TRANSACTIONS

of the

American Foundrymen's Society

Proceedings of the

Sixtieth Annual Meeting

Atlantic City, N. J.

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JOS. E. FOSTER
Editor

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OFFICERS AND DIRECTORS

of the

AMERICAN FOUNDRYMEN'S SOCIETY
(INCORPORATED)

(Year Ending May 10, 1957)

President, FRANK W. SHIPLEY, Caterpillar Tractor Co., Peoria, Ill.
Vice-President, HARRY W. DIETERT, Harry W. Dietert Co., Detroit, Mich.

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(In Addition to the President and Vice-President)

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G. P. PHILLIPS, International Harvester Co., Chicago.
A. PIRRIE, Standard Sanitary & Dominion Radiator, Ltd., Toronto, Ontario, Canada.
R. V. RIGHTER, Central Foundry Div., General Motors Corp., Danville, Ill.
G. R. RUSK, Freeman Supply Co., Toledo, Ohio.

Staff Officer

General Manager, WM. W. MALONEY

*Regional Vice-President

SUMMARY OF PROCEEDINGS OF THE 60TH ANNUAL MEETING

The 60th Castings Congress and Show of the American Foundrymen's Society was held in Atlantic City, N. J., May 3-9, 1956. The Biennial Foundry and Allied Industries Show was staged by the Society in conjunction with the Convention. Exhibits were displayed in Convention Hall with 239 companies showing their products.

Over 13,000 foundrymen, metallurgists, and representatives from allied industries throughout the United States and 21 countries outside of continental United States attended the Convention. During the seven days of Convention over 100 timely technical papers and committee reports were presented at 54 sessions.

The Charles Edgar Hoyt Memorial Lecture was delivered by S. C. Massari, National Engineering Co., Chicago, on the subject "Marketing Your Product" which starts on page 1 of this volume.

A summary of the sessions held follows:

Thursday, May 3, 12:00 Noon

AUTHORS LUNCHEON

Thursday, May 3, 12:00 Noon

JOINT AFS-NON-FERROUS FOUNDRYMEN'S SOCIETY LUNCHEON

Presiding—W. A. Gluntz, Gluntz Brass & Aluminum Co., Cleveland, Ohio.

Panel Discussion: What the Castings Buyer Expects from the Foundryman—Geo. W. Partington, General Electric Co., Bloomfield, N. J. and E. J. Schwarz, DeLaval Steam Turbine Co., Trenton, N. J.

Thursday, May 3, 2:30 P.M.

GRAY IRON SESSION

Presiding—C. K. Donoho, American Cast Iron Pipe Co., Birmingham, Ala.

Vice-Chairman—W. R. Jennings, John Deere Waterloo Tractor Works of Deere Mfg. Co., Waterloo, Iowa.

Secretary—S. G. Johnson, Jr., International Harvester Co., Indianapolis, Ind.

Service Life of Iron Castings Can Be Affected by Their Thermal Conductivity—J. A. Davis, H. W. Deem, and H. W. Lownie, Jr., Battelle Memorial Institute, Columbus, Ohio.

The Stress-Strain Relationship for Gray Iron—E. M. Stein, Battelle Memorial Institute.

Cracking and Life of Ingot-Molds—J. Duflot, Institut de Recherches de Siderurgie, St. Germain-en-Laye, France.

Thursday, May 3, 2:30 P.M.

MALLEABLE SESSION

Presiding—W. D. McMillan, International Harvester Co., Chicago.

Vice-Chairman—Milton Tilley, National Malleable & Steel Castings Co., Cleveland.

Secretary—Robert Greenlee, Auto Specialties Manufacturing Co., St. Joseph, Mich.

Effects of Charge Proportions, Furnace Atmosphere Flow Rate, and Melt-Down Time on Properties of Malleable Iron. (Malleable Research Progress Report)—A. H. Zrimsek, American Steel Foundries, Chicago, E. H. Belter and R. W. Heine, University of Wisconsin, Madison, Wis.

Annealing on Five-Day Week Basis—R. V. Righter, Central Foundry Div., General Motors Corp., Danville, Ill.

Consideration of Factors Governing Selection of Modern Malleable Heat Treating Equipment. (Panel discussion by Controlled Annealing Committee—*Chairman:* L. R. Jenkins, Wagner Malleable Iron Co., Decatur, Ill.

Thursday, May 3, 2:30 P.M.

SAND SESSION

Presiding—Charles Williamson, Columbia Malleable Iron Co., Columbia, Pa.

Vice-Chairman—R. H. Olmsted, Whitehead Bros. Co., New York.

Secretary—D. C. Williams, Ohio State University, Columbus, Ohio.

Evaluating Cereal Core Binders—E. C. Zuppann and H. Putz, Oliver Corp., South Bend, Ind.

Some Considerations on the Tensile and Transverse Strength Testing of Shell Mold and Core Sands—P. J. Ahearn and F. Quigley, Watertown Arsenal, Watertown, Mass., and J. I. Bluhm and J. F. Wallace, Case Institute of Technology, Cleveland.

Some Observations on the Transverse Test of Elevated Temperatures for Molded Sand Mixtures—D. C. Williams, Ohio State U., Columbus, Ohio.

Thursday, May 3, 2:30 P.M.

BRASS AND BRONZE SESSION

Presiding—R. A. Colton, Federated Metals Div., American Smelting & Refining Co., Newark, N. J.

Vice-Chairman—H. C. Ahl, National Bearing Div., American Brake Shoe Co., Meadville, Pa.

Secretary—R. F. Schmidt, Ajax Metal Div., H. Kramer & Co., Philadelphia, Pa.

Fluxing and Deoxidation Treatment for Copper—M. G. Neu, Foundry Services (Overseas) Ltd., Birmingham, England, and J. E. Gotheridge, Foundry Services, Inc., Columbus, Ohio.

Gas Porosity in Oxygen-Free Copper Castings—D. V. Ragone, University of Michigan, J. K. Sprinkle, Virginia Polytechnic Institute, H. F. Taylor and C. M. Adams, Massachusetts Institute of Technology.

Thursday, May 3, 3:00 P.M.

AFS OFFICIAL LADIES' TEA

Thursday, May 3, 7:00 P.M.

NON-FERROUS FOUNDERS' SOCIETY ANNUAL DINNER

Thursday, May 3, 8:00 P.M.

GRAY IRON SHOP COURSE

Presiding—E. J. Burke, Hanna Furnace Corp., Buffalo, N. Y.

Vice-Chairman—A. S. Hedberg, Wells Mfg. Co., Skokie, Ill.

Subject—**Cupola Operation.**

Speakers—Sam F. Carter, American Cast Iron Pipe Co., Birmingham, Ala.—Basic Cupola Operation. M. H. Horton, Deere & Co., Moline, Ill.—Water-Cooled Cupola Operation. H. E. Henderson, Lynchburg Foundry Co., Lynchburg, Va.—Acid Cupola Operation with Water-Cooled Tuyeres and Hot Blast.

Friday, May 4, 7:30 A.M.

AUTHORS BREAKFAST

Friday, May 4, 9:00 A.M.

BRASS AND BRONZE SESSION

Presiding—W. H. Baer, Bureau of Ships, Navy Dept., Washington, D. C.

Vice-Chairman—G. H. Bradshaw, Navy Dept., Philadelphia, Pa.

Secretary—T. E. Gregory, Michigan Smelting & Refining Div., Bohn Aluminum & Brass Corp., Detroit, Mich.

Age-Hardening Characteristics of a Cast Alloy of Copper-5.8% Titanium—N. P. Hehner, H. McCurdy, and R. E. Edelman, Pitman-Dunn Laboratories, Frankford Arsenal, Philadelphia.

Application of Radiography in Manufacture of Bronze Castings—N. A. Kahn, S. Goldspiel, and R. R. Waltien, New York Naval Shipyard, Brooklyn, N. Y.

Friday, May 4, 9:00 A.M.

GRAY IRON SESSION

(Sponsored by Cupola Research Committee)

Presiding—Dr. A. E. Schuh, U. S. Pipe & Foundry Co., Burlington, N. J.

Vice-Chairman—R. A. Clark, Electro Metallurgical Co., Union Carbide & Carbon Corp., Detroit, Mich.

Hot-Blast Cupola Practice—W. L. Heinrichs, Fried. Krupp, Essen-Bredeney, Germany.

Development of Hot-Blast Cupola Melting Technique in Europe—Ernst Loebbecke, W. Strikfeldt & Co., Rhineland, Western Germany.

Friday, May 4, 9:00 A.M.

MALLEABLE SESSION

Presiding—C. F. Joseph, Central Foundry Div., General Motors Corp., Saginaw, Mich.

Vice-Chairman—J. E. Rehder, Canada Iron Foundries, Montreal, Que.

Secretary—L. R. Jenkins, Wagner Malleable Iron Co., Decatur, Ill.

Comparison of Liquid and Air-Quenched Pearlitic Malleable Irons. (Report of Pearlitic Malleable Committee)—*Chairman*—R. W. Heine, University of Wisconsin, Madison, Wis.

Welding of Castings. (Report of Joint AWS-AFS Committee on Welding Iron Castings)—

Definitions—G. A. Meyer, Jr., Malleable Founders' Society, Cleveland, Ohio.

Specifications for Filler Metal—H. V. Inskeep, Linde Air Products Co., Newark, N. J.

Test Block for Welding Cast Iron—W. W. Edens, Allis-Chalmers Co., Milwaukee, Wis.

The Nucleation of Graphite During the Decomposition of Cementite—E. G. Haney, Aluminum Co. of America, New Kensington, Pa., M. F. Hawkes, Naval Ammunition Dept., Oahu, T.H.

Friday, May 4, 9:00 A.M.

PATTERN SESSION

Presiding—W. L. Allan, Southern Precision Pattern Works, Inc., Birmingham, Ala.

Vice-Chairman—A. F. Pfeiffer, Allis-Chalmers Mfg. Co., West Allis, Wis.

Secretary—M. K. Young, United States Gypsum Co., Chicago, Ill.

Use of Epoxy Resin as Pattern Material—E. J. McAfee, Puget Sound Naval Shipyard, Bremerton, Wash.

Opportunities and New Developments of the Patternmaking Industry—M. K. Young, U. S. Gypsum Co., Chicago, Ill.

Friday, May 4, 12:00 Noon

JOINT MALLEABLE AND GRAY IRON ROUND TABLE LUNCHEON

Presiding—B. C. Yearley, National Malleable & Steel Castings Co., Cleveland, Ohio.

Vice-Chairman—C. K. Donoho, American Cast Iron Pipe Co., Birmingham, Ala.

Malleable Base Spheroidal Iron—F. B. Rote, E. J. Chojnowski, and J. F. Bryce, Albion Malleable Iron Co., Albion, Mich.

Friday, May 4, 12:00 Noon

BRASS AND BRONZE ROUND TABLE LUNCHEON

Presiding—B. A. Miller, Crown Non-Ferrous Foundry, Inc., Chester, Pa.

Vice-Chairman—H. L. Smith, Federated Metals Div., American Smelting & Refining Co., Whiting, Ind.

Subject—**New Horizons in Copper Base Alloy Founding.**

Friday, May 4, 12:00 Noon

EDUCATION DIVISION ROUND TABLE LUNCHEON

Presiding—Dr. W. H. Ruten, Polytechnic Institute of Brooklyn, N. Y.

Vice-Chairman—Frank W. Shipley, Caterpillar Tractor Co., Peoria, Ill.

Subject—**Tomorrow's Foundryman.**

Speaker—L. J. Fletcher, Caterpillar Tractor Co., Peoria, Ill.

Friday, May 4, 2:30 P.M.

LIGHT METALS SESSION

Presiding—B. L. Meredith, Federated Metals Div., American Smelting & Refining Co., South Painfield, N. J.

Vice-Chairman—W. Babington, Bell Telephone Laboratory, Murray Hill, N. J.

Centrifugal Casting of Unusual Shapes in Non-Ferrous Alloys—J. P. Krishon, Centr-O-Cast & Engineering Co., Detroit, Mich.

Machinability of Aluminum Die Castings—D. L. Colwell and Oldrich Tichy, Apex Smelting Co., Cleveland, Ohio.

Novel Aluminum Engine Die Castings—E. N. Jacobi, Briggs & Stratton Corp., Milwaukee.

Friday, May 4, 2:30 P.M.

PLANT AND PLANT EQUIPMENT SESSION

Presiding—James Thomson, Continental Foundry & Machine Div., Blaw-Knox Co., East Chicago, Ind.

Vice-Chairman—H. C. Weimer, Beardsley-Piper Div., Pettibone-Mulliken Corp., Chicago, Ill.

Method of Melting Malleable Iron with Combination of Pulverized Coal and Oil—E. H. Nielsen, The Whiting Corp., Harvey, Ill.

Recent Development of Coreless Line Frequency Induction Melting Furnace in European Foundries—Otto Junker, Otto Junker GMBH, Lammersdorf, Aachen, Germany. (Presented by Manuel Tama, Ajax Engineering Corp., Trenton, N. J.)

Friday, May 4, 2:30 P.M.

SAND SESSION

Presiding—E. C. Troy, Riverton, N. J.

Vice-Chairman—W. R. Moggridge, Ford Motor Co., Windsor, Ont.

Secretary—H. F. Bishop, Naval Research Laboratory, Washington, D. C.

Molding Sands, Molding Methods, and Casting Dimensions—R. W. Heine, University of Wisconsin.

Does Sand Testing Give Us the Facts?—R. W. Heine, University of Wisconsin, E. H. King and J. S. Schumacher, Hill & Griffith Co., Cincinnati, Ohio.

Jolt Test for Sand—R. W. Heine, University of Wisconsin, E. H. King and J. S. Schumacher, Hill & Griffin Co.

Friday, May 4, 2:30 P.M.

GRAY IRON SHOP COURSE

Presiding—K. H. Priestley, Vassar Electroloy Products Co., Vassar, Mich.

Vice-Chairman—L. L. Clark, Armour Research Foundation, Chicago, Ill.

Subject—**Selection of Materials for Gray Iron Melting.**

Speaker—R. A. Clark, Electro Metallurgical Co., Div. Union Carbide & Carbon Co., Detroit, Mich.

Friday, May 4, 7:30 P.M.

AFS ANNUAL DINNER

Presiding—AFS President B. L. Simpson.

The AFS Annual Dinner of the Society's 60th Annual Meeting was called to order by President Bruce L. Simpson, Following singing of the National Anthem by those in attendance, President Simpson introduced Past President Fred J. Walls, Chairman of the AFS Board of Awards who presented Honorary Life Memberships and AFS Gold Medal Awards as follows:

HONORARY LIFE MEMBERSHIPS IN AFS

Awarded to Joseph C. Pendleton, retired from the Newport News Shipbuilding & Dry Dock Co., Newport News, Va., *"For outstanding contributions to the advancement of the art and sciences of metal castings, particularly in the field of naval architecture requirements."*

Awarded to William D. McMillan, Plant Metallurgist, International Harvester Co., McCormick Works, Chicago, *"For outstanding contributions to the Society and the ferrous castings industry."*

Awarded to Bruce L. Simpson, President, National Engineering Co., Chicago, *"On completion of his present term of office as President of AFS."*

THE THOMAS W. PANGBORN GOLD MEDAL

Awarded to Charles C. Sigerfoos, Foundry Instructor, Michigan State University, E. Lansing, Mich., *"For outstanding contributions to the Society and the Castings Industry, particularly in the field of education of engineering students for foundry careers."*

THE WILLIAM H. McFADDEN GOLD MEDAL

Awarded to James S. Vanick, Research and Development Metallurgist, The International Nickel Co., Inc., New York, *"For outstanding contributions to the Society and for valuable service to the ferrous castings industry over a period of many years."*

THE JOSEPH S. SEAMAN GOLD MEDAL

Awarded to Harold F. Bishop, Metallurgist, Naval Research Laboratories, Washington, D. C., *"For outstanding contributions in the field of castings research at the Naval Research Laboratories and for contributions to the Society."*

After presentation of awards, President Simpson introduced the guest speaker of the evening, Dr. Norman Vincent Peale, Minister, Marble Collegiate Church, New York, who spoke on the theme "Efficiency Through Right Thinking."

Saturday, May 5, 7:30 A.M.

AUTHORS BREAKFAST

Saturday, May 5, 9:00 A.M.

EDUCATION SESSION

Presiding—F. G. Sefing, International Nickel Co., New York.
Vice-Chairman—B. C. Yearley, National Malleable & Steel Castings Co., Cleveland, Ohio.
Secretary—C. J. Turner, Caterpillar Tractor Co., Peoria, Ill.

Subject—**Training Needs for the Foundry Industry.**

Speaker—John S. McCauley, Research Director, U. S. Dept. of Labor, Washington, D. C.

Saturday, May 5, 9:00 A.M.

REFRACTORIES SESSION

Presiding—W. R. Jaeschke, Whiting Corp., Harvey, Ill.
Vice-Chairman—J. P. Holt, Basic Refractories, Inc., Clayton, Mo.

Properties of Refractories Affect Air Furnace Bottom Service—L. R. Jenkins and C. C. Lawson, Wagner Malleable Iron Co., Decatur, Ill.

Experiences with Basic Cupola Refractories and Melting—A. P. Alexander and Edward Boywid, International Harvester Co., Memphis, Tenn.

Saturday, May 5, 9:00 A.M.

LIGHT METALS SESSION

Presiding—E. V. Blackmun, Aluminum Co. of America, Pittsburgh, Pa.
Vice-Chairman—F. L. Turk, Joy Manufacturing Co., New Philadelphia, Ohio.

Tensile Properties of Aluminum Silicon Magnesium Alloys and the Effects of Sodium Modification—R. C. Harris, S. Lipson, and H. Rosenthal, Pitman-Dunn Laboratories, Frankford Arsenal, Philadelphia, Pa.

An Evaluation of ZH62XA Magnesium Sand Casting Alloy—K. E. Nelson, The Dow Chemical Co., Midland, Mich., and W. P. Saunders, The Dow Chemical Co., Madison, Ill.

Effects of Section Size Variations in a Test Casting on Properties of Some Mg-Al-Zn Alloys—W. E. Pearson, The Dow Chemical Co., Midland, Mich.

Saturday, May 5, 9:00 A.M.

SAND SESSION

Presiding—J. O. Ochsner, Crouse-Hinds Co., Syracuse, N. Y.
Vice-Chairman—E. C. Zirzow, Werner G. Smith, Inc., Cleveland, Ohio.
Secretary—D. J. Jones, New Jersey Silica Sand Co., Millville, N. J.

Cold Process for Resin-Coated Foundry Sands—J. E. Bolt, General Electric Co., Pittsfield, Mass.

Developments of High Pressure Molding with CO_2 Process Sands—T. E. Barlow, Eastern Clay Products Dept., International Minerals & Chemical Corp., Chicago, Ill.

Factors Affecting Core Stickiness. (Progress Report of Sand Division Committee 8-c-f)—

Saturday, May 5, 12:00 Noon

AFS ANNUAL BUSINESS MEETING

Presiding—AFS National President Bruce L. Simpson, National Engineering Co., Chicago.

President Simpson called the meeting to order as the Annual Business meeting of the American Foundrymen's Society. President Simpson, after making some introductory remarks, called on General Manager Wm. W. Maloney who announced the 1956 AFS Robert E. Kennedy Apprentice Contest winners as follows:

Wood Patternmaking Division

1st—Herbert Nelson, Pacific Pattern Works, Tacoma, Washington.
2nd—Edward Poole, Washington Iron Works, Seattle, Washington.
3rd—Lewis Backus, Cleveland Standard Pattern Works, Cleveland.

Metal Patternmaking Division

1st—Herbert L. Zigler, Motor Patterns Co., Cleveland.
2nd—Stanley Stobierski, Cove Pattern Works, Cleveland.
3rd—James A. Smith, Annex Pattern Co., Detroit.

Gray Iron Molding Division

1st—Stanley Varr, Brown & Sharpe Mfg. Co., Providence, R. I.
2nd Eugene Calhoun, Howell Foundry Co., Inc., Los Nietos, Calif.
3rd—James R. Walton, Caterpillar Tractor Co., Peoria, Ill.

Steel Molding Division

1st—Mert West, The Ohio Steel Foundry Co., Springfield, Ohio.
2nd—Joseph Zohil, Dodge Steel Co., Philadelphia.
3rd—Jesse Lawson, Blaw-Knox Co., Continental Foundry Div., East Chicago, Ind.

Non-Ferrous Molding Division

1st—Donald Tetzlaff, Allis Chalmers, Mfg. Co., Milwaukee, Wis.
2nd—Raymond Montano, Ampco Metal, Inc., Milwaukee, Wis.
3rd—Arthur Berge, The St. Catharine Brass Works, Ltd., St. Catharines, Ontario, Canada.

The Society arranged to have the five first-prize winners present at the Convention and to receive their awards in person. The first-prize winners were called to the platform and president Simpson presented each with the first prize, a check for $100.00 and an engraved Certificate of Award prefaced by a few words of encouragement and commendation.

President Simpson then called on General Manager Maloney who reported on the nomination of Officers and Directors for the coming year and stated that no additional nominees had been received in accordance with the procedure prescribed in Art. XI of the Society By-Laws. He therefore cast a unanimous ballot of the membership of AFS for the election of the following:

President (to serve one year):
Frank W. Shipley, Caterpillar Tractor Co., Peoria, Ill.

Vice President (to serve one year):
Harry W. Dietert, Harry W. Dietert Co., Detroit, Mich.

Directors (to serve three years):
R. W. Griswold, Erie Malleable Iron Co., Erie, Pa.
Herbert Heaton, Letson & Burpee, Ltd., Vancouver, B. C.
A. V. Martens, Pekin Foundry & Mfg. Co., Pekin, Ill.
G. P. Phillips, International Harvester Co., Chicago.
A. W. Pirrie, Standard Sanitary & Dominion Radiator, Ltd., Toronto, Ontario, Canada.
R. V. Righter, Central Foundry Div., General Motors Corp., Danville, Ill.
G. R. Rusk, Freeman Supply Co., Toledo, Ohio.

Director (to serve one year):
B. L. Simpson, National Engineering Co., Chicago.

Saturday, May 5, 2:00 P.M.

CHARLES EDGAR HOYT MEMORIAL LECTURE

Subject—**Marketing Your Product.**

Speaker—S. C. Massari, National Engineering Co., Chicago.

Saturday, May 5, 3:00 P.M.

BRASS AND BRONZE SESSION

Presiding—R. B. Fischer, Ingersoll-Rand Co., Phillipsburg, N. J.
Secretary—J. E. Gotheridge, Foundry Services, Inc., Columbus, Ohio.

Casting Defects Clinic.

Saturday, May 5, 3:00 P.M.

INDUSTRIAL ENGINEERING SESSION

Presiding—M. T. Sell, Sterling Foundry Co., Wellington, Ohio.
Vice-Chairman—Milton Annich, American Brake Shoe Co., Mahwah, N. J.

Practical Foundry Application of Statistical Quality Control—Ross Martin, Jr., Rich Mfg. Co. of Calif., Los Angeles, Calif.

Engineering Cost, and Quality—J. M. Barrabee, International Harvester Co., Milwaukee, Wis.

Saturday, May 5, 3:00 P.M.

MALLEABLE SHOP COURSE

Presiding—F. B. Rote, Albion Mallable Iron Co., Albion, Mich.

Molding Methods for Malleable Iron Green Sand Molding—W. Tragarz, International Harvester Co., Chicago, Ill.

Dry Core Sand Method—R. H. Greenlee, Auto Specialties Mfg. Co., St. Joseph, Mich.

Mold Blowing—Medie Hakeman, John Deere Malleable Works, East Moline, Ill.

High Pressure Molding—T. E. Barlow, Eastern Clay Products Dept., International Minerals & Chemical Co., Chicago, Ill.

Shell Molding—C. F. Joseph, Central Foundry Div., General Motors Corp., Saginaw, Mich.

Saturday, May 5, 5:30 P.M.

INTERNATIONAL REUNION

(For all attending from countries outside U. S. and Canada, as Guests of AFS. Ladies welcome)

Saturday, May 5, 7:00 P.M.

SAND DIVISION DINNER

Presiding—O. Jay Myers, Archer-Daniels-Midland Co., Minneapolis, Minn.
Subject—**New Developments in Core Blowing.**
Speaker—Arthur M. Clark, Ford Motor Co., Dearborn, Mich.

Sunday, May 6, 12:30 P.M.

AFS PAST PRESIDENTS' LUNCHEON

Sunday, May 6, 7:30 P.M.

CANADIAN DINNER

("Fellowship Hour" 6:30 - 7:30)

Monday, May 7, 7:30 A.M.

AUTHORS BREAKFAST

Monday, May 7, 9:00 A.M.

LIGHT METALS SESSION

Presiding—A. G. Ealy, Littlestown Hardware & Foundry Co., Littlestown, Pa.
Vice-Chairman—C. Szego, Mold Cast Products, Newark, N. J.

Modifications in Vertical Gating Principles—K. R. Grube, R. M. Lang, and J. G. Kura, Battelle Memorial Institute, Columbus, Ohio.

New Aluminum Permanent Mold Casting Alloys C355 and A356—R. C. Lemon and H. Y. Hunsicker, Aluminum Co. of America, Cleveland, Ohio.

New Aluminum Casting Alloy XA140 for Elevated Temperature Applications—R. C. Lemon and W. E. Sicha, Aluminum Co. of America, Cleveland, Ohio.

Monday, May 7, 9:00 A.M.

STEEL SESSION

Presiding—C. K. Donoho, American Cast Iron Pipe Co., Birmingham, Ala.
Vice-Chairman—W. O. Igelman, National Malleable & Steel Castings Co., Melrose Park, Ill.
Secretary—D. N. Rosenblatt, American Foundry & Machine Co., Salt Lake City, Utah.

Effects of Aluminum and Titanium Deoxidation on Cast Steel—R. D. Ahles, W. R. Nestle, C. H. Weight, and R. S. Zeno, General Electric Co., Schenectady, N. Y.

Charpy Impact Tests Reveal Mysteries of Hot Tears—C. F. Christopher, Continental Foundry & Machine Co., Division of Blaw-Knox Co., East Chicago, Ind.

Some Aspects of Dust Suppression in Foundries. (Exchange Paper)—C. M. Stoch, British Steel Castings Research Association, Sheffield, England.

Monday, May 7, 9:00 A.M.

SAND SESSION

Presiding—R. H. Jacoby, St. Louis Coke & Foundry Co., St. Louis, Mo.
Vice-Chairman—T. A. Tarquino, American Brake Shoe Co., Chicago Heights, Ill.
Secretary—J. B. Stazinski, Foundry Div., General Electric Co., Everett, Mass.

Green Sand Casting Finish—R. F. Meader, Whitin Machine Works, Whitinsville, Mass.

Statistical Techniques for Classifying Foundry Sands— D. C. Ekey and John Leaman, Lebanon Steel Foundry, Lebanon, Pa.

Analysis of Factors Affecting Surface Finish of Gray Iron Castings—D. C. Ekey and R. L. Yard, Lebanon Steel Foundry, Lebanon, Pa.

Monday, May 7, 9:00 A.M.

JOINT PLANT AND PLANT EQUIPMENT AND AIR POLLUTION SESSION

Presiding—James Thomson, Continental Foundry & Machine Div., Blaw-Knox Co., East Chicago, Ind.
Vice-Chairman—H. W. Johnson, Wells Mfg. Co., Skokie, Ill.

Equipment For Cupola-Emission Control—John M. Kane, American Air Filter Co., Louisville, Ky.

Control of Make-Up Air in Industrial Plants. (Slide Demonstration)—K. E. Robinson, General Motors Corp., Detroit, Mich.

Monday, May 7, 12:00 noon

LIGHT METALS ROUND TABLE LUNCHEON

Presiding—D. L. LaVelle, Federated Metals Div., American Smelting & Refining Co., Barber, N. J.
Vice-Chairman—C. M. Adams, Dept. of Metallurgy, Massachusetts Institute of Technology, Cambridge, Mass.
Subject—**Gas in Light Alloys.**
Panel Members and Topics:

Methods and Apparatus for Measuring Gas Content of Melts—H. Rosenthal, Pitman-Dunn Laboratories, Frankford Arsenal, Philadelphia, Pa.

Application of Degassing Methods to Production—W. A. Mader, Oberdorfer Foundries, Inc., Syracuse, N. Y.

Production Results and Casting Quality—B. J. Alperin, Erie Foundries, General Electric Co., Erie, Pa.

A Tensile Bar Shell Mold for Light Alloys—K. L. Herrick and R. C. Harris, Pitman-Dunn Laboratories, Frankford Arsenal, Philadelphia, Pa.

Monday, May 7, 12:00 noon

AFS BOARD OF DIRECTORS LUNCHEON AND BUSINESS MEETING

Monday, May 7, 2:30 P.M.

HEAT TRANSFER SESSION

Presiding—Charles Locke, Westlectric Castings, Inc., Los Angeles, Calif.
Vice-Chairman—Dr. W. J. Childs, Lafayette College, Easton, Pa.

Temperature Drop in Pouring Ladles—V. Paschkis, Columbia University, N. Y. (Progress Report on AFS-sponsored Research Work).

The Risering of Nodular Iron. Part III—The Effect of Pouring Temperature on Shrink Depth—R. C. Shnay and S. L. Gertsman, Dept. of Mines & Technical Surveys, Ottawa, Ont., Canada.

Monday, May 7, 2:30 P.M.

GRAY IRON SESSION

Presiding—W. W. Levi, Lynchburg Foundry Co., Lynchburg, Va.
Vice-Chairman—E. J. Burke, Hanna Furnace Corp., Buffalo, N. Y.
Secretary—C. F. Walton, Gray Iron Founders' Society, Cleveland, Ohio.

Influence of Temperature on Mechanical Strength of Coke—John Varga, Jr. and H. W. Lownie, Jr., Battelle Memorial Institute, Columbus, Ohio.

Use of Pig Iron in Iron Foundries (with Particular Attention to Specifications, Prices, Secondary Elements and "Herdity")—H. W. Lownie, Jr., Battelle Memorial Institute, Columbus, Ohio.

Compressive Strength of Coke—D. E. Krause and E. A. Lange, Gray Iron Research Institute, Columbus, Ohio.

Monday, May 7, 2:30 P.M.

SAND SESSION

Presiding—Douglas James, Urick Foundry, Erie Pa.

Vice-Chairman—L. J. Pedicini, General Motors Corp., Detroit, Mich.

Secretary, William Ruprecht, ACF Industries, Inc., Berwick, Pa.

The Carbon Dioxide Process (Official Exchange Paper of The Institute of British Foundrymen)—D. V. Atterton, Foundry Services, Ltd., Birmingham, England.

Coremaking with CO_2 Process—F. M. Scaggs, Oklahoma Steel Castings Co., Tulsa, Okla.

Progress Report of Core Test Committee (8-C).

Monday, May 7, 2:30 P.M.

EDUCATION SESSION

Presiding—E. M. Strick, Erie Malleable Iron Co., Erie, Pa.

Vice-Chairman—Jess Toth, Harry W. Dietert Co., Detroit, Mich.

Subject—**Chapter Educational Activities.**

Speaker—H. E. Gravlin, Claude B. Schneible Co., Detroit, Mich.

Panel Discussion—Responsibilities of Chapter Educational Chairmen—Robert Herrmann, Penton Publishing Co., Cleveland, Ohio. J. A. Barrett, National Malleable & Steel Castings Co., Indianapolis, Ind. R. W. Schroeder, University of Illinois, Navy Pier Branch, Chicago, Ill. Morris McQuiggan, LaSalle Coke Co., Montreal, Que., Canada.

Monday, May 7, 7:30 P.M.

AFS ALUMNI DINNER

Monday, May 7, 8:00 P.M.

METALLOGRAPHY FOR THE FOUNDRYMAN (PART I).

Presiding—P. C. Rosenthal, University of Wisconsin, Madison, Wisconsin.

Vice-Chairman—H. J. Heine, American Foundrymen's Society.

Good Metallographic Practice—A. M. Montgomery, Aluminum Co. of America, Cleveland, Ohio.

Metallography of Aluminum and Its Alloys—L. F. Mondolfo, Director, Metallurgical Engineering Dept. Illinois Institute of Technology, Chicago.

Metallography of Magnesium and Its Alloys—G. E. Holdeman, Dow Chemical Co., Midland, Mich.

Monday, May 7, 8:00 P.M.

SAND SHOP COURSE

Presiding—R. L. Cleland, Great Lakes Foundry Sand Co., Detroit, Mich.

Moderator—C. A. Sanders, American Colloid Co., Chicago, Ill.

Subject—"What Sand Mixture Shall I Use?"

Panel Members—H. W. Meyer, General Steel Castings Corp., Granite City, Ill. R. F. Meader, Whitin Machine Works, Whitinsville, Mass. W. G. Parker, General Electric Co., Elmira, N. Y.

Tuesday, May 8, 7:30 A.M.

AUTHORS BREAKFAST

Tuesday, May 8, 9:00 A.M.

HEAT TRANSFER SESSION

Presiding—C. W. Briggs, Steel Founders' Society of America, Cleveland, Ohio.

Vice-Chairman—W. K. Bock, National Malleable & Steel Castings Co., Cleveland, Ohio.

Heat Transfer Fundamentals.

Heat Flow in the Transient State—V. Paschkis, Columbia University, New York, N. Y.

Solidification Phenomena—W. S. Pellini, Naval Research Laboratory, Washington, D. C. (Presented by) H. F. Bishop, Naval Research Laboratory, Washington, D. C.

Heat Treating Fundamentals—W. J. Childs, Lafayette College, Easton, Pa.

Tuesday, May 8, 9:00 A.M.

INDUSTRIAL ENGINEERING SESSION

Presiding—J. A. Westover, Westover Engineers, Milwaukee, Wis.

Vice-Chairman—R. A. Petersen, Lynchburg Foundry, Lynchburg, Va.

Basic Cost Concepts for the Small-Medium Foundry—A. C. Sinnett, Terre Haute Malleable & Mfg. Co., Terre Haute, Ind.

Incentives in a Small Jobbing Foundry—C. C. Erhart, Jr., Erhart Foundry & Machine Co., Cincinnati, Ohio.

Tuesday, May 8, 9:00 A.M.

PATTERN SESSION

Presiding—J. W. Costello, American Hoist & Derrick Co., St. Paul, Minn.

Vice-Chairman—G. E. Garvey, Garvey Pattern & Mfg. Co., South Bend, Ind.

Secretary—John F. Roth, Cleveland Standard Pattern Works, Inc., Cleveland, Ohio.

Subject—**Metal Patterns and Core Boxes with Metal Dryers.**

Speaker—O. C. Bueg, Arrow Pattern & Engineering Co., Erie, Pa.

Tuesday, May 8, 9:00 A.M.

GRAY IRON SESSION

Presiding—H. W. Lownie, Jr., Battelle Memorial Institute, Columbus, Ohio.

Vice-Chairman—G. A. Timmons, Climax Molybdenum Co., Detroit, Mich.

Subsurface Blowholes in Gray Irons and Their Association with Manganese Sulphide Segregation—W. G. Tonks, British Cast Iron Research Association, Birmingham, England.

Heat Treatment of Gray Cast Iron—J. B. Peck and A. H. Rauch, Deere & Co., Moline, Ill.

Causes and Effects of the Grain Size in Pearlitic Gray Irons. (Official French Exchange Paper from Association Technique de Fonderie, Paris, France)—Michael Ferry and Jean-Claude Margerie, Centre Technique des Industries de la Fonderie, Paris, France (Presented by) J. S. Vanick International Nickel Co., New York, N. Y.

Tuesday, May 8, 12:00 noon

NOISE CONTROL ROUND TABLE LUNCHEON

Presiding—Allen Brandt, Bethlehem Steel Co., Bethlehem, Pa.

Vice-Chairman—K. M. Smith, Caterpillar Tractor Co., Peoria, Ill.

Subject—**Legal Aspects of the Noise Problem.**

Speaker—F. E. Frazier, National Association of Mutual Casualty Companies, Chicago, Ill.

Tuesday, May 8, 12:00 noon

PATTERN ROUND TABLE LUNCHEON

Presiding—O. C. Bueg, Arrow Pattern & Engineering Co., Erie, Pa.

Vice-Chairman—Joseph M. Kreiner, National Malleable & Steel Castings Co., Cleveland, Ohio.

Secretary—E. T. Kindt, Kindt-Collins Co., Cleveland, Ohio.

Subject—**Patternmaking—Man to Man.**

Speaker—Dr. Ralph L. Lee, Lee Hobby Foundry, Birmingham, Mich.

Tuesday, May 8, 12:00 noon

INDUSTRIAL ENGINEERING ROUND TABLE LUNCHEON

Presiding—W. E. Boswell, Glamorgan Pipe & Foundry Co., Lynchburg, Va.

Vice-Chairman—E. C. Reid, Ford Motor Co., Windsor, Ont.

Subject—**The Human Factor in New Equipment.**

Speaker—John Taylor, Norris & Elliott, Inc., Columbus, Ohio.

Tuesday, May 8, 12:00 noon

PRESIDENT'S LUNCHEON FOR EXHIBITS COMMITTEE

Tuesday, May 8, 2:30 P.M.

FUNDAMENTAL PAPERS SESSION

Presiding—H. F. Taylor, Massachusetts Institute of Technology, Cambridge, Mass.

Vice-Chairman—W. A. Snyder, University of Washington, Seattle, Wash.

Metal Solidification in a Flowing Stream—M. C. Flemings and C. M. Adams, Massachusetts Institute of Technology, Cambridge, Mass.; E. E. Hucke, University of Michigan, Ann Arbor, Mich.

A New Method for Determining the Effect of Solidification Range of Fluidity—D. V. Ragone, University of Michigan; C. M. Adams and H. F. Taylor, M. I. T.

Tuesday, May 8, 2:30 P.M.

STEEL SESSION

Presiding—A. J. Kiesler, General Electric Co., Schenectady, N. Y.
Vice-Chairman—E. W. O'Brien, Oklahoma Steel Castings Co., Tulsa, Okla.

Pad Washing with Carbon Arc Process—F. Newberry, Oklahoma Steel Castings Co., Tulsa, Okla.

Inorganic Binders Solve Shell Molding Problems—Jose Navarro and H. F. Taylor, Massachusetts Institute of Technology, Cambridge, Mass.

Anionic of Cationic Agents—A Solution to Sand Problems—G. J. Vingas and A. H. Lewis, Dominion Engineering Works, Montreal, Que., Canada.

Tuesday, May 8, 2:30 P.M.

NOISE CONTROL SESSION

(Sponsored by Safety, Hygiene and Air Pollution Control Committee).
Presiding—H. T. Walworth, Lumbermen's Mutual Casualty Co., Chicago, Ill.
Vice-Chairman—J. C. Radcliffe, Ford Motor Co., Dearborn, Mich.

Foundry Noise Control—A Demonstration—F. A. Patty, General Motors Corp., Detroit, Mich.

Typical Foundry Noise Exposures—Edward G. Meiter, Employers Mutual Liability Insurance Co. of Wisconsin, Milwaukee, Wis.

The Mechanics of Hearing and Hearing Loss in Noise Deafness—Abraham S. Goldner, M.D., Flushing, N. Y.

Panel Discussion—Short Case Histories in Foundry Noise Control—A. L. Cudworth, Liberty Mutual Insurance Co., James Botsford, Bethlehem Steel Co., Bethlehem, Pa. E. L. Lund, and W. O. Hanson, Allis-Chalmers Mfg. Co., LaCrosse, Wis.

Tuesday, May 8, 7:00 P.M.

MANAGEMENT DINNER

Presiding—R. L. McIlvaine, National Engineering Co., Chicago, Ill.

Subject—**People—The Other Side of the Air Pollution Problem.**

Speaker—Dr. E. M. Adams, Dow Chemical Co., Midland, Mich.

Tuesday, May 8, 8:00 P.M.

METALLOGRAPHY FOR THE FOUNDRYMAN (PART II).

Presiding—P. C. Rosenthal, University of Wisconsin. Madison, Wis.
Vice-Chairman—R. L. Lindsay, Pennsylvania State University, University Park, Pa.

Ferrous Metallographic Practice—P. C. Rosenthal, University of Wisconsin, Madison, Wis.

Metallography of Copper and Nickel-Base Alloys and Stainless Steels—J. J. Connelly, American Brake Shoe Co., Mahwah, N. J.

Management of a Metallographic Laboratory—Walter W. Edens, Allis-Chalmers Mfg. Co., Milwaukee, Wis.

Tuesday, May 8, 8:00 P.M.

SAND SHOP COURSE

Presiding—E. C. Troy, Foundry Engineer, Riverton, N. J.
Moderator—C. A. Sanders, American Colloid Co., Chicago, Ill.

Subject—**Eliminating Sand Pile Blues.**

Panel Members—Ray Olson, Shell Process Inc., Chicopee, Mass., Charles W. Mooney, Jr., Olney Foundry Div., Link Belt Co., Philadelphia, Pa., Edward H. Berry, Dodge Steel Co., Philadelphia, Pa.

Wednesday, May 9, 7:30 A.M.

AUTHORS BREAKFAST

Wednesday, May 9, 9:00 A.M.

JOINT GRAY IRON AND PLANT AND PLANT EQUIPMENT SESSION

Presiding—James Thomson, Continental Foundry & Machine Div., Blaw-Knox Co., East Chicago, Ind.
Vice-Chairman—K. M. Smith, Caterpillar Tractor Co., Peoria, Ill.

Casting Tolerances As Affected by Automation in the Machine Shop—E. L. Buchman, Cleveland Foundry, Ford Motor Co., Beria, Ohio.

Blow-Hot Press Automatic Shell Molding Machine—R. S. Amala, J. H. Smith, A. L. Boegehold, and R. F. Thomson, General Motors Corp., Detroit, Mich.

Movie—**Techniques for Tomorrow.** (Various Automation Innovations in Ford's Cleveland Foundry.)

Wednesday, May 9, 9:00 A.M.

STEEL SESSION

Presiding—Erwin C. Tetzlaff, Pelton Steel Castings Co., Milwaukee, Wis.
Vice-Chairman—E. C. Troy, Foundry Engineer, Riverton, N. J.

Bentonite—Properties and Composition (Their Relation to Casting Defects)—Wm. D. Emmett, Los Angeles Steel Castings Co., Los Angeles, Calif. (Presented by B. G. Emmett, Los Angeles Steel Castings Co., Los Angeles, Calif.)

Removal of Gases from Molten Steel—Clyde H. Wyman, Burnside Steel Foundry Co., Chicago, Ill.

Ceramic-Mold Process for Steel Castings—D. C. Ekey and E. G. Vogel, Lebanon Steel Foundry, Lebanon, Pa.

Wednesday, May 9, 9:00 A.M.

SAND SESSION

Presiding—D. F. Sawtelle, Malleable Iron Fittings Co., Branford, Conn.
Vice-Chairman—B. H. Booth, Carpenter Bros., Inc., Milwaukee, Wis.
Secretary—L. E. Wile, Lynchburg Foundry Co., Lynchburg, Va.

Influence of Sand Distribution and Surface Coatings on Metal Penetration—Report by Sand Division Mold Surface Committee—8-H.

Casting Finish, Precision and Tolerance—C. A. Sanders, American Colloid Co., Chicago, Ill.

Wednesday, May 9, 9:00 A.M.

VENTILATION AND SAFETY SESSION

Presiding—J. R. Allan, Allan Industries, Melbourne, Fla.

Subject—**Foundry Ventilation.**

Speaker—J. G. Liskow, American Air Filter Co., Louisville, Ky.

Subject—**Aluminum Therapy.**

Speaker—Dr. J. W. G. Hannon, McIntyre Research Foundation, Washington, Pa.

English Motion Pictures—**Dust Control on Pedestal Grinders and Dust Control on Swing Frame Grinders**—C. M. Stoch, British Steel Castings Research Association, Sheffield, England.

Subject—**AFS Safety Training Course.**

Speaker—R. A. Oster, National Director, Beloit Vocational and Adult School, Beloit, Wis.

Wednesday, May 9, 12:00 Noon

STEEL ROUND TABLE LUNCHEON

Presiding—Dale L. Hall, Oklahoma Steel Castings Co., Tulsa, Okla.
Vice-Chairman—P. C. Fuerst, Falk Corp., Milwaukee, Wis.

Subject—**What's New With Snotters—and Other Steel Foundry Problems.**

Wednesday, May 9, 5:30 P.M.

60TH CASTINGS CONGRESS AND EXHIBIT OFFICIALLY CLOSED

REPORT OF THE AFS TREASURER

(Fiscal Year Ended June 30, 1956)

The Auditor's Report for 1955-56, as prepared by the firm of George V. Rountree & Co., Chicago, certified public accountants, contained the following data on the Society's financial condition as of June 30, 1956:

Total Income—Actual $937,833, Budgeted $894,600; Total Expense—Actual $822,667, Budgeted $765,275; Excess Income—Actual $115,165, Budgeted $129,325.

General Observations

Total Income in 1955-56 increased 78.6% or $412,597 over 1954-55, while Expense increased 27.4% or $176,815. The increases are accounted for as follows:

Income Increases		Expense Increases	
M/S Dues	$ 28,300.	Salaries	$ 25,100.
All Publications	12,100.	Burdens	5,700.
"Modern Castings"	39,700.	"Modern Castings"	54,900.
Convention & Exhibit	318,900.	Convention & Exhibit	56,800.
Investments	14,400.	Technical	23,600.
		Chapters	12,400.
Total Increase	$413,400.	Total Increase	$178,500.

Since the above compares an Exhibit year and a non-Exhibit year, comparison should be made of two successive 2-year periods, as follows:

	1952-1954	1954-1956
Total Income	$1,344,921.	$1,463,069.
Total Expense	1,232,989.	1,468,519.
Excess Income or (Expense) 2 Years	$ 111,932.	($ 5,450.)

The above figures re-emphasize the point made in 1955 that the upward trend of expense-producing activities is greater than for income-producing activities, and that the Society must develop additional sources of substantial income (such as the Engineered Castings Show), or reduce the rate of activities and expense to match present factual income.

Attention is directed to the Auditor's Report for the year, showing that only three activities (Dues, Exhibit, Investments) produced substantial Net Income in 1955-56, whereas six activities (Magazine, Technical, Chapters, Publications, Conventions, Gen. Administration) produced substantial Net Expense.

Income and Expense 1955-56

The relative importance of each of the Society's income sources is illustrated in the following table, showing shifts in income emphasis between three successive two-year periods:

Income Sources (%) 1952-1956

	1952-1954	1953-1955	1954-1956
Member Dues	35.1%	36.6%	38.7%
Advertising	33.1	31.7	29.7
Conventions & Exhibits	24.8	23.9	22.3
Publications	6.2	5.8	6.1
Investments and Miscellaneous	0.8	2.0	3.2
Total	100.0%	100.0%	100.0%

These shifts in income percentages point up two conclusions: (1) Dues are today carrying the major load of financing AFS activities; (2) the consistent drop in Advertising percent and in Convention & Exhibit percent, is probably not justified in terms of the extent of AFS expenditures solely on behalf of the entire castings industry, suppliers as well as foundrymen.

Relationship between Income and Expense per member is illustrated in the following table:

	Total Income per Member	Total Expenses per Member	Dues Income per Member
1952-53	$45.43	$50.55	$21.92
1953-54	73.04	58.46	20.00
1954-55	47.90	58.90	24.51
1955-56	75.09	65.87	23.79

Again it will be noted that the increase in Expense per member, between 1953-54 and 1955-56, was considerably greater than the increase in Income per member. In other words, while AFS has been spending more per member in service, the use of its income-producing services has been decreasing instead of increasing. This trend must be reversed.

Since the present Dues schedule went into effect in 1954, the Society's Dues Income has shown pronounced increase, as follows:

	Income 1953-54	Income 1954-55	Income 1955-56
Sustaining	$ 28,005.	$ 48,818.	$ 50,906.
Company	78,466.	97,376.	104,640.
Personal	66,711.	66,333.	75,161.
Affiliate	52,288.	51,208.	60,520.
Others	5,762.	4,987.	5,829.
Total	$231,232.	$268,722.	$297,056.

Even considering an increase of 937 members, the increase is significant, since the only two membership classes where dues increases occurred (Sustaining and Company) showed a loss of 321 between 1954 and 1956.

Investments as of June 30, 1956

The Auditor's Report summarizes Income from Investments held in the Society's Investment Agent Trust Account at the Harris Trust & Savings Bank, Chicago. The report shows the following Income sources:

Interest on U. S. Govt. Bonds	$ 1,227.50
Interest on Canadian Govt. Bonds	1,050.00
Interest on Industrial Bonds	7,660.87
Dividends on Common Stocks	7,017.60
Discount on Treasury Bills	1,441.60
Gain on Securities Sold	11,542.34
Total	$29,939.91

Recap of Investments held by the Society on June 30, 1956, is as follows:

	Book Value (Cost)	Mkt. Value June 30	% Incr. or (Decr.)	% of Total (Mkt.)
U. S. Govt. Bonds	$ 42,000.	$ 41,006.	(2.37) %	8.05%
Canadian Govt. Bonds	20,000.	19,425.	(2.88) %	3.82%
Industrial Bonds	245,000.	240,030.	(2.03) %	47.11%
Common Stocks	118,509.	209,051.	76.40 %	41.02%
Total	$425,509.	$509,512.	19.74 %	100.0 %

Between June 30, 1955 and 1956, the percent held in Industrial Bonds decreased from 53.8% to 47.11%; the percent in Common Stocks increased from 27.0% to 41.02%. This follows the present trend of trust fund investments, to increase common stock percentages as a "hedge" against anticipated developments.

In 1950 the Board of Directors determined, as a policy, to build up a General Reserve Fund "equal to the Society's average expense over the previous four years," subject to biennial adjustment. The result of this policy may be seen in the following statement covering the period 1948 to date:

Cash and Investments, 9 Years (Including all AFS Funds)

	Cash	Investments	Total	Incr. or (Decr.)
June 30, 1948	$106,394.	$178,805.	$285,199.	—
June 30, 1949	85,785.	133,805.	219,590.	$ (65,609.)
June 30, 1950	229,497.	118,805.	348,302.	128,712.
June 30, 1951	103,913.	308,805.	412,718.	64,416.
June 30, 1952	297,608.	330,600.	628,208.	215,490.
June 30, 1953	129,920.	448,761.	578,681.	(49,527.)
June 30, 1954	127,053.	564,573.	691,626.	112,945.
June 30, 1955	31,076.	456,904.	487,980.	(203,644.)
June 30, 1956	115,605.	460,960.	576,565.	88,585.

Financial Condition as of June 30, 1956

The Society's financial condition as of June 30 is compared in the Auditor's Report for the past three years, showing an *increase* of $107,603 or 13.43% in Fund Principals over June 30, 1955 . . . and a *decrease* of $3,830 or 0.42% from June 30, 1954.

2-Year Budget 1955-1957

The Society is now operating on a 2-year budget, in order to counteract the "feast-and-famine" financing in effect since 1938, the last year that AFS held an Exhibit annually. Two-Year budgets are established at the beginning of each *odd-numbered* year, subject to review annually. The following data shows the current 2-year budget for 1955-1957, including *Actual* Income and Expense for 1955-56, *Budgeted* Income and Expense 1956-57:

	Income	Expense	Excess Income or (Expense)
Year 1955-56, Actual	$ 937,834.	$ 822,668.	$115,166.
Year 1956-57, Budgeted	748,500.	821,800.	(73,300.)
Total	$1,686,334.	$1,644,468.	$ 41,866.

Conclusions, Finances 1955-56

During the fiscal year 1955-56, the Society's financial operations developed precisely as budgeted, again with the single exception of the AFS magazine. In consequence, Excess Income was $14,159 below the budget. Advertiser acceptance of MODERN CASTINGS, solely on the basis of proven ability to produce returns in inquiries and sales, is still far short of the acceptance of *all other* Society activities. Until this condition is rectified, Society finances will still vary widely from year to year, even though Expenses are kept under close budgetary control.

Respectfully submitted,

WM. W. MALONEY
General Manager and Treasurer

Condensed Statement of Income and Expense

(All Expenses Distributed to Major Activities) Fiscal Years July, 1954 — June, 1956

INCOME

	Total Income Amount	Per Cent
Membership Dues	$ 495,420.35	34.7
General Publications	20,867.00	1.5
Special Publications	76,212.10	5.4
MODERN CASTINGS	505,455.34	35.5
Conventions & Exhibit	325,361.29	22.9
TOTAL INCOME	$1,423,316.08	100.0

EXPENSE

	Total Expense Amount	Per Cent
Membership Service	$ 91,308.66	6.4
Technical & Research	153,732.45	10.8
General Publications	68,794.75	4.8
Special Publications	89,136.55	6.2
MODERN CASTINGS	539,886.86	37.8
Chapter Operations	182,836.85	12.8
Conventions & Exhibit	163,359.31	11.4
General Administration	97,905.34	6.9
Safety, Hygiene & Air Pollution Control	41,804.71	2.9
TOTAL EXPENSE	$1,428,765.48	100.0
EXCESS EXPENSE, 2 YEARS	$ 5,449.40	

Comparative Condensed Balance Sheets

As of June 1953 to June 1956, inclusive (4 Years)

ASSETS	June 30 1953	June 30 1954	June 30 1955	June 30 1956	Incr. + (Decr.—) Over 1955	Incr. + (Decr.—) 4 Years
Cash	$129,920.33	$127,052.55	$ 31,075.51	$115,604.32	$ 84,528.81	$ (14,316.01)
Investment Securities	448,761.14	564,572.71	456,904.43	460,960.20	4,055.77	12,199.06
Inventories	39,667.57	37,135.44	52,322.67	52,217.54	(105.13)	12,549.97
Accounts Receivable	37,662.98	17,266.10	36,179.42	10,283.93	(25,895.49)	(27,379.05)
Deferred & Prepaid Items	12,950.68	9,925.30	11,153.74	8,103.29	(3,050.45)	(4,847.39)
Furniture & Fixtures (net)	14,616.25	13,045.65	27,151.35	33,984.58	6,833.23	19,368.33
Land and Building	24,341.01	155,260.20	277,338.88	273,101.93	(4,236.95)	248,760.92
TOTAL	$707,919.96	$924,257.95	$892,126.00	$954,255.79	$ 62,129.79	$246,335.83
LIABILITIES						
Current Liabilities	$ 27,946.24	$ 11,803.38	$ 91,104.99	$ 45,630.97	$ (45,474.02)	$ 17,684.73
Deferred Income	—	—	—	—	—	—
Fund Principals	679,973.72	912,454.57	801,021.01	908,624.82	107,603.81	228,651.10
TOTAL	$707,919.96	$924,257.95	$892,126.00	$954,255.79	$ 62,129.79	$246,335.83

NOTE: Includes all Awards, Building, S&H&APC, and Retirement Funds.

REPORT OF THE AFS GENERAL MANAGER

(Fiscal Year Ended June 30, 1956)

This report sums up all major activities of the Society during the past year, including Technical activities, and also includes outlines of programs planned for the year 1956-57. The report is confined to non-financial matter, finances being discussed in a separate report of the Treasurer.

1. Membership

During the past year, the Society's membership increased to an all-time high of 12,488, a net gain of 1,523 or 13.9% above the total of 10,965 on June 30, 1955. The increase was also 488 above the year's "Target" of 12,000 members.

Membership Gains & Losses, 4 Years

	1952-53	1953-54	1954-55	1955-56
Members gained (all Classes)	1,947	1,723	1,533	2,560
Members lost (all Causes)	980	1,205	2,119	1,037
Net Gain or (Loss)	967	518	586	1,523
Avg. Gain or (Loss) per month	81.0	43.2	48.8	126.9
Total on June 30	11,033	11,551	10,965	12,488
Rate of Turnover	8.9%	4.5%	(5.3%)	12.2%
Pct. of New Members Held (Lost)	49.6%	30.0%	(38.2%)	59.2%

It should be noted that 60% of the 1,523 increase (912 members) came from the Chapters active in July 1955 . . . and 40% (611) from the three new Chapters established during the year. (Actually, the gain from the three new Chapters was less than 611, because of transfers from adjoining Chapters.) Of the 42 Chapters active last July, only 5 showed net losses for the entire fiscal year . . . a remarkable tribute to the membership activity of the Chapters and Board Contacts Committee.

Membership activity by AFS Regions:

	No. Chapters	No. Target Chapters	% Target Chapters	Total Increase	Pct. Increase
I	14	5	35.7%	826	32.7%
II	10	6	60.0	366	12.0
III	6	2	33.3	185	9.1
IV	10	3	30.0	158	11.1
V	5	5	100.0	116	14.1
TOTAL	45	21	46.7%	1,651	16.8%

It will be noted that the Members Gained total (2,560) was the largest in four years; the Members Lost (1,037) lowest in three years, due to a larger-than-usual percent of reinstatements. However, the percent of reinstatements of members dropped as delinquent in dues was the lowest in four years, as follows:

	1952-53	1953-54	1954-55	1955-56
Delinquents Dropped	1,533	1,784	2,999	1,924
Delinquents Reinstated	968	958	1,392	1,163
Pct. Reinstated	63%	54%	46%	40%

A total of 21 Chapters made their assigned membership "Targets" in 1955-56, compared with 4 the previous year, 11 in 1953-54, and 21 in 1952-53. The work of the Field Membership Director during the first year of this activity has been well received. Results from the 28 Chapters visited: 11 (39%) made their assigned "Targets;" member increase 678 (45% of total AFS increase). These encouraging results, plus closer Chapter relations, indicate that the field work "experiment" probably paid off.

2. Chapters

Three new Chapters were formed in 1955-56 . . . Connecticut, Pittsburgh, New England . . . making a total of 45 regular Chapters. No new Student Chapters were formed, the total remaining 12.

During the year, 9 Regional Conferences were sponsored by a total of 27 Chapters, all attended by the President, Vice-President, General Manager and/or Technical Director.

The 13th Annual Chapter Officers Conference was held at the LaSalle Hotel, Chicago, June 28-29, with total attendance of 118, including 96 Chapter delegates representing all 45 Chapters. The total included 41 Chairmen, 43 Program Chairmen, 8 Secretaries and/or Treasurers, and 2 Directors. A total of 42 Chapters had at least two delegates present, 8 more than two. President Shipley presided, and Vice-President Dietert and 7 National Directors were present, compared with 12 in 1955, six in 1954, four in 1953. All but one of the 7 new Directors attended.

3. Research

A total of 18 research projects carried on under AFS auspices during the past year are detailed in a separate report on Research. The projects deal with brass and bronze, light metals (2), steel (2), malleable iron, gray iron (2), cupola research, heat transfer, and sand (8). With a total of $48,000 budgeted, $43,082 was expended during 1955-56, with additional funds committed.

New projects already initiated or now commencing include: pressure tightness of 85-5-5-5 leaded red brass castings, at University of Michigan; study of "snotter" defects in steel castings, at Steel City Testing & Engr. Co.; mechanics of metal flow in molds; study of gray iron gating and risering, at Case Inst. of Technology; solidification of metals, at Cornell University; study of loss of heat of molten metal while in the ladle; critical review of world literature on technical progress of cupola operation; study of physical-chemical aspects of sand particles; study of main methods of producing pearlitic malleable.

Periodic research reports are issued to the Board Research Committee. A "Research Advisory Committee" may be formed to periodically evaluate and coordinate all AFS research.

Further development of AFS research now hinges on establishment of a proposed AFS Training and Research Foundation.

4. Technical Committees

The Society's Divisions and General Interest Committees held over 100 meetings during the past year and deserve the unanimous thanks of the Board and membership. These groups prepared Convention programs and reviewed papers presented, prepared technical publications, supervised research, and cooperated closely with the Central Office.

Several new committees were formed in 1955-56, including: Research Advisory Committee, Pearlitic Malleable Committee, Fundamental Papers Committee, and Management Committee. Others were discontinued, as no longer needed or work completed or tabled. Committee work culminated in a 5-day National Committee Week, June 4-8, with a meeting of the Technical Council.

Castings Clinics, planned for individual Chapter needs, were presented in 1955-56 at Southern California, Northern California, Northwest Regional Conference, Corn Belt, University of Alabama, and Texas. A short course on "Principles of Molding and Core Sand Practice" was developed with O. Jay Myers and C. A. Sanders and presented at the Northwest Regional and Corn Belt and Metropolitan Chapters.

A new Chapter Speaker List and a new Film List were developed for the 1956 Chapter Officers Conference. The library of films now available at the Central Office includes 10 separate subjects, in constant use, and more are constantly sought for non-commercial nature.

The National Committee Roster was prepared and published; a portable Book Display developed and displayed at all Regional Conferences possible; and a film on the CO_2 Process was shown frequently throughout the year. The AFS Library has acquired over 200 books since March 1955, and currently receives 138 periodicals which are scanned for items of member interest. The Librarian also maintains a file of clippings and abstracts, prepares technical bibliographies, and will compile an initial bibliography on cupola operation for the Cupola Research Committee.

Other Staff activities include the answering of frequent inquiries of a technical foundry nature, received from many countries and ranging from simple to highly complex questions requiring considerable search of the literature. Liaison also is maintained with a number of professional engineering societies such as ASM, AIME, ASTM, ASEE, and Society for Non-Destructive Testing.

5. Publications

Bound volume of TRANSACTIONS, vol. 63 (1955) was mailed in November. Total printing 2,680 copies, 962 shipped on pre-publication orders, 881 distributed gratis (Sustaining, Company, Honorary Life members) on request, 301 shipped post publication, 536 on hand June 1956. Number of gratis copies requested was nearly 200 less than for the 1954 volume.

Four new Special Publications were produced during the past year, and one reprinted, as follows:

Symposium on Non-Destructive Testing. (New)
Principles of Metal Casting. (New)
Engr. Manual for Control of In-Plant Environment. (New)
1955 Annual Hoyt Lecture, F. J. Walls. (New)
Patternmaker's Manual. (Reprint)

An impressive schedule of Special Publications has been recommended by the Publications Committee, as follows:

Cast metals Handbook.
Symposium on Metallography.
Die Casting Methods.
Air Pollution Control Manual.
Foundry Safety Manual.
Manual of Welding Hazards Control.
1956 Annual Hoyt Lecture.

In addition to the above, at least 7 other major publications are now in work, with completion anticipated during the next two years, as follows:

Patternmakers Manual (complete revision).
Foundry Molding Sand Practice.
Recommended Practices for Light Alloys.
Metallography of Cast Alloys.
Manual on Design and Stress Analysis.
Foundry Noise Control Manual.
Foundryman's Radiation Protection Manual.

As a result of the studies made by the Board's Special Committee on Publications Pricing, Director Morley, chairman, Staff attention has been directed to pricing Special Publications so as to produce a net income instead of net expense. Finance Committee recommendations at this Board meeting are expected to make this possible.

6. Convention and Exhibit

Approximately 14,500 registered at the 60th AFS Castings Congress & Exhibit held in Atlantic City, May 3-9. This included 4,781 Members, 2,694 Non-Members, 3,100 Exhibitor representatives, 291 Foreign, 1,779 "Free Day" guests, and 546 Ladies. Convention days were Thursday through Wednesday, with no Exhibits or sessions on Sunday, Banquet on Friday, Annual Business Meeting on Saturday (as a Luncheon meeting) followed by Annual Lecture, "Free Day" for Exhibits on Saturday.

Advance registration was 5,989, heavier than ever before. The Ladies program was arranged and financed entirely by AFS, with a paid Staff Hostess. Daily Attendance lists were printed and distributed on 6 days. No organized Plant Visitations were arranged.

A total of 236 exhibitors occupied 91,045 sq. ft., compared with 290 occupying 97,586 sq. ft. in 1954. Thus the 1954 Show was the largest held thus far by AFS, while gross revenue of the 1956 Show was largest.

A total of 80 papers were presented in 1956, of which 40 were preprinted, 5 distributed at the Convention, and 9 published in MODERN CASTINGS. Some 3,000 requests were received for Preprints in advance of the meeting.

The Divisions and Committees cooperated closely with the Society in reducing the total number of sessions held so as to allow ample time for visitors to inspect exhibits as well as attend technical sessions. Innovations of the program included the following:

(1) Symposium on Metallography — two evening sessions.
(2) Malleable Base Spheroidal Iron — presented at joint Malleable-Gray Iron luncheon.
(3) Vacuum Melting and Pouring — German film, presented at Steel luncheon.
(4) Principles of Core Blowing — at Sand Division Dinner.
(5) Fundamental Papers — initial session, well attended.
(6) The CO_2 Process — I.B.F. official exchange paper, best attended 1956 session.
(7) Dust Control in Steel Foundry — British film, presented at S&H&AP session.
(8) AFS Research Exhibit — first time attempted.

The Annual Banquet, with Dr. Norman Vincent Peale as guest speaker, was a "sell-out" with 845 present, and well received. A record total of 16 past Presidents of AFS attended the Past Presidents Luncheon, and received a bronze plaque at the Alumni Dinner. Registration was handled very smoothly, the daily Author-Chairman breakfasts were continued and now are well established, and the Annual Lecture on "Marketing Your Product" by S. C. Massari was well received.

Convention Comments. The following comments can be considered as "future intentions" at AFS Conventions and Exhibits, based on experience of the past two years:

(1) Advance Registration — Continue and broaden, but with separate badges for: Members, Member-Exhibitors, Guests, Exhibitors.

(2) Ladies Program — Continue as arranged by AFS with paid Staff Hostess and "reasonable" registration fees.

(3) Annual Business Meeting and Lecture — Hold as a morning session, not a luncheon meeting.

(4) Daily Attendance Lists — Continue, at Exhibit meetings.

(5) Convention & Exhibit Days — Hold future Exhibit Conventions Monday thru Friday, with "Free Day" on Saturday preceding opening day.

(6) Annual Banquet — Continue policy of outstanding guest speakers, without entertainment program.

(7) Author-Chairman Breakfasts — Continue.

(8) Technical Sessions — Endeavor to reduce total number of sessions, and number of luncheon and dinner sessions in Exhibit years only.

(9) Membership Work — Confine to an attractive AFS booth, either separate or a part of AFS Publications booth.

(10) AFS Research Exhibit — Do not attempt unless funds available for an outstanding display.

7. Education

A major new undertaking in 1955-56 was the first AFS Foundry Instructors Seminar, held June 14-16 at Michigan State University for instructors in foundry and patternmaking at high schools and vocational schools. Vice-President Dietert and Prof. W. H. Ruten acted as Co-Chairmen. Attendance totaled 101, including 91 Instructors from 11 states, and 10 Guest Speakers and Guests.

Consensus of all present was decidedly enthusiastic and in favor of an annual Seminar. AFS received a great deal of information helpful to its future work at the secondary school level. The program was approved and sponsored by the Education Division, with Directors Oster and Cech lending valuable assistance.

The 1956 Apprentice Contest was again held in five divisions, and total entries of 431 set a new high, including 15 Chapters and 153 companies. First prize winners in all five divisions received their awards in person at Atlantic City.

The Joint Committee on Foundry Vocational Training, established in 1954 to write a series of in-plant training manuals, is at present inactive although included in "Area No. 3" of the Education Division's program. Recent resignation of the chairman makes revitalization of this program imperative.

8. Safety, Hygiene & Air Pollution Control

The work of the S&H&AP group still is largely concentrated on preparation of six basic engineering manuals on:

Control of In-Plant Environment (interior).
Control of Air Pollution (exterior).
Safe Practices for Welding, Cutting, Brazing, Soldering in Foundries.
Foundry Safe Practices.
Foundry Noise Control.
Radiation Protection for the Foundryman.

All of these essential manuals are of a "pioneering" nature, with no existing publications covering the same field for foundry use. When completed, the six will form a major task whose value to foundry management is incalculable but great.

During the past year, the program Director assisted a number of foundry groups in formulating local civic codes acceptable to foundry operators, at: Toronto, Philadelphia, New York, Beloit, and Portland, Ore. He has also induced revision of the ASME "model" Smoke Abatement Code for the protection of foundry operations.

9. "MODERN CASTINGS"

While the AFS magazine staged a substantial recovery in Advertising during 1955-56, total pages for the past 12 months were still below the previous 12 months. In readership and appearance, MODERN CASTINGS showed marked improvement editorially. Ample evidence exists that advertising inquiries have increased remarkably, and the "bonus" sections are widely read, clipped and retained for permanent use. A new magazine file holder will bring even greater retention of the magazine.

However, magazine expense has risen sharply . . . out of proportion to increased acceptance by advertisers and readers. As a result, the net for 1955-56 indicates that closer control must be exercised.

Replacement of the magazine Editor, resigned July 1, now has been accomplished and completes the present "team" for the job ahead.

10. International Affairs

Due to unforseen events, it became impossible for the Central Office to be represented at the 1956 International Foundry Congress in Germany. Official AFS representatives in attendance were: Vice-President Dietert, who presented the official AFS exchange paper; past President W. L. Seelbach; and Vincent Delport, AFS European representative. Mr. Dietert presented an AFS invitation for the International Committee to hold the 1962 "International" in U.S.A.

The 1956 Congress was accomplished by a large Foundry Fair of both foundries and suppliers, and a number of American foundry equipment firms exhibited.

As directed, AFS contributed $100 to the I.B.F. Tom Makemson Scholarship Fund, in England, and received grateful acknowledgment.

11. General Administration

It is suggested that the National Directors, in telegraphing the Central Office in Des Plaines, utilize the AFS teletype equipment by sending telegrams via WXU Western Union Service, to save time.

Respectfully submitted,
WM. W. MALONEY
General Manager

July 23, 1956

5-Year Comparative Membership Report

As of June 30, 1952-56, Inclusive

	June 30, 1952	June 30, 1953	June 30, 1954	June 30, 1955	June 30, 1956
Sustaining Members	191	193	191	150	145
Company Members	1,259	1,278	1,280	1,014	1,005
Personal Members	7,574	8,471	8,916	8,718	10,087
Junior Members	501	567	639	564	719
Honorary Life Members	79	84	87	88	97
International Members	429	440	438	431	435
TOTAL	10,033	11,033	11,551	10,965	12,488
New Members, 12 Months	1,850	1,947	1,723	1,533	2,560
Resignations	193	265	324	273	244
Delinquent Dropped	813	565	826	1,607	761
Removed by Death	17	32	29	24	30
Net Gain (Loss) for Year	827	967	518	(586)	1,523
Members in Chapters	9,320	10,272	11,504	10,917	11,882

AFS Membership and Distribution

(As of June 30, 1956)

Chapter	Sustaining	Company	Personal	Affiliate	Associate	Junior	Honorary	Total
Birmingham	8	22	108	320	9	2	5	474
British Columbia	0	5	56	15	5	3	0	84
Canton District	0	19	44	74	0	1	0	138
Central Illinois	2	6	37	179	3	19	1	247
Central Indiana	2	32	64	204	5	10	2	319
Central Michigan	2	12	39	147	7	11	4	222
Central New York	1	19	79	94	0	1	0	194
Central Ohio	2	17	57	95	3	2	3	179
Chesapeake	3	28	89	86	34	1	10	251
Chicago	21	60	209	456	8	76	7	837
Cincinnati District	8	39	100	135	5	9	3	299
Connecticut	4	15	53	37	3	2	0	114
Corn Belt	0	3	9	21	3	2	0	38
Detroit	9	38	140	340	12	6	12	557
Eastern Canada	5	27	77	185	18	8	3	323
Eastern New York	1	9	28	67	7	1	0	113
Metropolitan	9	45	144	202	20	5	8	439
Mexico City	0	4	39	21	0	5	0	69
Michiana	4	17	69	184	3	2	0	279
Mid-South	0	2	24	32	0	0	0	58
Mo-Kan	1	4	40	17	2	0	0	64
New England	0	13	111	37	2	0	2	165
Northeastern Ohio	13	67	167	480	10	31	9	777
Northern California	0	19	116	115	8	1	0	259
Northern Illinois & Southern Wisconsin	1	14	25	75	1	3	1	120
Northwestern Pennsylvania	0	17	53	103	1	2	2	178
Ontario	2	56	133	200	2	3	0	396
Oregon	1	10	41	82	4	3	0	141
Philadelphia	8	44	139	163	12	4	5	375
Pittsburgh	5	17	223	71	6	3	1	332
Quad City	3	20	50	135	4	6	1	219
Rochester	1	2	40	42	0	2	0	87
Saginaw Valley	6	14	34	252	2	55	3	366
St. Louis District	2	18	133	92	6	2	0	253
Southern California	2	40	178	139	10	8	0	377
Tennessee	0	11	54	152	1	3	0	221
Texas	2	27	75	102	4	1	1	212
Timberline	1	2	37	21	3	4	0	68
Toledo	1	16	35	123	2	3	1	181
Tri-State	0	8	52	49	1	5	1	116
Twin City	1	22	75	96	13	6	1	214
Washington	0	2	41	17	16	4	1	81
Western Michigan	1	24	48	161	3	2	1	240
Western New York	1	21	77	135	2	1	1	238
Wisconsin	9	70	125	359	5	4	5	577
TOTAL IN REGULAR CHAPTERS	142	977	3567	6124	265	322	94	11491
STUDENT CHAPTERS								
Massachusetts Institute of Technology	—	—	—	—	2	19	—	21
Michigan State University	—	—	—	—	—	26	—	26
Missouri School of Mines	—	—	—	—	1	59	—	60
Ohio State University	—	—	—	—	—	16	—	16
Oregon State College	—	—	—	—	—	15	—	15
Pennsylvania State University	—	—	—	—	—	32	—	32
Polytechnic Institute of Brooklyn	—	—	—	—	—	24	—	24
Texas A. & M. College	—	—	—	—	—	23	—	23
University of Alabama	—	—	—	—	—	48	—	48
University of Illinois	—	—	—	—	—	71	—	71
University of Michigan	—	—	—	—	—	21	—	21
University of Wisconsin	—	—	—	—	—	34	—	34
TOTAL IN STUDENT CHAPTERS					3	388		391
TOTAL IN ALL CHAPTERS	142	977	3567	6124	268	710	94	11882
Foreign	0	17	404	8	23	9	2	463
Non-Chapter	3	11	27	23	18	0	1	83
Members in Military Service	—	—	—	—	—	—	—	60
GRAND TOTAL	145	1005	3998	6155	309	719	97	12,488

Minutes

Special Meeting 1955-56 Executive Committee

Drake Hotel, Chicago — August 8, 1955

(1) **ROLL CALL:**

President Bruce L. Simpson, *presiding*
Vice-President Frank W. Shipley
Past-President Frank J. Dost
Regional Vice-Presidents:
Region 1, Thomas Kaveny, Jr.
Region 3, C. V. Nass
Region 4, L. H. Durdin
} Executive Committee
Region 2, M. J. Lefler
Region 5, W. R. Pindell

Wm. W. Maloney, *General Manager*

(2) A quorum having been established, President Simpson called the special meeting to order.

(3) The by-laws of the Society were read describing the method by which the Nominating Committee is appointed to fulfill its functions for election of a Vice-President and new Directors of the Society at the 1956 AFS Convention. Names of candidates were presented from lists submitted by 26 eligible Chapters (4 eligible Chapters not submitting; 12 Chapters ineligible).

(4) In accordance with the by-laws, 5 members were appointed by the Executive Committee to form, together with Past Presidents Frank J. Dost and Collins L. Carter, the 1955-56 Nominating Committee as follows:

Past President Frank J. Dost (1954-55), *Chairman.*
Past President Collins L. Carter (1953-54).
W. T. Shute, Works Mgr., Steel Foundry Div., Canadian Car & Foundry Co., Montreal, Que. (Rep. Region No. 1, Chapter Group D — Eastern Canada Chapter and Steel).
C. D. Schopp, Asst. General Supt., Link-Belt Co., Indianapolis, Ind. (Rep. Region No. 2, Chapter Group J — Central Indiana Chapter and Malleable).
W. D. McMillan, Supervisor of Met., International Harvester Co., Chicago (Rep. Region No. 3, Chapter Group K — Chicago Chapter and Gray Iron).
Earl White, Sales Repr., M. A. Bell Co., St. Louis (Rep. Region No. 4, Chapter Group N — Corn Belt Chapter and Supplies).
Howard H. Havies, Chief Estimator, Vivian Diesels & Munitions, Ltd., Vancouver, B. C. (Rep. Region No. 5, Chapter Group R — British Columbia Chapter and Gray Iron).

(5) The General Manager was instructed to notify those selected, for prompt acceptance.

(6) There being no further business to be considered, the Executive Committee adjourned *sine die.*

Approved:
BRUCE L. SIMPSON, *President*
October 6, 1955

Minutes

Annual Meeting 1955-56 Board of Directors

Drake Hotel, Chicago — August 8 - 9, 1955

(1) **ROLL CALL:**

President Bruce L. Simpson, *presiding*
Vice-President Frank W. Shipley

Directors:

(Terms expire 1956)	*(Terms expire 1957)*
Charles E. Brust	Frank C. Cech
Frank J. Dost	Lewis H. Durdin
Edw. C. Hoenicke	B. George Emmett
Thos. Kaveny, Jr.	Mm. W. Hamilton
Chester V. Nass	Wm. A. Morley
Wm. R. Pindell	Harold L. Ullrich

(Terms expire 1958)
Curtis C. Drake
Harold C. Erskine
Charles W. Gilchrist
O. Jay Myers
Charles E. Nelson
Richard A. Oster
Robert W. Trimble

General Manager Wm. W. Maloney
Assistant Secretary A. B. Sinnett

Absent: Directors Martin J. Lefler (exp. 1956), G. Ewing Tait (exp. 1956), and Eugene R. Oeschger (exp. 1957).

(2) A quorum having been established, President Simpson called the Annual Meeting of the Board of Directors to order.

(3) Reading of Minutes

Minutes of the meeting of the Board of Directors held in Houston, May 28, 1955, were read and on motion duly made, seconded and carried, approved with one correction: Chairman of the 1955-56 Board of Awards to be F. J. Walls instead of S. V. Wood.

Minutes of the Executive Committee meeting held June 15, 1955, in Chicago, were read and accepted on motion duly made, seconded and carried, with one correction: Deferment of final action on the Pangborn scholarship offer is permitted to December 31, 1956 (not 1955), if additional time needed.

(4) Report of General Manager

Membership: Total membership 10,965 on June 30, a net loss of 586 or 5.1% from total of 11,551 on June 30, 1954; lowest total in four years and first time in years that members lost exceeded members gained. The General Manager stated that a target of 12,000 members had been established for the year 1955-56.

Chapters: No new regular or Student Chapters established during the fiscal year. Chapter total remaining at 42 regular Chapters, 14 Student Chapters (2 inactive). The report included data on the 1955 Chapter Officers Conference held June 16-17: attendance 104, 77 Chapter delegates.

1955 Convention: Attendance at the 59th Annual Convention in Houston, May 23-27, between 1900 and 2000, including ladies and guests; total *registered* attendance 1500. Commendations were given the Texas Chapter for the very efficient housing bureau, operated without previous experience.

Research projects on Brass & Bronze, Gray Iron, Light Metals, Malleable, Sand, and Steel were carried on during the year. Steel research on hot tears facilitated at Armour Research Foundation with balance of funds donated by American Steel Foundries; Light Metals research at Battelle Memorial Institute, by additional funds from Frankford Arsenal.

A separate and more detailed report on AFS Research by the Technical Director was presented, outlining progress in each research project and work planned for the coming year.

Publications issued during the fiscal year 1954-55:
Time and Motion Study for the Foundry.
Processing Molding Sands.
Control of Emissions from Metal Melting Operations.
Melt Quality and Pressure Tightness of Copper-Base Alloys.
Engineering Manual for Control of In-Plant Environment (Secs. 1, 2, 3).

The long-awaited College Foundry Textbook, available for distribution in September. All priority to be given the following publications in 1955-56:

S&H&AP Environmental Control Manual (completion).
Symposium on Non-Destructive Testing.
S&H&AP Safety Manual.
Cast Metals Handbook.

Technical Activities: The report praised the work of AFS Technical committees during the past year and emphasized several developments: holding of Casting Clinics and Sand Schools; growth and further development of AFS Library as a technical tool; availability of two new films through AFS, one on Vertical Gating Systems by the Light Metals research project, and a re-editing of a German film on the CO_2 Process; development of a number of valuable bibliographies; increased number of technical inquiries from members; committee meetings, Chapter talks and other outside visits by the Technical Staff.

Educational Activities: Greater contacts with secondary schools; record Apprentice Contest, with 365 entries from 138 companies and 14 Chapters; complete revision of "The Foundry Is A GOOD Place to Work."

S&H&AP Program: Work of the S&H&AP group still largely concentrated on development of essential manuals on Dust Control & Ventilation, and Air Pollution Control — the former in final stages, the latter still many months away; S&H&AP activities now being completely financed from general operating funds; $1100 in uncompleted pledges written off as of June 30, 1955, or only 1.5% of the contributed total.

Headquarters Building: Occupied October 1, 1954, formal dedication November 18, some projects still remaining for completion. F. C. Riecks continues as the owner's sole representative in charge of engineering and construction until general contractor completely discharged. As of June 30, total of $540 in outstanding pledges written off as uncollectible, only 0.2% of the contributed total.

MODERN CASTINGS. The General Manager recalled steps leading to present reorganization of AMERICAN FOUNDRYMAN: change in name, change in size and format, changes in sales representation, and appointment of C. G. Fuller as Managing Director in charge of editorial, advertising and production. Mr. Fuller entered the meeting on invitation and explained why the new size and format were sound from a publishing standpoint, and stated that steps were being taken to increase reader acceptance.

The General Manager pointed out that all changes in the magazine had been with two main points in mind: (1) to develop a magazine completely different from any other publication serving the foundry field, and (2) to progress the Board requirement for greater net income from the Society's official publication.

International Congresses: Two international congresses held during 1954-55: Florence, Italy in September, and London, England in June, with only six Americans in total attendance. It was recommended for Board consideration that one of the technical men of the Central Office be assigned to participate in the 1956 International Foundry Congress and Exposition at Duesseldorf, Germany.

Administrative Reassignments: A. B. Sinnett, Assistant Secretary; D. J. Hayes, membership field work; W. N. Davis, Exhibits Manager; H. J. Weber, Director of S&H&AP Control; C. G. Fuller, Managing Director MODERN CASTINGS, and H. F. Scobie, Editor.

Report of the General Manager was accepted on motion duly made, seconded and carried.

(5) Report of the Treasurer

For the fiscal year ended June 30, 1955; Income $525,236 or 12.1% under budget of $597,675; Expense $645,852 or 2.3% below budget of $661,120; Excess Expense $120,615 or 90.1% over budget of $63,445. Individual items of Income and Expense were examined in detail as audited by the firm of George Rountree & Co., Chicago.

AFS Income sources for the two years 1953-1955, as audited:

Membership Dues	37.2% of total income
MODERN CASTINGS and American Foundryman	30.4%
Conventions & Exhibit	24.4%
Publications	6.0%
Miscellaneous	2.0%
Total	100.0%

The Society's financial condition as of June 30, 1955, was reported as showing a decrease of $111,433 or 12.2% in Fund Principals from June 30, 1954 . . . and an increase of $121,048 or 17.8% over the two-year period since June 30, 1953.

The General Manager stated that, from the experience of the past fiscal year, the rate of activities of the Society have now reached the state where the following three alternatives should be considered:

(a) The Society must find additional sources of substantial income to support the present and contemplated future rate of activities; or

(b) the Society must reduce its present rate of activities and expenses in order to match its present factual income; or

(c) the Society must reconsider a present Board policy and revert to industry contributions for the support of approved special activities.

Reports of the Auditors and the Treasurer were accepted on motions duly made, seconded and carried, being made a part of these minutes.

(6) Annual Reports of Board Committees 1954-55

Chapter Contacts Committee. President Simpson reported as Chairman of the Committee during 1954-55, stating that more concentrated effort is needed to increase membership. He thanked the Directors for their fine work during the past year, and promised to keep them posted monthly on membership in their respective areas.

Exhibits Committee. The General Manager stated that the Exhibitors Committee had agreed upon exhibit hours and rules and regulations for the 1956 AFS Exhibit. Recommendations of the committee: (1) That a survey of attendance be made at the 1956 show by AFS to determine traffic flow during exhibit hours; (2) that AFS prepare a daily attendance list in 1956; (3) that a questionnaire on convention and exhibit details be developed and mailed.

National Castings Council. President Simpson and past President Dost reported on the Council meeting held July 15, 1955. They announced that a special task committee, comprising the staff heads of trade association members, was developing a brochure publicizing the basic foundry industry, for the information of governmental agencies and distribution by NCC members, cost to be shared by Council members.

President Simpson stated that a CIO-UAW pamphlet on the silicosis "hazard," harmful to the foundry industry and called to the attention of AFS, had been referred to the trade association members of NCC as beyond the scope of AFS for action.

Publications Committee recomendations: (1) a special book promotion campaign in 1955-56 to reduce existing inventories; (2) a survey of office reproduction equipment to determine cost savings from new and larger press equipment; (3) publication of the Cast Metals Handbook by AFS rather than an outside publisher.

Results of a preliminary survey on office reproduction equipment were presented by the General Manager with recommendation for purchase of a new 14x20 offset press at a cost of approx. $7,000. Following discussion, the Board deferred action, requesting more explicit information on estimated cost savings and amortization period.

The President appointed a special Board committee to consult with the existing Publications Committee for possible revision of present pricing formulae, with a goal of recovering full costs of publications. The committee: Director Morley, chairman; Directors Ullrich and Oster.

Board Research Committee. Director Hoenicke, chairman, reported for the Committee: the desirability of personal contacts by committee members with the various divisional research committees, for more intelligent estimations of projects undertaken and proposed; that some definite percent of AFS income be allocated annually for research.

(7) Special Reports on 1954-55 Activities

Pangborn Grant. The President reported that the Pangborn Foundation had agreed to hold open until December 31, 1956, if necessary, the offer of a $50,000 donation for scholarship purposes, so as to allow AFS time for reclassification of AFS tax-exempt status. He stated that the accounting firm of Lybrand, Ross Bros. & Montgomery, of Chicago, had been requested to seek such tax reclassification, as authorized by the Executive Committee. On motion made, seconded and carried, the Board approved retaining the above-mentioned firm for the purpose stated.

(8) Election of AFS Chief Administrative Officer

In executive session, the incumbent General Manager was reelected for the fiscal year 1955-56.

(9) Staff Salaries and Compensations

In executive session, recommendations of the Financial Committee for Staff salaries and compensations during the fiscal year 1955-56 were considered and approved.

(10) Budget of Income and Expense

Recommendations of the Finance Committee for budgets of estimated Income and Expense for the 2-year period 1955-1957 were considered in detail and approved, as follows:

1955-56: Est. Income $894,600; est. Expense $765,275; est. Excess Income $129,325.

1956-57: Est. Income $637,500; est. Expense $711,620; est. Excess Expense $74,120.

1955-1957 (2 Years): Est. Total Income $1,532,100; est. total Expense $1,476,895; est. Excess Income (2 years) $55,205.

It was pointed out that recommendations for 1956 Exhibit income budget included: (a) elimination of $25.00 space permit fee; (b) increasing space rental rate from $3.00 to $3.30 per square foot. Following discussion, all budget recommendations were approved on motion duly made, seconded and carried, with one negative vote.

(11) Recommendations of Finance Committee

Chapter Dues Refunds. It was pointed out that AFS had agreed when the new dues structure was made effective July 1954, to reimburse for dues refund losses in 1954-55 all Chapters maintaining their memberships during the year. The General Manager stated that the British Columbia, Metropolitan, Saginaw Valley and Northeastern Ohio Chapters had maintained membership but suffered refund losses. On motion made, seconded and carried, the General Manager was authorized to reimburse these four Chapters for refund losses.

Sales of Investment Securities. President Simpson reviewed the sale of $30,000 industrial stocks held in the AFS Investment Agency Trust account at Harris Trust & Savings Bank, Chicago, sale having been consummated to provide operating cash until such time as dues and 1956 exhibit revenues become available. On motion duly made, seconded and carried, action approved.

(12) Chapter and Membership Activities

At the request of Vice-President Shipley, Chairman of the 1955-56 Chapter Contacts Committee, all Directors were presented with data on Chapter contacts during the coming year. He urged all Directors to make their contacts as early as possible, with strong emphasis on the importance of increasing membership.

(13) New Business

Approval of Resolutions. The following resolutions required by banks of deposit for the withdrawal of Society funds and documents were approved on blanket motion duly made, seconded and carried:

RESOLVED that resolutions required by the Harris Trust and Savings Bank of Chicago and by the First National Bank of Des Plaines, authorizing the withdrawal of Society funds, are hereby approved and the Secretary authorized to certify thereto.

RESOLVED that checks for the withdrawal of funds deposited in the name of the Society with depository banks, including all Checking accounts and Interest Savings accounts, and for the disposition of all securities held in the various funds of the Society by the Harris Trust & Savings Bank of Chicago and/or the First National Bank of Des Plaines, shall require the signature of any two of the following: President, Vice-President, Secretary-Treasurer, Assistant Secretary, Chief Bookkeeper.

RESOLVED that the Secretary be authorized to maintain, if deemed feasible, a safety deposit box in the name of the Society at some convenient location for the safekeeping of Society documents, and that any two of the following have authority to obtain access to such safety deposit box: President, Vice-President, Secretary-Treasurer, Assistant Secretary, Chief Bookkeeper.

RESOLVED that the Secretary be authorized to execute all contracts for the administration of Society affairs, subject to specific approval by the Board of Directors. In the case of AFS-sponsored research projects, approval of the project by the Board of Directors includes authority for the Secretary to execute contracts for performance of such projects on a bid basis.

RESOLVED that the Secretary be authorized to reimburse traveling expenses of members of the Society in attendance at any regularly-called Board of Directors, Executive Committee, or Technical Committee meeting, with the following exceptions: No expenses shall be paid to Directors or Committee members for attendance at meetings held during the week of the Annual Convention of the Society, unless specifically authorized by the Board of Directors. When meetings are held in conjunction with other committees or associations, the Secretary is authorized to determine what portion of the expense of such attendance shall be paid by the Society.

RESOLVED that the Secretary be authorized to negotiate the compensations of necessary Staff employees below $6,000 per year, and that Finance Committee approval be required on all compensations of $6,000 per year or more.

RESOLVED that Indemnity Bonds be maintained covering responsible financial officers of the Society as follows: (1) Blanket Position Indemnity Bond covering all staff members in the amount of $5,000; (2) additional Indemnity Bonds in the amount of $15,000 each covering the Bookkeeping Staff; (3) additional Indemnity Bonds in the amount of $30,000 each covering the Secretary-Treasurer, Assistant Secretary and Chief Bookkeeper; (4) Indemnity Bonds in the amount of $25,000 each covering the President and Vice-President of the Society. All premiums for such Indemnity Bonds to be paid by the Society.

AFS Advisory Service on S&H&AP Problems. President Simpson reviewed findings of the Executive Committee on a proposal to offer a plant advisory service on S&H&AP problems: that AFS should not enter into any form of consulting service but should continue to accept the same type of advisory service requests as in the past, on the same non-charge basis as in the past, from communities and foundry groups. On motion duly made, seconded and carried, recommendations of the Executive Committee approved.

Proposed Fundamental Papers Committee. President Simpson reviewed a proposal by Prof. H. F. Taylor of Massachusetts Institute of Technology that AFS establish a special member group to obtain and screen highly technical and theoretical papers on fundamentals of cast metallurgy, and that AFS make provision for presentation and publication of such papers. He stated that the proposal, for a special group to receive papers rejected by regular convention Program and Papers Committees and to select those worthwhile, had been approved in principle by the Technical Council and Executive Committee. Organization of a new Fundamental Papers Committee was approved on motion made, seconded and carried.

Proposed Foundry Research Foundation. The President outlined a broad plan for expanding and consolidating foundry research as now undertaken separately by AFS and various foundry associations. He reviewed AFS research efforts in recent years and proposed the establishment of a Foundry Research Foundation, separate from but under supervision of AFS, and directed by trustees selected from AFS and members of the National Castings Council. He stated that the proposal had been discussed with and informally accepted in principle by both the Executive Committee and Board Research Committee.

Board discussion, while favorable to the plan, emphasized the inherent difficulties involved and indicated the need for further intensive study. Accordingly, the President appointed the following special committee to determine willingness of the trade associations to participate: Past President Dost, chairman; Director Hoenicke, representing gray iron; Directors Emmett and Oeschger, rep. steel; Director Brust, rep. malleable iron; Director Durdin, rep. brass and bronze; Director Erskine, rep. light metals. The committee was requested to report back to the Board in November.

Proposed "Foundryman of the Year" Award. President Simpson offered for consideration a plan to establish an AFS "Foundryman of the Year" award for notable achievements by foundry management, in the interest of attracting more attention of top management to AFS and cementing closer relations with the foundry trade associations. Following considerable discussion, the proposal was tabled.

Appointment of Nominating Committee. The General Manager announced that a 1955-56 Nominating Committee has been appointed by the Executive Committee in accordance with the AFS by-laws, as follows:

Past President Frank J. Dost, President, Sterling Foundry Co., Wellington, Ohio — Chairman.

Past President Collins L. Carter, President and General Manager, Albion Malleable Iron Co., Albion, Mich.

W. T. Shute, Works Mgr., Steel Foundry Div., Canadian Car & Foundry Co., Montreal, Que. (rep. Region No. 1, Chapter Group D — Eastern Canada Chapter and Steel).

C. D. Schopp, Asst. General Supt., Link-Belt Co., Indianapolis, Ind. (rep. Region No. 2, Chapter Group J — Central Indiana Chapter and Malleable).

W. D. McMillan, Supervisor of Met., International Harvester Co., Chicago (rep. Region No. 3, Chapter Group K — Chicago Chapter and Gray Iron).

Earl White, Sales Repr., M. A. Bell Co., St. Louis (rep. Region No. 4, Chapter Group N — Corn Belt Chapter and Supplies).

Howard H. Havies, Chief Estimator, Vivian Diesels & Munitions, Ltd., Vancouver, B. C. (rep. Region No. 5, Chapter Group R — British Columbia Chapter and Gray Iron).

Death of R. E. Kennedy. President Simpson reported with deep regret the death of Robert E. Kennedy, AFS Secretary Emeritus, on August 8 at his home in Westville, Ind. He recalled the great services rendered by Bob Kennedy to AFS and the industry, and urged those who could do so to remain in Chicago for the funeral. On motion made, seconded and unamiously carried, the General Manager was instructed (1) to prepare appropriate resolutions to be spread upon the minutes of the Board of Directors in recognition of Bob Kennedy's services, a copy to be forwarded to his daughter, and (2) to assume for AFS all expenses attendant upon the funeral of Mr. Kennedy.

(14) There being no further business to be considered, the Board of Directors adjourned until November 10.

Approved:

BRUCE L. SIMPSON, *President*
October 6, 1955

Minutes

Fall Meeting 1955-56 AFS Board of Directors

Drake Hotel, Chicago — November 10 - 11, 1955

(1) ROLL CALL:

President Bruce L. Simpson, *presiding*
Vice-President Frank W. Shipley

Directors:

(Terms expire 1956)
Charles E. Brust
Frank J. Dost
Edw. C. Hoenicke
Thos. Kaveny, Jr.
Martin J. Lefler
Chester V. Nass
Wm. R. Pindell
G. Ewing Tait

(Terms expire 1957)
Frank C. Cech
Lewis H. Durdin
B. George Emmett
Wm. M. Hamilton
Wm. A. Morley
Harold L. Ullrich

(Terms expire 1958)
Curtis C. Drake
Harold C. Erskine
Charles W. Gilchrist
O. Jay Myers
Charles E. Nelson
Richard A. Oster
Robert W. Trimble

General Manager Wm. W. Maloney
Assistant Secretary Ashley B. Sinnett

Absent: Director Eugene R. Oeschger (exp. 1957)

(2) A quorum having been established, President Simpson called the meeting of the Board of Directors to order.

(3) Reading of Minutes

Minutes of the meeting of the Board of Directors held August 8-9, 1955, were read and accepted, on motion duly made and seconded.

(4) Report of the General Manager

Membership: increased from 10,965 on June 30 to 11,328 on October 31 . . . an increase of 363 or 3.3%, as a direct result of three efforts: field membership work by D. J. Hayes, personal contacts by AFS Directors with the Chapters, and more stimulated efforts by the Chapters.

Chapters. After a lapse of several years renewed activity in Chapter expansion: A Connecticut Chapter, No. 43 and the first in New England, organized and a petition for Chapter status submitted; a new Southern section of the Chesapeake Chapter, to cover Virginia and North Carolina, formed and operating.

Research. A brief report was presented on all phases of AFS-sponsored research projects, backed up by a later report from the Board Research Committee: progress reports being prepared for the steel project on hot tears and heat transfer project at Columbia University; new research project on brass and bronze assigned to Professor Richard Flinn at the University of Michigan, dealing with a study of the effect of feeding distance and riser size on pressure tightness of 85-5-5-5 bronze alloys; Gray Iron Division preparing for new research project, probably to be assigned to Case Institute of Technology, Cleveland, for development of a graphic treatment of solidification phenomena; sand research on effect of elevated temperatures on properties of steel foundry molding sands now under way at Locomotive Finished Materials Co. (Committee 8-L); sand project of Committee 8-J on effect of elevated properties on gray iron molding sands being conducted jointly between the Universities of Michigan and Wisconsin; sand project of Committee 8-H on the effect of mold washes on surface finish going on at Michigan State University.

Publications: the 1955 TRANSACTIONS available in November; "Principles of Metal Castings" available for distribution; new edition of Cast Metals Handbook being edited for printing; S&H&AP Manuals on Air Pollution, Safety, and Welding Hazards available early next summer.

Technical Activities. The General Manager reported that several divisions have added new committees: Nodular Iron Committee, Gray Iron Division; Pearlitic Malleable Committee, Malleable Division; Research Committee, Pattern Division.

On motion made, seconded and carried, the report of the General Manager was accepted.

(5) Report of the Treasurer

Financial statements as of October 31, 1955, were presented in the report of the Treasurer, covering: Income and Expense, Actual vs. Forecasts; Balance Sheet; Investment Agency Trust Account; Official Travel and Expense; also operating statements for the first quarter for AFS and for the Society's magazine. On motion made, seconded and carried, the report of the Treasurer was accepted.

(6) "Modern Castings and American Foundryman"

The General Manager analyzed comments of advertisers and agencies on the new size and format of the magazine, as requested

at the August Board meeting: 65 of 78 comments received approved the new size and changes made; some objections of a traditional nature opposed to any change; an 87.5% increase in reader inquiries during the first 30 days of the new size; a 50% increase in inquiries per item published; and 24 new advertisers since July 1. The General Manager concluded that the magazine now is doing a better job in meeting requirements of the Board, but that the full effect would not become evident prior to evidence of contract renewals.

On motion made, seconded and carried, it was agreed that the magazine name should be simplified to MODERN CASTINGS, effective January 1956.

(7) Reports of Board Committees

Chapter Contacts Committee. Vice-President Shipley reported on Chapter contacts by Directors, stating that the combination of visits by Board members and the Field Director has materially aided prospects of increased membership.

Board Liaison to Education Division. Directors Cech and Oster reported on activities of the Education Division. President Simpson stated that a meeting had been requested by the division chairman with the General Manager, President and Educational Director to discuss future activities of the latter in view of new responsibilities, it being feared that the work of the Division is not now being developed to the point originally intended.

In discussion, it was expressed that the educational work of the Society should be developed along more specific lines; that one field of endeavor be chosen for complete development rather than a number of programs receiving only partial assistance; that the foundry industry must develop more semi-skilled and skilled personnel; and that the Society develop a full-term vocational education program in cooperation with vocational schools throughout the country.

President Simpson appointed a special Board committee to meet with the Education Division Chairman to study the problem of specific fields to be pursued by the Educational Director and Division: Vice-President Shipley, Chairman; Directors Oster, Lefler and Cech. This committee was asked to present recommendations at the February Board meeting.

(8) Pangborn Award Agreement

The General Manager reported that the Pangborn Gold Medal Award agreement had been signed and returned to the Pangborn Corporation, and thus becomes available for award for the first time in 1956, the agreement containing no stipulation as to frequency of award.

(9) Petition for Connecticut Chapter

The General Manager presented a petition for formation of a 43rd Chapter, the Connecticut Chapter, first in New England. He stated that requirements of the by-laws for Chapter information had been fully met, and recommended approval by the Board of Directors. On motion duly made, seconded and carried, the petition was unanimously approved.

(10) Proposed Pittsburgh Chapter

Director Kaveny reported that the Pittsburgh Foundrymen's Association was currently balloting its eligible membership on the question of affiliation with AFS as a Chapter, and that the Board would be kept advised of developments.

(11) Proposed New England Chapter

Vice-President Shipley, following attendance at the regional conference of the New England Foundrymen's Association in October, reported that the New England group plans to ballot its eligible membership in December on the question of affiliation with AFS as a Chapter, results of such ballot to be advised to the Board of Directors.

(12) Proposed "Foundry Research Foundation"

Past President Dost, Chairman of a special Board committee appointed to investigate possibilities of a "Foundry Research Foundation," under AFS sponsorship but with the other foundry associations involved, requested more time for study, so as to present a comprehensive report at the next Board meeting.

(13) Report on Publications Pricing

Director Morley, chairman of a special Board committee appointed to investigate present policies on pricing publications for sale, reported on a study made with the Central Office Staff and Publications Committee. In the opinion of the committee, (a) promotional costs should be included in the pricing of publications, as well as proper percentages of salaries and burden items; (b) the present pricing for Class II — General publications, which calls for a member price of "out-of-pocket expense plus 125%," may not be sufficient today even if adequate when established as a policy in 1952.

Following discussion by Director Oster and other Directors, it was pointed out that an established Board policy, as on Publications, cannot be changed at the same Board meeting at which a policy revision is proposed. Accordingly, motion was made, seconded and carried for an enabling measure to permit change in the present Publications Policy at the February Board meeting, if requested by Chairman Morley's committee.

(14) Report on Headquarters Facilities

As requested at the August Board meeting, the General Manager presented a detailed study on the need for new reproduction equipment in the Central Office, and estimates on amortization by savings in outside printing costs. Following discussion, motion was made, seconded and carried, authorizing the purchase of offset-printing equipment to a maximum cost of $8,000 installed.

It was also brought out that processing memberships in minimum time might shortly require additional and more modern Addressograph (or equivalent) equipment. On motion made, seconded and carried, the General Manager was authorized to purchase such addressing equipment when necessary and advisable.

(15) Proposal of H. W. Dietert

The President presented a letter from past AFS Director H. W. Dietert offering to the Society Mr. Dietert's sand testing laboratory and library, now located in Texas, at some future date when no longer required by him. Without motion, the President expressed the Society's grateful appreciation of the offer and great interest in both the library and laboratory facilities.

(16) Tom Makemson Memorial Fund

The General Manager announced that the Institute of British Foundrymen had established a Tom Makemson Memorial Fund, in honor of their deceased and long-time Secretary, for the giving of scholarships. He pointed out that Mr. Makemson was also the Honorary and only Secretary of the International Committee of Foundry Technical Associations from the time of the Committee's inception, and recommended a donation to the fund by AFS in memory of Mr. Makemson. On motion made, seconded and unanimously carried, a donation of $100 to the Tom Makemson Memorial Fund was approved.

(17) 1956 International Congress

The General Manager stated that a number of members had inquired as to whether AFS intended to organize some form of "Foundry Tour" in 1956 to include attendance at the International Foundry Congress, at Duesseldorf, Germany, September 1-9. He recalled the heavy American attendance at the 1953 Congress in Paris, and the very meager American representation of 1954 in Italy, and 1955 in England. Without vote, it appeared to be the consensus of the Board that AFS should not undertake development of official participation in the 1956 International Congress.

(18) Proposed Student Redesign Contest

Director Pindell presented a proposal that AFS initiate and sponsor an annual "Student Redesign Contest" at AFS Student Chapter schools, involving the awarding of cash prizes for the redesigning by students of non-cast parts as castings. He stated that considerable interest had become evident in discussions held in the Pacific Northwest area, and asked for Board reaction. It was agreed, without vote, that further development of interest was necessary; that the Directors should sound out schools and Faculty Advisors of Student Chapters in their respective areas on the feasibility and workability of the proposed contest and report their findings to President Simpson, for recommendations to the February Board meeting.

(19) Proposed "Engineered Castings Show"

The General Manager presented a recommendation of the Finance Committee that AFS organize and operate an "Engineered Castings Show" in conjunction with the 1956 Castings Congress & Exhibit at Atlantic City, May 3-9. Basic intention of the proposed event would be to demonstrate the quality, utility and economy of modern cast metals to a specialized audience of design engineers and purchasers of metals for engineering purposes, thus helping cast metals producers to broaden their markets.

The General Manager stated that the idea of an exhibit by castings producers had been discussed on several occasions over a number of years, culminating in a suggestion by past Director H. W. Dietert. He stated that he had personally contacted the elected presidents and administrative officers of the gray iron, malleable, steel, non-ferrous and foundry equipment trade associations as to feasibility of the plan and estimates of support. Following these contacts, the Finance Committee recommended that AFS initiate an Engineered Castings Show on the following general plan:

(a) That the proposed Castings Show be held at the same time and place as the 1956 Castings Congress & Exhibit, but as a separate and distinct event.

(b) That all exhibit space at the proposed Castings Show be sold to producing foundries at specified rental rates, assignments of space to be made in the same manner as for the regular Exhibit, and all exhibitors receiving like treatment in the Official Program.

(c) That the same rules and regulations apply to the proposed Castings Show as to the regular Exhibits, with all power and other facilities equally available.

(d) That AFS provide free booth space in the proposed 1956 Castings Show for member organizations of the National Castings Council for the purpose of product development displays.

(e) That stress be laid on obtaining exhibits of (1) producers of all cast metals at the proposed Castings Show: Gray iron, steel, malleable, brass and bronze, and light metals; (2) all known modern casting methods: Sand casting, shell molding, investment casting, permanent molding, die casting, centrifugal casting; and (3) specific cast metals such as nodular iron, meehanite, semi-steel, specialty alloys, etc. Fundamental purpose: To demonstrate the fact that "The Shortest Distance between Raw Material and Finished Product is a CASTING."

(f) That AFS, recognizing the necessity of attracting *a new and additional attendance* for exhibitors at the proposed Castings Show, undertake a strong, specialized promotional campaign to attract design engineers and those who specify and purchase metals for engineering structures. Campaign to include direct mail, publicity releases, editorial coverage, correspondence stickers, exhibitor mail enclosures, announcements to engineering societies, and paid advertising in selected industrial trade journals.

(g) That exhibitors at the proposed Castings Show be accorded equal facilities for exhibitor and guest registrations as regular exhibitors.

(h) That AFS develop, specifically for the proposed Castings Show, a corollary technical program dealing with casting design, product conversions, and the inherent engineering value of cast metals.

In presenting the proposal, the General Manager analyzed in detail the advantages and possible disadvantages of staging the proposed event initially in 1956 at Atlantic City, including points raised in discussions with trade association officers. He concluded that the advantages definitely out-weighed possible disadvantages, especially as to: (1) importance of initial staging in conjunction with an old established and accepted event; (2) extra promotional effort due to the regular Congress & Exhibit; (3) complete and excellent facilities available at Atlantic City, including hotel accommodations; (4) value of both regular and specialized audiences to both supplier and foundry exhibitors in 1956; (5) ability of AFS to do the job successfully.

The General Manager stated, in response to question, that in his experience there remained sufficient time to undertake the Castings Show successfully in 1956. He presented estimates of income and expense in support of the statement that the event, if held in 1956, would be entirely self-supporting, and should produce substantial net income for other Society activities. In response to question, he expressed confidence in being able to attract a substantial additional audience of design engineers for a 1956 initial event.

Summing up, the General Manager stated that the proposed Engineered Castings Show should benefit the industry and the Society in several ways: (1) by offering a new and dramatic means of developing new castings business; (2) by emphasizing the importance of castings to all manufacturing; (3) by encouraging other foundry managers to improve their practices and products; (4) by offering a needed new vehicle to further emphasize the importance of castings design and "designing for castings;" (5) by giving the entire Castings Industry an opportunity, if approved for 1956, to put its best foot forward in a single, all-inclusive Castings Congress; (6) by providing additional revenue for AFS as pointed out in the General Manager's report at the 1955 annual meeting of the Board; (7) that a Castings Show in 1956 would violate no existing Board policy concerning exhibits; (8) that the intended AFS program of trade paper advertising to attract design engineers and purchasing men would, in itself, greatly benefit the industry by stressing the quality, utility and economy of modern cast metals to the entire engineering profession.

After extensive discussion by the Directors, the proposal for an AFS Engineered Castings Show was voted on in four motions, as follows:

(1) THAT AFS organize and operate an Engineered Castings Show as a further service to the castings industry. MOTION CARRIED.

(2) THAT the initial Engineered Castings Show be held in conjunction with the AFS Castings Congress & Exhibit in May 1956. MOTION FAILED.

(3) THAT the initial Engineered Castings Show be held in conjunction with an AFS non-exhibit Castings Congress in 1957. MOTION CARRIED.

(4) THAT the approved Castings Show be held in accordance with the general overall plan submitted by the General Manager. MOTION CARRIED.

Prior to the vote for staging the proposed event in conjunction with a non-exhibit Castings Congress in 1957, the General Manager stated that he could not recommend holding the initial event in Cincinnati for several reasons: (1) hotel accommodations considered insufficient for a combined Castings Congress and Castings Show; (2) lack of facilities for holding both exhibits and technical meetings in one location; (3) hotel exhibit space insufficient for Castings Show estimates; (4) commercial exhibit hall facilities below AFS normal standards.

(20) Suggested AFS Die Casting Division

Director Nelson announced that an "Association of Die Casting Engineers" had recently been organized in Michigan, and expressed the belief that the new group might be induced to affiliate directly with AFS, as an organization with strictly technical purposes. The consensus of the Board being favorable, the President requested Director Nelson to contact the new die casting group and to negotiate toward affiliation with AFS through formation of a Die Casting Division of the Society. Director Nelson was requested to report back to the February Board meeting.

There being no further business to be considered, the meeting was declared adjourned.

Approved:

BRUCE L. SIMPSON, *President*
December 12, 1955

Minutes

Special Policy Meeting — Education Division

Union League Club, Chicago — December 8, 1955

(1) ROLL CALL:

President Bruce L. Simpson, *presiding*
Vice-President Frank W. Shipley

F. G. Sefing
B. C. Yearley
Jess Toth
Frank C. Cech
B. L. Bevis
} representing Education Division

Wm. W. Maloney, *General Manager*
A. B. Sinnett, *Education Director*

(2) President Simpson called the meeting to order and stated that it was being held at the request of the Education Division for a survey and discussion of present and future AFS educational activities. He then turned the discussion over to Mr. Sefing, acting for the Education Division at the request of the Division Chairman, W. H. Ruten.

(3) Mr. Sefing reviewed briefly the history of AFS educational work for some years past, with emphasis on lack of consecutive and concentrated Staff direction. He expressed the concern of the Education Division over the appointment of Mr. Sinnett, employed in 1954 as Education Director, as assistant to the General Manager. He then stated that the Division's program requires more concentration on education at the Chapter level, encouragement of more Chapter educational programs, and definition of the term "education," insofar as AFS activity is concerned, to mean "practical education of high school, vocational school and in-plant students in the arts and sciences of foundry practice."

All present agreed that the points presented by Mr. Sefing represented a consensus of opinion. Mr. Yearley, as chairman of the Joint Committee on Foundry Vocational Training, stated that in his estimation the contemplated foundry training manuals could not be produced in any reasonable period of time except with specialized and compensated assistance. Mr. Cech urged concentration on fewer and more specific educational objectives. The AFS Board policies on Education, adopted in 1953, were reviewed.

(4) President Simpson stated that progress in AFS educational work to date has largely been determined by the degree of interest shown by Chapter education committees, whose personnel changes frequently and in many instances may have been appointed for reasons other than knowledge of or interest in the educational process. He stated that the AFS program could only be progressed in the future by Staff effort and direction, concentrating on a few carefully chosen targets rather than many broad and diversified objectives.

Mr. Simpson recommended an intensified program of fewer objectives and, it being the consensus that this was desirable, presented a 9-point Education program for AFS, in order of intended chronological development, as follows:

1. Continuation of National Apprentice Contests and broadening present base.
2. Development of proposed Education Seminar for Foundry Instructors in high schools and vocational schools to be held June 21-23, 1956 at Michigan State University, East Lansing, Mich.
3. Development and completion of agreed Foundry Training Manuals by selected writers, subject to review by Joint Committee on Foundry Vocational Training.
4. Development of Teaching Aids for foundry instructors in high schools, vocational schools and in-plant training programs.
5. Activation of Pangborn Scholarships, when available.
6. Development of proposed Student Redesign Contests, if considered feasible.
7. Development of AFS Foundry Correspondence Courses.
8. Possibility of summer Foundry Short Courses for plant supervisory personnel. (Similar to short courses held at colleges in agriculture, coaching, etc.)
9. Consideration of national Student Essay Contest . . . at outset, in AFS Student Chapter and F. E. F. schools.

It was agreed that AFS should not undertake at this time the development of summer Foundry Short Courses (Point No. 8); deferred for two years. It was also agreed that the suggested Essay Contest (Point No. 9) should be dropped as being more in the province of F. E. F. work at college levels.

After full discussion of the remaining points, it was agreed to recommend to the Education Division that this 7-point program be adopted.

(5) Mr. Simpson indicated that approximately $12,000 might be required to finance the Seminar and Training Manuals (Points 2 and 3), and that Board approval would be necessary for budgeting. He agreed to seek letter ballot approval of the Board for the program, as soon as accepted by the Education Division.

(6) For the proposed Instructors Seminar at Michigan State, the following general provisions were agreed upon: (a) Instructors would be invited to attend, their expenses to be paid by AFS only while at the university; (b) 20 places to be reserved for industrial plant classroom instructors; (c) invitations to be limited to instructors at schools where demonstration facilities are available for molding and pouring metals, and for pattern-making; (d) invitations to be originally limited to the five states of Michigan, Ohio, Indiana, Illinois and Wisconsin, except for certain specific schools in other states and unless sufficient acceptances from the original five states fail to develop.

(7) There being no further business to be considered, the meeting was declared adjourned.

Respectfully submitted,
A. B. SINNETT
Education Director

Minutes

Meeting of the 1955-56 Nominating Committee

Union League Club, Chicago — December 12, 1955

(1) **ROLL CALL:**

Presiding, Frank J. Dost, Chairman, as the immediate Past-President of AFS.

Collins L. Carter, Past-President of AFS

W. T. Shute (Rep. Region No. 1, Chapter Group D — Eastern Canada Chapter, and Steel.)

C. O. Schopp (Rep. Region No. 2, Chapter Group J — Central Indiana Chapter, and Malleable.)

W. D. McMillan (Rep. Region No. 3, Chapter Group K — Chicago Chapter, and Gray Iron.)

Earl White (Rep. Region No. 4, Chapter Group N — Corn Belt Chapter, and Supplies.)

Howard H. Havies (Rep. Region No. 5, Chapter Group R — British Columbia Chapter, and Gray Iron.)

(2) Nomination of President

On motion made, seconded and unanimously carried, the incumbent AFS Vice-President, Frank W. Shipley, was nominated President of the Society for the year 1956-57.

(3) Nomination of Vice-President

Following submission of names of candidates and on motion made, seconded and unanimously carried, Harry W. Dietert, Chairman of the Board, H. W. Dietert Co., Detroit, was nominated Vice-President for the year 1956-57.

(4) Nomination of Directors

By-Laws of the Society were read as relating to organization of the Board and Nominating Committee procedures. The Chairman stated that it is the intention of the By-Laws that each of the 18 Chapter Groups of the Society would have constant representation on the Board of Directors. He announced that the Nominating Committee was expected to nominate Directors from the following six Chapter Groups which otherwise would have no representation after May 1956:

Chapter Group D, Region 1 — Eastern Canada, Ontario.

Chapter Group E, Region 1 — Northwestern Pennsylvania, Canton District, Pittsburgh area.

Chapter Group H, Region 2 — Detroit, Toledo.

Chapter Group J, Region 2 — Central Indiana, Michiana.

Chapter Group K, Region 3 — Chicago.

Chapter Group R, Region 5 — British Columbia, Oregon, Washington.

The Chairman named those Chapters already represented by Directors serving beyond May 1956, and candidates submitted by all Chapters then were presented and discussed. On motions made, seconded and carried, the following Directors were nominated, each to serve 3-year terms:

Chapter Group D — Eastern Canada Chapter
Alex W. Pirrie, Foundry Supt., Standard Sanitary & Dominion Radiator, Ltd., Toronto, Ontario — representing Gray Iron.

Chapter Group E — Northwestern Pennsylvania Chapter
Roger W. Griswold, Erie Malleable Iron Co., Erie, Pa. — representing Malleable.

Chapter Group H — Toledo Chapter
Gerald R. Rusk, President, Freeman Supply Co., Toledo, Ohio — representing Supplies.

Chapter Group J — Central Indiana Chapter
R. V. Righter, Plant Mgr., Central Foundry Div., General Motors Corp., Danville, Ill. — representing Gray Iron and Malleable.

Chapter Group K — Chicago Chapter
Garnet P. Phillips, General Supervisor, Foundry Research, International Harvester Co., Chicago — representing Gray Iron.

Chapter Group R — British Columbia Chapter
Herbert Heaton, Foundry Supt., Letson & Burpee, Ltd., Vancouver, B. C. — representing Gray Iron, Brass & Bronze, Aluminum.

(5) It was agreed that the Secretary of the Society, as in previous years, would notify all nominees by wire for prompt acceptance, Presidential and Vice-Presidential nominees having accepted previously by telephone.

It was agreed, in the event of non-acceptance by any nominee, that the past Presidents on the committee were empowered to select alternate nominees following notification of intention to the committee.

(6) General

The Chairman stated that names of all candidates not acted upon by the Nominating Committee would be referred to the Board of Directors for consideration in the appointment of one additional National Director, as provided in the by-laws.

(7) The Chairman pointed out that additional nominations may be received by written petition signed by 200 members in good standing, any time 45 days prior to the date of the Annual Business Meeting, and that newly elected officers and directors take office on the day following the close of the Annual Meeting (Convention) of the Society.

(8) There being no further business to be considered, the meeting was declared adjourned.

Respectfully submitted,

FRANK J. DOST, *Chairman*
Nominating Committee

Minutes

Meeting of the 1955-56 AFS Board of Awards

Union League Club, Chicago — December 13, 1955

(1) ROLL CALL:
Past-President Fred J. Walls (1945-46), *presiding*
Past-President E. W. Horlebein (1949-50)
Past-President Walter L. Seelbach (1951-52)
Past-President Collins L. Carter (1953-54)
Past-President F. J. Dost (1954-55)
W. W. Maloney, Board of Awards Secretary

Absent:
Past-President W. B. Wallis (1948-49)
Past-President I. R. Wagner (1952-53)

(2) The Secretary was requested to extend to absentee members W. B. Wallis and I. R. Wagner regrets over their absence, and to Past-President Wood the hope of the Board of Awards for improved health.

(3) It was announced that, by agreement with the donor, the Simpson Gold Medal henceforth could be considered by the Board of Awards on the same basis of award frequency as all other medals of the Society. The Board of Awards expressed its appreciation to H. S. Simpson and B. L. Simpson for their cooperation.

(4) Number of Awards

On motion duly made, seconded and carried, the Board of Awards voted to bestow three Gold Medals of the Society and three Honorary Life Memberships (including the retiring President) in 1956.

(5) Selection of Gold Medalists

The meeting being open for receipt of recommendations, the Secretary first presented a list of nominees properly submitted in 1955 and in previous years. Members of the Board of Awards presented additional recommendations. After full discussion, the following motions were made, seconded and carried, recommending to the Board of Directors the awarding of Gold Medals of the Society in 1956:

(a) *To Charles C. Sigerfoos* (Professor of Mechanical Engineering, Michigan State University, East Lansing, Mich.) the Thomas W. Pangborn Gold Medal of AFS "For outstanding contributions to the Society and the Castings Industry, particularly in the field of education of engineering students for foundry careers."

(b) *To James S. Vanick* (Foundry Engineer, International Nickel Co., New York) the William H. McFadden Gold Medal of AFS "For outstanding contributions to the Society and for valuable service to the ferrous castings industry over a period of many years."

(c) *To Harold F. Bishop* (Metallurgist in charge of Foundry Research, U. S. Naval Research Laboratory, Washington, D.C.) the Joseph S. Seaman Gold Medal of AFS "For outstanding contributions in the field of castings research at the Naval Research Laboratories and for contributions to the Society."

On motion duly made, seconded and carried, nominations for 1956 Gold Medal awards were closed.

(6) Selection of Honorary Life Memberships

Following discussion, motion was made, seconded and carried, recommending that Honorary Life Memberships in the Society be awarded in 1956 as follows:

(a) *To Joseph C. Pendleton* (retired Supt. of Foundry Div., Newport News Shipbuilding & Dry Dock Co., Newport News, Va.), Honorary Life Membership in AFS "For outstanding contributions to the advancement of the art and sciences of metal casting, particularly in the field of Naval architecture requirements."

(b) *To William D. McMillan* (Metallurgical Supervisor, McCormick Works, International Harvester Co., Chicago), Honorary Life Membership in AFS "For outstanding contributions to the Society and the ferrous castings industry."

(c) *To Bruce L. Simpson* (President, National Engineering Co., Chicago), Honorary Life Membership in AFS on completion of his present term of office as President of the American Foundrymen's Society.

On motion duly made, seconded and carried, nominations for 1956 Honorary Life Memberships were closed.

(7) Finances

It was stated that the $5,000 Pangborn Gold Medal Award endowment had been invested in Government securities, the additional $1,000 of the endowment being retained in cash to defray the costs of art work, plaster molds, dies, etc.

Following analysis of Board of Awards cash reserves, it was determined that no additional investments need be made at this time. It was understood, however, that the Secretary of the Board of Awards was authorized to reinvest increments of $1,000 or more as accrued beyond the maximum of $1,500 desirable for annual award presentation expense.

(8) General Considerations

It was agreed that medals and life memberships again should be presented at the Annual Banquet of the Society during the 1956 Convention; that the Chairman of the Board of Awards would make all presentations; that histories of all recipients would be printed in the Banquet program, with each recipient receiving a leather-bound copy thereof.

It was agreed that the call for nominations of candidates to be considered by the 1956-67 Board of Awards would go only to the Board of Awards, Board of Directors, Chairmen and Vice-Chairmen of AFS Divisions and General Interest Committees, plus publicity in MODERN CASTINGS magazine.

Lists of nominees, as previously submitted for consideration, were studied and the Secretary instructed to prepare an up-to-date list for the 1956-57 Board of Awards.

(9) Notifications

It was agreed that recipients of all 1956 awards would be notified over the signature of the Chairman of the Board of Awards, with no publicity of any kind given to the recommendations until after prior release by the Society following letter ballot approval by the Board of Directors. It was agreed that the Secretary would obtain prompt action by the Board of Directors on such approval.

(10) Proposed New Type Award

Some concern was expressed over the frequency of awarding Honorary Life Memberships, it being the consensus that the prestige of this honor should be carefully guarded and, for the most part, the award confined to retiring Presidents of the Society and Gold Medalists. The thought was also expressed that, in many instances, the Society is unable to recognize particularly outstanding AFS service or technical developments except by the awarding of Gold Medals and Life Memberships.

It was therefore felt a third type of award should be considered, in the nature of a "Certificate of Merit," which might be awarded more frequently than either the Gold Medals or Life Memberships. Thus, the Society is at present unable to take any official recognition of such developments as shell molding, the CO_2 process, the "D" process and pressure molding, primarily because it is believed that such developments should be materially advanced and accepted before being considered for Gold Medal awards. It was pointed out that the proposed "Certificate of Merit" would bridge this gap and enable the Society to encourage more developments and greater service efforts to AFS.

Members of the Board of Awards agreeing on the principle of such an additional type of award, it was decided on motion made, seconded and carried, that the matter should be further considered at an adjourned meeting of the Board of Awards, date to be announced later.

(11) Special Recommendations

Past Presidents of AFS present expressed the belief that AFS should play a more prominent and continuous part in relations with the International Committee of Foundry Technical Associations, and expressed concern over poor representation of the American foundry industry at the International Congresses held in Italy in 1954 and England in 1955. They expressed the hope that the AFS Board of Directors would take appropriate steps to develop greater attendance at the 1956 International in Germany.

While realizing that such recommendation might be considered extraneous to a meeting of the Board of Awards, on motion made, seconded and carried, the past Presidents made the following recommendation:

That the AFS Board of Directors, in recognition of the work of the Technical Director, seriously consider sending him to attend the International Foundry Congress in Duesseldorf, Germany, in September 1956 as an official representative of AFS and preferably to present a technical paper thereto; and

That the Board of Directors should seriously consider extending an official invitation to the International Committee of Foundry Technical Associations to hold an International Foundry Congress in the United States in 1962.

(12) 1956 Meeting

It was agreed that the 1956 Annual meeting of the Board of Awards would be held in Chicago on approximately the same Tuesday as in 1955, with the Nominating Committee meeting preferably on the preceding Monday, but that these meetings should avoid the meeting dates of the National Association of Manufacturers. The Secretary was directed to arrange accordingly.

(13) There being no further business to be considered, the meeting of the Board of Awards was declared adjourned.

Respectfully submitted,
FRED J. WALLS, *Chairman*
Board of Awards

Minutes

Meeting of AFS Finance Committee

Union League Club, Chicago — January 25, 1956

(1) ROLL CALL:
President B. L. Simpson, *presiding*
Vice-President F. W. Shipley
Past President F. J. Dost
General Manager Wm. W. Maloney
Vice-Presidential Nominee H. W. Dietert

Guests:
Henry Parker, Harris Trust & Savings Bank, Chicago.
P. L. Mitchell, Harris Trust & Savings Bank, Chicago.
Jos. E. Tansill, Lybrand Ross Bros. & Montgomery, Chicago.

(2) AFS Investment Agent Trust Account

The President referred to a Harris Trust letter of Dec. 5, 1955, recommending that the ratio of common stocks held in the AFS Investment Agent account be increased from approx. 35% to 40% of the total market value of the account. He asked Harris Trust representatives to provide more data why the increase was recommended, and its proposed timing.

Mr. Parker stated that an analysis of various trust fund policies, made by the bank's Trust Committee in November 1954, showed that common stock ratios in most trust funds exceeded 33⅓% and expressed the view that the present inflationary spiral would continue. Consequently, the Trust Committee had decided to increase common stock ratio in trust funds under its control by 1960 . . . up to 40% by June 1956, and up to 50% by 1960. After review in June 1956 it may be decided, where the Harris Trust has full responsibility for the account, to increase common stock ratios to 45%, and to 50% by 1960 or earlier. Mr. Parker stated that approx. 75% of existing large trust funds now have a common stock ratio of 40% or more.

The General Manager stated that the AFS Investment Agent Trust account had been set up with three Board-approved policies in mind:

First, to protect the principal of invested sums, as an obligation of the Society;

Second, to provide sufficient liquidity as to make cash available for operating purposes as may be needed;

Third, to seek an income from investments commensurate with the first two policies.

The General Manager also stated that establishment of the Investment Agent Trust in 1952 was based on two primary considerations not stated in the Board resolution of approval:

(1) Acceptance of an operating principle, as stated in a letter from the Harris Trust dated July 8, 1953, whereby the common stock ratio would be built up to 33⅓% within approx. three years, this ratio being calculated to produce an investment income satisfactory to AFS.

(2) Acceptance of the Finance Committee's strong opinion that the Harris Trust should have full discretionary powers in administering the Investment Agent account, so as to relieve AFS officials of personal responsibility for the selection of specific securities, constant observance of fluctuating values, and careful analysis of market trends. It was understood that continuation of the AFS Trust should depend on AFS satisfaction with the bank's administration of the account.

Following full discussion, motion was made, seconded and carried, accepting the Harris Trust recommendation to increase the ratio of common stocks in the AFS Investment Agent account from 33⅓% to 40% as cash is made or becomes available to them.

(3) AFS Income Tax Classification

The President reviewed briefly efforts of the Society toward reclassification of AFS for income tax purposes under Sec. 501(c)(3) of the 1954 Internal Revenue Act. Mr. Tansill stated that there appeared to be two stumbling blocks which might have to be removed before reclassification could be accomplished: (1) the present AFS charter, which is too broad; (2) present AFS exhibit policy.

The President stated that, on learning of the Internal Revenue Dept. attitude, some fears were expressed as to whether continued AFS efforts toward tax reclassification might jeopardize the Society's present exemption status under Sec. 501(c)(6). He had therefore requested legal opinion from Robert F. Spindell, attorney, of Chicago, on three specific questions, as follows:

(a) That "pressing for tax exemption under 501(c)(3) would not jeopardize the Society's present tax exempt status as a business league."

(b) That rentals paid for the use of the exhibit space for the display of products does not constitute "unrelated business taxable income" for a 501(c)(3) organization.

(c) That exemption of AFS under 501(c)(3) could be accomplished, but that a "substantial" amount of time would be involved and two main bars would have to be removed: revision of the AFS charter, and a material change in AFS exhibit policy.

Ensuing discussion centered around three possibilities open to the Society in its efforts to obtain permanent reclassification under 501(c)(3) and qualify for acceptance of the preferred Pangborn scholarship fund, as follows:

(a) Continue efforts toward reclassification under 501(c)(3) by Dec. 31, 1956. The AFS Finance Committee unanimously agreed that no present activities of AFS should be given up or materially altered, and that AFS should accept no restrictions on those activities.

(b) Set up, as recommended, a separately incorporated fund or foundation of AFS for the handling of scholarship, research and other funds, and qualify such separate fund or foundation under Sec. 501(c)(3). This can be done even though the Society itself remains a 501(c)(6) organization. Organization and administration would require no drastic changes in the present AFS organization. Although 501(c)(3) status calls for a full year of operation prior to qualification, it might be accomplished in less time. . . . In the opinion of the Finance Committee this plan probably should be accomplished in any event for permanent flexibility of AFS progress and activities.

(c) Suggest that the proposed Pangborn scholarship fund remain in possession of the Pangborn Foundation and the original scholarship plan be set up to function as "The Pangborn Foundation Scholarship Program administered by AFS." It is not known whether such a suggestion would be acceptable, although the purposes behind the original offer would apparently be served. This plan also is in the nature of a short-term objective, and, even if tendered and accepted, the Finance Committee unanimously agreed that AFS should still work toward 501(c)(3) classification of the Society as a whole or of a foundation thereof.

In its discussion of three possible actions, the Finance Committee stressed 501(c)(3) classification as a primary objective of the Society. On motion made, seconded and carried, the Finance Committee requested Mr. Tansill to continue efforts to obtain 501(c)(3) classification of the Society and to report on June 1, 1956, as to whether further efforts may be worth while.

(4) There being no further business to be considered, the meeting was declared adjourned.

Approved:

BRUCE L. SIMPSON, *President*
February 8, 1956

Minutes

1955-56 AFS Board of Directors

Sea Island, Georgia — February 9 - 10, 1956

(1) ROLL CALL:

President Bruce L. Simpson, *presiding*
Vice-President Frank W. Shipley
General Manager Wm. W. Maloney
Assistant Secretary Ashley B. Sinnett

Directors (exp. 1956)	Directors (exp. 1957)
Frank J. Dost	Lewis H. Durdin
Thos. Kaveny, Jr.	Wm. M. Hamilton
Chester V. Nass	Wm. A. Morley
Wm. R. Pindell	Eugene R. Oeschger
G. Ewing Tait	Harold L. Ullrich

Directors (exp. 1958)
Harold C. Erskine
Charles W. Gilchrist
O. Jay Myers
Robert W. Trimble

Absent:

Directors	Charles E. Brust	(exp. 1956)
	Edw. C. Hoenicke	(exp. 1956)
	Martin J. Lefler	(exp. 1956)
	Frank C. Cech	(exp. 1957)
	B. George Emmett	(exp. 1957)
	Curtis C. Drake	(exp. 1958)
	Charles E. Nelson	(exp. 1958)
	Richard A. Oster	(exp. 1958)

(2) A quorum having been established, President Simpson called the meeting of the Board of Directors to order.

(3) Reading and Approval of Minutes

Minutes of the Board of Directors meeting held November 10-11, 1955 were briefed for information of the Board, having previously been approved by letter ballot.

Minutes of the Board of Awards meeting held December 13, 1955 were reviewed briefly for information of the Board, having previously been approved by letter ballot.

(4) Report of the General Manager

Membership: Total membership 12,002 on January 31, a net gain of 1037 or 9.5% over the 10,965 total of June 30, 1955, and exceeding the Society's fiscal year "target" of 12,000.

Director Hamilton requested that the assigned membership expiration date of Tennessee Chapter be made earlier in the year so as to permit work on delinquent members prior to the Chapter's final meeting. Change agreed to, without vote.

Chapters. Two new Chapters installed: No. 43 — Connecticut, No. 44 — Pittsburgh. Progress continuing toward affiliation of New England Foundrymen's Association

Convention & Exhibit: Exhibit space practically sold, technical program practically finalized, 47 papers preprinted as separate pamphlets (35) or in MODERN CASTINGS (12).

Publications completed and distributed:

"Symposium on Non-Destructive Testing"
"Principles of Metal Casting"
"Engineering Manual for Control of In-Plant Environment in Foundries"

Six major publications scheduled for early completion:
Die Casting Design and Practice.
Control of Welding Hazards.
Cast Metals Handbook.
Air Pollution Control Manual.
Foundry Safety Manual.
Foundry Molding Sand Practice.

A major publications problem: reducing present book stock inventories, by stronger promotional effort and perhaps price adjustments.

Research: 14 research projects under way, 8 included in 1955-56 budget, 6 requiring no budgeted funds. Major need henceforth: evaluation of results in terms of industry application.

Education. Program finalized for 1st AFS Foundry Instructors Seminar scheduled June 14-16; Apprentice Contest 1956 well advanced with entries equal to or exceeding 1955. Balance of 7-point Education program recommended, not yet accepted by Education Division.

S&H&AP. Work of Program Director at present primarily toward completion of engineering manuals on Dust Control & Ventilation, and Air Pollution; activity with foundry groups at Philadelphia, Beloit, Wis., New York, Ontario, and Portland, Ore., assisting in development of acceptable local codes governing air pollution and in-plant conditions.

Administration. AFS Headquarters building completed and accepted by the owners on recommendation of the owner's sole representative, F. C. Riecks. General Contractor and all sub-contractors paid in full and discharged.

On motion made, seconded and carried, report of the General Manager accepted and made a part of these minutes.

(5) Report on "Modern Castings"

While the enlarged format still is a source of some contention, both readership and advertising acceptance indicate pronounced improvement. Present editorial styling and presentation widely complimented, editorial coverage broadened to cover operating and management interests in addition to purely technical interest, readership inquiries increased remarkably.

Advertising acceptance definitely increased, 1956 space contracts showing a 15-20% increase over previous year. Improvement attributed to format advantages, better coverage of products and literature, stronger promotional and contact efforts, advertiser realization of increased readership. MODERN CASTINGS can no longer be ignored as a potent factor in the advertising-sales plans of any company whose products are useful to castings producers.

Effective January 1956, the Society's magazine dropped the name "American Foundryman," as recommended by the Board, and now is known solely as MODERN CASTINGS.

On motion made, seconded and unanimously carried, the Board gave a vote of confidence to C. G. Fuller, Managing Director of MODERN CASTINGS, for progress thus far reported.

(6) Report of Chapter Contacts Committee

Vice-President Shipley, chairman, thanked the Directors for a conscientious job of contacting the Chapters, and stated that their emphasis on membership work was producing obvious results. He urged completion of all contacts at the earliest possible time.

(7) Report of Finance Committee

The President requested Board confirmation of Finance Committee action approving a proposal of the Harris Trust & Savings Bank that the common stock ratio in the AFS Investment Agent Trust Account be increased from 33⅓% to 40%, with a probable further increase to 50% by 1960. Funds required, he stated, were available from matured and called securities. On motion made, seconded and carried, the Committee's action was approved.

It was pointed out that the Society's Investment Agent Trust account was authorized as a "discretionary trust," with the Harris Trust & Savings Bank exercising full discretion in its administration. It was accepted without vote that the Harris Trust be requested to consider purchase of Canadian common stocks with funds derived from matured Canadian government bonds as presently held.

(8) Report of Board Research Committee

In the absence of the Chairman, the General Manager reported briefly on various AFS research projects:

Brass & Bronze — new project on factors affecting pressure tightness of 85-5-5-5 alloy, assigned to University of Michigan. *Steel* — injection of deoxidizers in molten steel, at cooperating organizations. *Light Metals* — new work on mold design as affecting fundamental mechanics of metal flow, at Battelle Memorial Institute. *Malleable* — melting variables and properties of malleable iron, continuing at University of Wisconsin. *Gray Iron* — new study on application of the theory of gating and risering, at Case Institute of Technology. *Heat Transfer* — new study on heat loss of molten metal in the ladle, at Columbia University.

Sand — (1) properties of steel sands at elevated temperatures, continuing at Locomotive Finished Materials Co.; (2) properties of iron molding materials at elevated temperatures, continuing at University of Michigan and cooperating foundries; (3) metal penetration on dry cores with mold washes, at cooperating foundries; (4) testing procedures and equipment for sands at producer plants, at cooperating plants; (5) evaluation of mulling techniques, at cooperating plants; (6) testing procedures for evaluation of shell molding materials, at cooperating plants; (7) study of physical-chemical aspects of sand particles, a theoretical investigation.

(9) Report of F. E. F. Trustees

Directors Nass and Durdin reported briefly as AFS trustees to the Foundation, stating that F.E.F. enrollment now is the largest in its history, with all schools progressing the program's objectives.

(10) New Pittsburgh Chapter

The General Manager reported that letter ballot of the Board, on a petition for formation of a Pittsburgh Chapter of AFS, resulted in unanimous approval. The Chapter, No. 44, was officially installed January 16.

(11) Foundry Instructors Seminar

The General Manager reported that letter ballot of the Board, on a recommendation that AFS sponsor a Foundry Instructors Seminar at Michigan State University in June 1956, resulted in unanimous approval. The dates of June 14-16, while admittedly not best for maximum attendance, had been accepted rather than postpone the initial seminar for more favorable dates in 1957.

(12) Proposed Foundry Research Foundation

Chairman Dost reported that the proposal to establish a Foundry Research Foundation had been discussed with the several foundry trade associations which would be expected to cooperate, but that a majority had indicated they were not fully in accord with the proposal as submitted. He stated that the Board committee was in full agreement on the merits of such a program, but that additional study was necessary before final recommendations could be made to the Board.

(13) Proposed Die Casting Division

The President and General Manager reported on a meeting held February 3rd with Director Nelson and officials of the Society of Die Casting Engineers, to explore possible integration of the group into AFS, through a Die Casting Division and Staff technical assistance. Discussion indicated that policy and administrative problems involved, particularly as to territorial activity and relationships with the various groups serving the die casting field, required further investigation before definite recommendations could be made.

Accordingly, the President appointed the following committee to investigate the matter further and to formulate a recommendation for the Board meeting May 7: Erskine, Shipley, Nelson, Vice-Pres.-elect Dietert.

(14) Proposed Student Redesign Contest

Director Pindell, reporting on AFS Student Chapter schools asked to evaluate the merit of a proposed Student Redesign Contest, stated that while replies were not wholly in favor of the proposal, five schools still had not responded. The President requested that the remaining schools be contacted for reactions, for a full report to the May 7 meeting of the Board.

(15) Report on Publication Pricing

Chairman Morley reported that his committee, after considerable study of the problem of pricing AFS special publications to produce a greater income, had developed several points for consideration by the Publications Committee and Board;

(1) Members availing themselves of the Society's publication facilities should bear the major costs of publication, by fully reflecting preparation costs in selling prices. Costs should include allowance for obsolescence. Pricing to recognize the principle of member discounts off list (non-member) prices. Pricing policy must be flexible and acknowledge that the cost-price relationship cannot be constant for all publications.

(2) In the overall yearly operation, Special Publications should be produced at little or no cost to the Society.

(3) Publications pricing policies or formulae should be re-examined by the Publications Committee not less frequently than every three years.

(4) Minimum inventories should be maintained on all publications, printings generally not to exceed 2,000 copies (except the Cast Metals Handbook).

Chairman Morley indicated that his committee would meet jointly with the Publications Committee for further discussion.

(16) Report on AFS Tax-Exempt Status

The President reported at length on discussions held with tax accountants Lybrand, Ross Bros. & Montgomery, and with Robt. F. Spindell, attorney, on continuing efforts toward reclassification of AFS for tax-exempt purposes under Sec. 501(c)(3) of the 1954 Revenue Act. Existing stumbling blocks, including too broad provisions in the AFS charter, were pointed out. In response to question, the President stated that efforts to obtain reclassification would not, according to legal counsel, jeopardize in any way the Society's present tax-exempt status.

The President stated that, in the opinion of the Finance Committee and its tax and legal advisors, AFS might qualify itself for the Pangborn grant (and future grants) either by reclassification under Sec. 501(c)(3) or by setting up a separate AFS Foundation, itself tax-exempt under Sec. 501(c)(3) even though the Society as a whole may not be so exempt.

It was the consensus of the Board that Lybrand, Ross Bros. & Montgomery should continue efforts to obtain reclassification of AFS under Sec. 501(c)(3), and to report by June 1, 1956 success or failure.

On motion made, seconded and carried, minutes of the Finance Committee meeting held January 25, 1956, were approved.

(17) Appointment of New F. E. F. Trustee

The President stated that AFS, as a charter member of FEF, had the privilege of appointing two trustees to the Foundation, each to serve for two years, terms to overlap. Present trustees and terms: C. V. Nass, 1954-1956; L. H. Durdin, 1955-1957. In replacement of Director Nass, the President recommended appointment of Director O. Jay Myers. Approved on motion made, seconded and carried.

The President and Board expressed appreciation to retiring trustee Nass for his conscientious work with the Foundation during the previous two years.

(18) Appointment of New AFS Director

The President called attention to AFS By-Laws (Art. IX, Sec. 10) requiring that the Board of Directors appoint one National Director annually for a 3-year term of office. Vice-President Shipley and Director Durdin, appointed to explore available candidates, then placed in nomination the following:

A. V. Martens, President, Pekin Foundry & Mfg. Co., Pekin, Ill.

On motion made, seconded and unanimously carried, nominations were declared closed and A. V. Martens unanimously appointed National Director for the term 1956-1959.

(19) Recommendations of Board of Awards

Unofficial recommendations of past Presidents attending the Board of Awards meeting held December 13, as presented for Board consideration:

(1) That AFS be officially represented at the 1956 International Foundry Congress, and at all future International Congresses.

(2) That an official AFS delegate to the 1956 International Congress present personally an invitation from AFS to hold the 1962 International Foundry Congress in the United States.

On motions made, seconded and carried, the Board voted to appoint Vice-President-elect H. W. Dietert as official AFS delegate to the 1956 International at Düsseldorf, Germany, to present in person the AFS invitation for an "International" in the United States.

(20) Organization of 1956-57 Board of Directors

Vice-President Shipley, taking the floor as President-elect for 1956-57, pointed out the desirability of early Board organization, particularly the election of Regional Vice-Presidents for the coming year. He requested the incumbent Regional Vice-Presidents (Regn. 1 — Kaveny, Regn. 2 — Lefler, Regn. 3 — Nass, Regn. 4 — Durdin, Regn. 5 — Pindell) to ballot hold-over and incoming Directors for recommendations for new Regional Vice-Presidents, to be elected at the Board meeting May 7.

(21) There being no further business to be considered, the meeting was declared adjourned.

Approved:

BRUCE L. SIMPSON, *President*
April 19, 1956

Minutes

Final Meeting 1955-56 AFS Board of Directors

Hotel Claridge, Atlantic City, N. J. — May 7, 1956

(1) ROLL CALL:

President Bruce L. Simpson, *presiding*
Vice-President Frank W. Shipley
General Manager Wm. W. Maloney
Assistant Secretary Ashley B. Sinnett

Directors (exp. 1956)
Charles E. Brust
Frank J. Dost
Edw. C. Hoenicke
Thos. Kaveny, Jr.
Chester V. Nass
G. Ewing Tait

Directors (exp. 1957)
Frank C. Cech
Lewis H. Durdin
B. George Emmett
Wm. M. Hamilton
Wm. A. Morley
Harold L. Ullrich

Directors (exp. 1958)
Curtis C. Drake
Harold C. Erskine
Charles W. Gilchrist
O. Jay Myers
Charles E. Nelson
Robert W. Trimble

Officers and Directors-elect, as Observers:
Harry W. Dietert, Vice-Pres.-elect

Directors-elect (exp. 1959)
Roger W. Griswold
Herbert Heaton
A. V. Martens
Alex Pirrie
R. Verne Righter
Gerald R. Rusk

Absent:

Directors W. R. Pindell (exp. 1956), E. R. Oeschger (exp. 1957), R. A. Oster (exp. 1958), G. P. Phillips (exp. 1959).

(2) Call to Order

A quorum having been established, President Simpson called the meeting to order. He stated that newly-elected Director Garnet P. Phillips, unable to attend because of sudden illness, was well on the road to complete recovery.

(3) Reading and Approval of Minutes

Minutes of the meeting of the Board of Directors held February 9-10, 1956, were briefed, having previously been approved by letter ballot. Also briefed were the following letter ballots of the Board, both approved: (a) March 10, 1956, accepting petition for a New England Chapter; (b) April 10, 1956, for election of Regional Vice-Presidents.

Director Morley, as Chairman of the special Board Committee on Publications Pricing requested that Minutes of the February Board meeting be revised to properly indicate the Board's feeling, previously expressed, as to the importance of this co-operative investigation of publications pricing. So ordered.

President Simpson expressed his appreciation for the Board's vote in granting him sufficient latitude of action as to make possible immediate acceptance and installation of the New England Chapter in April.

(4) Report of the General Manager

Membership: Total membership April 25 was 12,435 (a new all-time high), compared with 12,146 on March 31 and 10,965 on July 1, 1955; total of 20 Chapters exceeded their assigned "Targets."

Chapters: Installation of New England Chapter on April 11 as Chapter No. 45.

Research. New Gray Iron project developing at Cornell University on pictorial presentation of solidification phenomena.

Technical Programs. "National Committee Week" scheduled for week of June 4-8 in Chicago, for review of activities of past year and to plan 1956-57 programs. Chairman of Technical Council, Frank B. Rote; President-elect Shipley, vice-chairman.

Publications. New S&H&AP publication, "Engineering Manual for Control of In-Plant Environment in Foundries," distributed and well received.

Education Seminar. First annual Foundry Instructors Seminar scheduled June 14-16 at Michigan State University. Vice-Pres.-elect Dietert, chairman; Prof. W. H. Ruten, chairman of Education Division, co-chairman.

Administration. Resignation of H. F. Scobie, Editor of MODERN CASTINGS, announced effective July 1, to become Asst. Executive Secretary, Non-Ferrous Founders' Society.

On motion made, seconded and carried, the report of the General Manager was accepted.

(5) Proposed AFS Foundation

President Simpson, reporting further on AFS efforts toward obtaining reclassification of non-tax status, stated that an opinion was due approximately June 1 from Lybrand Ross Bros. & Montgomery of Chicago, but that there seemed little likelihood that the Society as a whole could obtain reclassification as a 501(c)(3) organization under the Internal Revenue Act of 1954 as presently interpreted. He therefore stated that, if the opinion of the tax consultants was negative, the possibility of setting up a "Foundation" within the structure of AFS would be thoroughly explored to the end of recommendations to the Annual Board Meeting.

(6) Proposed Student Redesign Contest

President Simpson stated that no report could be presented on reactions of various universities to the proposed Student Redesign Contest. The General Manager was instructed to contact Director Pindell for a final report, the matter being tabled until the Annual Board Meeting.

(7) Proposed Die Casting Division

President Simpson outlined briefly the background to consideration of a Die Casting Division of the Society, and contacts with the Society of Die Casting Engineers toward possible AFS amalgamation. Director Nelson stated that membership of the S.D.C.E. totaled 278 in March; that AFS should show strong interest in greater die casting activities; that the Die Casting Committee of the Light Metals Division was planning a joint meeting with S.D.C.E.; and that AFS should indicate prime interest in that organization.

Director Erskine reported on early negotiations between S.D.C.E. and the American Die Casting Institute, trade association of the die casting industry. He stated that, as a principle, no organization should be allowed to operate independently within the framework of AFS. A consensus of the Board agreed to seek greater interest of die casters through AFS membership and other activities, but not as independent organizations.

On motion made, seconded, amended and carried, the Board approved creation of a Die Casting Division of AFS, subject to approval of the Light Metals Division and its Die Casting Committee.

(8) Proposed AFS Education Program

President Simpson reviewed correspondence with the Education Division following his suggestion that AFS concentrate on a Board-approved 7-point program in chronological order of undertaking, as follows:

1. Continuation of National Apprentice Contests.
2. Initiation of Foundry Instructors Seminar.
3. Development of In-Plant Training Manuals.
4. Development of better Teaching Aids for Secondary Schools.
5. Initiation of Scholarships for Secondary School Graduates (as funds made available).
6. Development of Student Redesign Contest (if deemed feasible).
7. Development of Foundry Correspondence Courses.

The President stated that the program thus far had not been accepted by the Education Division. Discussion developed the following points: that AFS should concentrate on vocational training with a full-time Staff "Director of Vocational Education"; that the new Board should seriously consider employment of such a Director; that educational activities could best be carried out by a Staff Director rather than by volunteers; and that the Education Division should be asked to define a proper vocational education program if the proposed 7-point program is considered inacceptable.

(9) Recommendation of Board of Awards

President Simpson stated that the 1955-56 Board of Awards, meeting in adjourned session, had prepared several important recommendations for additional AFS awards. In the interest of time, recommendations tabled for presentation at the Annual Board Meeting.

(10) Selection of AFS Auditor

President Simpson stated that the by-laws empowered him to select certified public accountants to audit the financial books of the Society. He expressed the belief, however, that Board action was preferable to Presidential selection. On motion made, seconded and carried, the firm of George W. Rountree & Co., Chicago, was selected to make the audit as of June 30, 1956.

(11) Recommendations of Retiring President

In the interest of continuity, President Simpson offered the following recommendations for consideration by President-elect Shipley and the 1956-57 Board of Directors:

(1) *Convention and Exhibits.* Go ahead full blast with the Engineered Castings Show in 1957 and a permanent Exhibit Manager.

(2) *Membership.* Set the goal for membership in 1956-57 at 13,500 with a permanent field membership contact.

(3) *Chapters.* Make strong effort to obtain Chapters in the Reading and Williamsport areas, and Salt Lake City.

(4) *Financial.* Transfer only $40,000 into the invested reserve, the balance above immediate cash needs to be held in short-term treasury notes.

(5) *"Modern Castings."* Recommend no change in policy as to size or personnel except in replacement of the resigned Editor.

(6) *Research.* Recommend AFS go ahead and set up a separate foundation as a 501(c)(3) tax-exempt organization, but as a totally owned subsidiary of the AFS with an interlocking Board of Trustees; that money voted for education and research by the Board be transferred to this foundation; that outside societies be invited to join on a voluntary basis, and that AFS explore its own facilities, including the possibility of building a research laboratory.

(7) *Technical.* Recommend the continued travel of the Technical Director, particularly to all Regional Conferences. Recommend continual approachment, on friendly basis, to other societies doing technical work and refusal to engage in any words of controversy with these societies.

(8) *Safety, Hygiene & Air Pollution.* Recommend continuance of a Director of Safety, Hygiene & Air Pollution Control and coverage with public bodies such as the public health agencies in the twelve or fifteen highly industrialized States to make them aware of the work AFS has done. The possibility here is to tie-in with Chapters involved.

(9) *Education.* Recommend that the Education Division give AFS a complete program stripped of generalities, and that a full-time Director of Education be employed.

(10) Recommend that AFS allow the report of Lybrand, Ross Bros. & Montgomery to come through on June 1, but under no conditions to jeopardize major revenue sources as they now stand.

(11) Recommend, in conjunction with the Pangborn Scholarship Fund, a minimum of $5,000 a year for scholarships at the secondary school level when the money becomes available to AFS.

(12) Recommend, after investigation completed on the proposed Castings Redesign Contest, that this matter be turned over to the Education Division for action prior to any final approval by the Board of Directors.

(12) International Congress Representation

In executive session of the Board, motion was made, seconded and carried appropriating the sum of $2,000.00 from 1955-56 excess income for expenses of the General Manager to attend the 1956 International Foundry Congress in Germany as an official representative of the Society.

(13) Introduction of New President

Before adjourning the final meeting of the 1955-56 Board of Directors, President Simpson presented the gavel of office to President-elect Frank W. Shipley and expressed confidence that the Board would give the new President of the Society its full and unqualified support.

(14) Adjournment

There being no further business to be considered by the retiring Board, the meeting was declared adjourned.

Approved:

BRUCE L. SIMPSON, *President*
July 23, 1956

Minutes

First Meeting 1956-57 AFS Board of Directors

Hotel Claridge, Atlantic City, N. J. — May 10, 1956

(1) ROLL CALL:

President Frank W. Shipley, *presiding*
Vice-President Harry W. Dietert
General Manager Wm. W. Maloney

Directors (exp. 1957)
Frank C. Cech
Lewis H. Durdin
B. George Emmett
Wm. M. Hamilton
Wm. A. Morley
Bruce L. Simpson
Harold L. Ullrich

Directors (exp. 1958)
Curtis C. Drake
Harold C. Erskine
C. Wm. Gilchrist
O. Jay Myers
Charles E. Nelson
Robert W. Trimble

Directors (exp. 1959)
Roger W. Griswold
Herbert Heaton
A. V. Martens
Alexander Pirrie
R. Verne Righter
Gerald R. Rusk

Retiring Directors (as Observers)
Frank J. Dost
Edw. C. Hoenicke
Thos. Kaveny, Jr.
G. Ewing Tait

Absent:

Directors E. R. Oeschger (exp. 1957), R. A. Oster (exp. 1958), G. P. Phillips (exp. 1959).

(2) Call to Order

President Shipley expressed appreciation for the honor accorded him to serve as President of AFS and stated that it would be his intention to give all Directors opportunity to express opinions on any subject, but that the main consideration of all should be: "Is it for the good of the Society?" The President then introduced Vice-President Dietert.

(3) Election of Regional Vice-President

President Shipley announced that new Regional Vice-Presidents had been elected by letter ballot of the Board, as follows:

Region No. 1 — Harold L. Ullrich
Region No. 2 — Charles E. Nelson
Region No. 3 — O. Jay Myers
Region No. 4 — Wm. M. Hamilton
Region No. 5 — B. George Emmett

(4) Formation of Executive Committee

The President stated that under the by-laws it was necessary to elect three Regional Vice-Presidents who, together with the President, Vice-President and immediate past President, would form an Executive Committee of six members. He stated that it was his intention to continue the practice of inviting all Regional Vice-Presidents to attend meetings of the Executive Committee. By secret ballot of the Board, the 1956-57 Executive Committee was formed as follows:

President Frank W. Shipley, *Chairman*
Vice-President Harry W. Dietert
Past President Bruce L. Simpson
Regional Vice-Pres. Chas. E. Nelson (Regn. 2)
Regional Vice-Pres. O. Jay Myers (Regn. 3)
Regional Vice-Pres. Wm. M. Hamilton (Regn. 4)

(5) Appointment of N.C.C. Representatives

The President stated that AFS, as a charter member of the National Castings Council, was obliged to name two official representatives, one of which must be the incumbent President. On motion duly made, seconded and carried, Vice-President Dietert was named, together with President Shipley, to represent the Society.

(6) Appointment of Board Committees

President Shipley stated that, as empowered under the by-laws, it was his desire to appoint Board and other official committees for 1956-57. On motion made, seconded and carried, the following committees were approved:

Executive Committee (as elected)
F. W. Shipley, Chairman
H. W. Dietert, Vice-Pres.
B. L. Simpson, Past Pres.
Chas. E. Nelson
O. Jay Myers
Wm. M. Hamilton

Finance Committee (self-appointive)
F. W. Shipley, Chairman
H. W. Dietert
B. L. Simpson
Wm. W. Maloney, Treas.

Board of Awards (self-perpetuating)
W. B. Wallis (1948-49), Chairman
E. W. Horlebein (1949-50)
W. L. Seelbach (1951-52)
I. R. Wagner (1952-53)
Collins L. Carter (1953-54)
Frank J. Dost (1954-55)
B. L. Simpson (1955-56)

Chapter Contacts Committee
H. W. Dietert, Chairman
H. L. Ullrich — Region 1
C. E. Nelson — Region 2
O. Jay Myers — Region 3
W. M. Hamilton — Region 4
B. George Emmett — Region 5
All Members of Board, members of Committee.

Publications Committee
S. C. Massari, Chairman
W. A. Morley, Board Liaison
B. N. Ames
D. L. Colwell
Thomas T. Lloyd
J. A. Rassenfoss
C. A. Sanders

Board Research Committee
C. E. Nelson, Chairman
L. H. Durdin
H. C. Erskine
O. Jay Myers
E. R. Oeschger
G. P. Phillips
H. L. Ullrich

Board Liaison to S&H&AP Steering Committee
R. V. Righter
W. M. Hamilton

Board Liaison to Education Division
F. C. Cech
R. A. Oster

Annual Lecture Committee
R. F. Thomson, Chairman
G. P. Halliwell
C. F. Joseph
S. C. Massari
W. E. Sicha

Technical Council
1956 —
Frank B. Rote, Chairman
F. W. Shipley, Vice-Chairman
1957 —
Frank B. Rote, Chairman
H. W. Dietert, Vice-Chairman

National Castings Council Representatives
F. W. Shipley
H. W. Dietert

AFS Trustees to Foundry Educational Foundation
L. H. Durdin (1955-57)
O. Jay Myers (1956-58)

Official International Representatives
H. W. Dietert
Walter L. Seelbach
Vincent Delport (Europe)
Wm. A. Gibson (Australia)
Wm. W. Maloney

(7) Chapter Contacts Organization

Vice-President Dietert, as chairman of the Chapter Contacts Committee, completed Director assignments for the year (accepted by all Directors present), as follows:

Harry W. Dietert — Texas, Texas A. & M. College.
Frank C. Cech — Canton, Northeastern Ohio.
C. C. Drake — Corn Belt, Timberline.
Lewis H. Durdin — Birmingham, University of Alabama, Mexico.
B. George Emmett — Northern California, Southern California.
Harold C. Erskine — Pennsylvania State University, Pittsburgh.
C. Wm. Gilchrist — Central Ohio, Cincinnati, Ohio State Univ.
Roger W. Griswold — Western New York, Northwestern Pennsylvania.
W. M. Hamilton — Mid-South, Tennessee.
Herbert Heaton — British Columbia, Oregon, Oregon State College, Washington.
A. V. Martens — Central Illinois, Quad City.
William A. Morley — Chesapeake, Philadelphia.
O. Jay Myers — Twin City.
Charles E. Nelson — University of Michigan, Michigan State Univ., Saginaw Valley, Western Michigan, Central Michigan.
Eugene R. Oeschger — Central New York, Eastern New York, Rochester.
Richard A. Oster — Northern Illinois-Southern Wisconsin, Wisconsin, University of Wisconsin.
Garnet P. Phillips — Chicago.
Alexander Pirrie — Eastern Canada, Ontario.
R. Verne Righter — Central Indiana, University of Illinois.
Gerald R. Rusk — Detroit, Michiana, Toledo.
R. W. Trimble — Mo-Kan, St. Louis, University of Missouri School of Mines, Tri-State.
Harold L. Ullrich — Brooklyn Polytechnic Institute, Connecticut, Massachusetts Institute of Technology, Metropolitan, New England.

(8) Selection of 1959 Convention City

The General Manager stated that it was necessary to select promptly a 1959 Convention City in order to make proper arrangements for meeting rooms, hotel accommodations, etc. He stated that Chapter invitations had been received from Eastern Canada for a non-exhibit convention in Montreal, and from Chesapeake for a non-exhibit meeting in Washington, D.C.

Director Simpson suggested that the 1959 meeting be held in Boston, in view of the New England Chapter, good facilities for an Engineered Castings Show, and because AFS has not met in Boston since 1917.

Retiring Director Tait recommended that Montreal be dropped from consideration in 1959, due to lack of facilities for an Engineered Castings Show, and that Toronto be considered for 1961 on the basis of the success of the 1957 Engineered Castings Show. He stated that American exhibits would be required to enter Canada under bond, and would have to clear U. S. customs on re-entry.

On motion made, seconded and carried, the Board voted to hold the 1959 Convention in Boston with an Engineered Castings Show, and instructed the General Manager to express appreciation to Director Tait and the Eastern Canada Chapter for withdrawal of Montreal, and to propose Toronto for an AFS Convention at a later date.

(9) Selection of 1960 Convention City

The General Manager, emphasizing the necessity of booking exhibit halls four to six years in advance to be assured of desirable dates and facilities, asked for decision on location of the 1960 Castings Congress and Show. On request, he outlined advantages and disadvantages of Atlantic City, Chicago, Cleveland, Detroit, New York and Philadelphia.

He stated that it had long been a policy of AFS to alternate its Conventions between various foundry areas. With the 1958 Castings Congress & Show scheduled for Cleveland, and projected new auditoriums in Chicago and Detroit uncertain as to completion by 1960, he recommended meeting either in Philadelphia or Atlantic City in 1960.

It being the consensus that meetings in Atlantic City should be held at intervals of more than four years, on motion duly made, seconded and carried the Board voted to hold the 1960 Castings Congress and Show in Philadelphia, subject to concurrence by the AFS Committee of Exhibitors.

(10) Chapter Officers Conference

President Shipley reminded all new Directors of their obligations to attend the 1956 Chapter Officers Conference, and stated that Directors attending would be seated with officers of Chapters assigned to them for Board contacts.

(11) There being no further business to be considered, the meeting was declared adjourned.

Approved:

FRANK W. SHIPLEY, *President*
July 19, 1956

13th ANNUAL CHAPTER OFFICERS CONFERENCE

LaSalle Hotel, Chicago — June 28 - 29, 1956

PROGRAM

June 28

9:30 A.M. Call to Order.....Gen. Mgr. Maloney
9:31 A.M. Welcome to Conference .....Gen. Mgr. Maloney
9:34 A.M. Introductions.....By the Conference
9:44 A.M. Your Society and Its Work.....Chairman Shipley
9:55 A.M. AFS as a Technical Society.....Tech. Dir. Heine
10:40 A.M. The 4 Essential Committees.....Gen. Mgr. Maloney
10:50 A.M. Developing the Chapter Program.....Panel: (Bob Colton, Ken Smith, John O'Meara, Frank Kellam, Dick Westendorf)
12:15 P.M. Luncheon
1:15 P.M. The F.E.F. Place in the Foundry Industry .....E. J. Walsh
2:00 P.M. The Chapter Contacts Committee.....Vice-Pres. H. W. Dietert
2:15 P.M. Building the Membership.....Gen. Mgr. Maloney
3:40 P.M. Successful Regional Conference .....Panel: (Geo. Barker, Clyde Armstrong, Fred Menzel, John Drenning, Alex Barczak)
4:35 P.M. Organization of AFS Board.....Chairman Shipley
4:50 P.M. The AFS Nominating Committee, Gen. Mgr. Maloney
5:30 P.M. Social Hour
6:30 P.M. Annual Conference Dinner
Presiding: Frank W. Shipley
Guest Speaker: Dave Condon, Sports Editor, Chicago Tribune. "You Don't Have To Be Crazy."
8:15 P.M. Special Delegate Meetings

June 29

9:02 A.M. Chapter Educational Programs .....Panel: (Bill McFatridge, Harold Henderson, John Roth, Morris McQuiggan)
10:00 A.M. SH&AP Control Program .....Dir. Weber
10:35 A.M. AFS Financing and Finances.....Chairman Shipley
10:50 A.M. Conventions and Exhibits .....Gen. Mgr. Maloney
11:00 A.M. "MODERN CASTINGS" .....Mng. Dir. Fuller
11:15 A.M. Organization for the Chapter Job, Gen. Mgr. Maloney
11:45 A.M. How To Use Conference Kits....Chairman Shipley
12:15 P.M. Getaway Luncheon

ATTENDANCE

AFS President & Conference Chairman — F. W. Shipley, Foundry Mgr., Caterpillar Tractor Co., Peoria, Ill.

AFS Vice-President — H. W. Dietert, Chairman of the Board, H. W. Dietert Co., Detroit, Mich.

AFS General Manager — Wm. W. Maloney.

Guests — R. W. Schroeder, University of Illinois, Chicago. E. J. Walsh, Exec. Director, Foundry Educational Foundation, Cleveland, Ohio.

Chapter Officers

BIRMINGHAM — Chairman J. F. Drenning, Sales Engr., Kerchner Marshall & Co., Birmingham, Ala. Treasurer T. L. Turner, Asst. Supt., American Brake Shoe Co., Birmingham, Ala.

BRITISH COLUMBIA — Chairman C. C. Smith, Mgr., General Metals, Vancouver, B. C. Vice-Chairman & Program Chairman E. C. J. Bird, President, Bird Aluminum Fdy Ltd., Vancouver, B. C.

CANTON DISTRICT — Chairman W. W. Snodgrass, Planning Engr., Massillon Steel Castings Co., Massillon, Ohio. Vice-Chairman & Program Chairman L. A. Weaver, Prod., Met., United Engineering & Foundry Co., Canton, Ohio.

CENTRAL ILLINOIS — Chairman K. M. Smith, Staff Engr., Caterpillar Tractor Co., Peoria, Ill. Vice-Chairman & Program Chairman G. Lloyd, Treasurer, Brass Foundry Co., Peoria, Ill. Secretary L. Kinsinger, Caterpillar Tractor Co., Peoria, Ill.

CENTRAL INDIANA — Chairman R. H. Brookes, Foundry Supt., Link Belt Co., Indianapolis, Ind. Secretary W. Boyd, Dist. Sales Mgr., Mexico Refractories, Indianapolis, Ind.

CENTRAL MICHIGAN — Chairman G. D. Strong, Fdy. Engr., Battle Creek Foundry Co., Box 278, Battle Creek, Mich. Vice-Chairman & Program Chairman R. D. Dodge, Sales Engr., Archer-Daniels-Midland Co., Fdy. Products Div., Detroit, Mich.

CENTRAL NEW YORK — Vice-Chairman & Program Chairman N. W. Meloon, Pres., Meloon Bronze Fdy. Inc., Syracuse, N. Y.

CENTRAL OHIO — Chairman T. M. Cusack, Fdy. Supt., O. S. Kelly Co., Springfield, Ohio. Vice-Chairman & Program Chairman J. R. McConnell, Treas. & Sls. Mgr., The Globe Steel Abrasive Co., Mansfield, Ohio.

CHESAPEAKE — Chairman J. H. Schaum, Fdy. Met., National Bureau of Standards, Washington, D. C. Vice-Chairman & Program Chairman L. H. Gross, Met., American Standard Corp., Baltimore, Md. Chairman Southern Section L. A. Howell, Supt., Fdy. Div., Newport News Shipbuilding & Dry Dock Co., Newport News, Va.

CHICAGO — President R. P. Schauss, Chicago Mgr., Werner G. Smith, Inc., Chicago, Ill. Vice-Chairman & Program Chairman W. O. McFatridge, Supv., Fdy. Rsch. Lab., International Harvester Co., Chicago, Ill.

CINCINNATI DISTRICT — Chairman R. J. Westendorf, Vice-Pres., Dayton Casting Co., Dayton, Ohio. Vice-Chairman & Program Chairman E. J. James, Sales Engr., The Dayton Oil Co., Dayton, Ohio.

CONNECTICUT — Program Chairman J. Baxter, Malleable Fdy. Supt., Malleable Iron Fittings Co., Branford, Conn.

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DETROIT — Chairman J. Nabrezny, Sales Mgr., Michigan Malleable Iron Co., Detroit, Mich. Vice-Chairman & Program Chairman C. W. Yaw, Asst. Fdy. Supt., Cadillac Motor Car Div., Detroit, Mich.

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1956

CHARLES EDGAR HOYT MEMORIAL LECTURE

MARKETING YOUR PRODUCT

By

S. C. Massari*

Having spent most of my life in the foundry industry and having a real attachment for it, I felt highly honored when invited to present the Charles Edgar Hoyt Memorial Lecture – an assignment which I deeply appreciate. I am proud to have the privilege of addressing fellow foundrymen especially in honor of a man so highly respected in the industry. His life was devoted to the foundry industry in all its ramifications, but particularly in promoting the utility of castings. It is quite appropriate for me to select as my subject, the marketing of our industry's product, castings.

Most of the past Charles Edgar Hoyt Memorial Lectures have presented the results of extensive investigations of cast metals, foundry processes, or equipment. I would like to present some observations relating to marketing of castings, the backbone of the industry in our competitive economy.

We, as foundrymen, thoroughly believe in the merits of castings but it is equally important that our customers recognize their value. In our highly competitive market we must keep our product before the eyes of the purchaser, develop improved manufacturing techniques, and maintain a competitive selling price. We must also improve casting quality, develop new applications, and satisfy increasingly rigid specifications to meet competition from other manufacturing methods and materials.

Customers of tomorrow will expect even more from us and if we are to have their business, castings will have to meet their needs or our competitors' product will. It is well to remember that the customers' needs today may be totally different tomorrow, emphasizing the need for close observation and liaison with the purchaser. Obviously, only by satisfying the real needs and wants of a customer can the casting industry continue on a profitable selling basis. Rendering the greatest service becomes the means for making the greatest profit.

The casting industry had its beginning some four thousand years before the beginning of the Christian era and has continued because it is basic to our society. Since that time foundry technology made great advances, especially in the last twenty years. But, it must advance at a more accelerated rate if the foundry is to exist in the future by meeting other technical competition.

Technological Status

In the past, the making of castings was largely an art while today, it is developing into an exact science. The industry comprises a group of men who are willing to exchange ideas and welcome plant visitation by even their closest competitors. Few industries enjoy such a fine spirit of cooperation.

Foundrymen are also fortunate to have technical groups such as the American Foundrymen's Society and the several industry trade groups where management and technologists gather for the exchange of information relating to metals, processes and manufacturing equipment. Such a program must continue and deserves the best support of management. Tremendous developments in industry are taking place creating a serious challenge to our industry. Research is one of the most effective means in avoiding obsolescence by a competitive product or process which serves in a new, better or cheaper way.

In short, through research and process control, the industry must energetically pursue a program aimed in producing the finest product at the lowest cost. Small or large, within its means, every foundry should employ well trained technologists and allocate some portion of its budget to research in the fields of better raw materials, process control, and new and improved castings at a lower cost.

New Market and Product Development

The greatest need for research and new product development may not be to increase profits but to maintain them. Obsolescence of existing castings applications results in diminishing tonnage. Typically, the chilled iron wheel is being replaced by either the wrought- or cast-steel wheel; large quantities of steel castings formerly used on steam locomotives are no longer required with the advent of diesel-electric motive power; the modern domestic gas range, made

*Director of Research, National Engineering Co., Chicago, Ill.

of stampings, has totally outmoded the large quantities of stove plates once made in foundries.

Today, even such a simple casting as a sash weight is being supplanted by various forms of spring-loaded devices. Our industry must improve foundry processes to retain existing markets, recapture at least some of those already lost and, equally important, discover and develop new markets. This is progress!

Mr. Ralph J. Cordiner, President of the General Electric Company, recently stated, "In the next five years, 90 per cent of all industrial production will emanate from the research and development work currently in process." In an era of such rapid industrial changes, the foundry must be constantly alert for new applications for castings to counteract the loss of business because of obsolescence.

We must do more than just wonder how the foundry will prosper in this Atomic Age. Many of the progressive foundrymen are alert to the necessity for new product development and the application of casting methods so as to do the job better by lowering the manufacturing cost or producing a structural member which performs better in ultimate service.

Through the cooperation and courtesy of a number of friends in the industry, it has been possible to assemble a variety of typical examples of where a change in design and the application of good foundry engineering has replaced less desirable methods of manufacture. It is hoped that they will show the potentiality of market expansion through new product development. In some instances, these developments are the result of new foundry techniques to produce a better casting at a lower overall cost; in others, the casting replaces a structural member formerly made of a weldment which was inadequate in actual service because of failure.

Still another approach lies in the field of alloys especially well suited for specific application of cast alloys which cannot be wrought or are practically unmachineable. These alloys find their use in the category of corrosion or wear resistance as well as applications where high temperatures are involved.

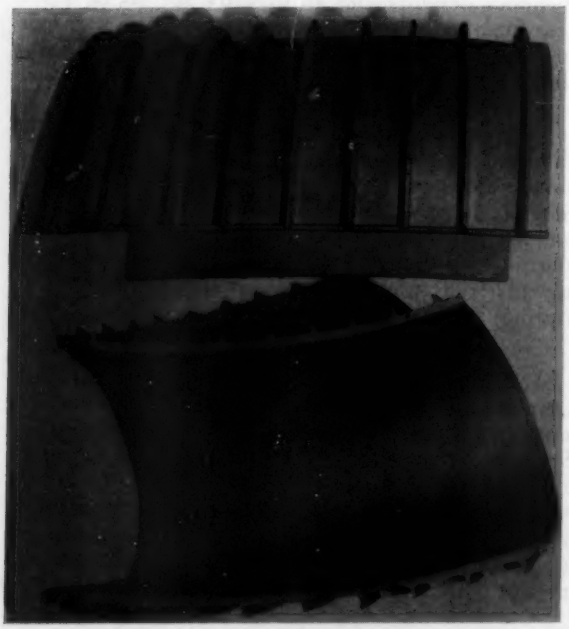


Fig. 1 — Cast magnesium alloy air duct.

Some Examples of Product Development

Magnesium

Through the cooperation and kindness of Mr. M. E. Brooks, of the Dow Chemical Co., and Mr. E. V. Lanfranke, of the Rolle Mfg. Co., Inc., it was possible to obtain some excellent examples of magnesium alloy casting developments.

A cast magnesium alloy air duct is shown in Fig. 1. This duct is 38 in. long, 20 in. across and 20 in. high. Originally these air ducts were fabricated, but due to excessive vibration in service, stresses, nuisance of fabricating and lower overall cost, it was redesigned and cast from AZ91 alloy. The foundry recommended that the duct be made of two half castings which were subsequently welded along the flange line to make the complete duct. The finished duct (two half castings) weighs only 24¾ pounds and has a wall thickness held to 0.125 in.

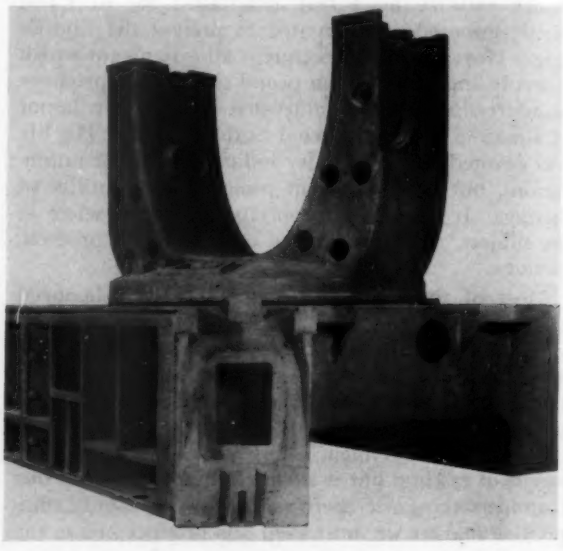


Fig. 2 — Base casting for mobile electronic unit.

A large base casting for a mobile military electronic unit is shown in Fig. 2. The design engineers originally contemplated fabricating this base, but because maximum dimensional stability was desired, it was made as a sand casting. The finished casting weighs 1500 pounds and was cast of AZ63T6 magnesium alloy. The overall dimensions are 8 ft x 8 ft x 3 ft, with a wall thickness of ⅜ in. This provides an excellent example of where a casting not only eliminated a substantial amount of welding labor, but more important, fulfilled one of the most important ultimate requisites – dimensional stability.

A comparatively small, permanent mold alloy magnesium casting is illustrated in Fig. 3. The approxi-

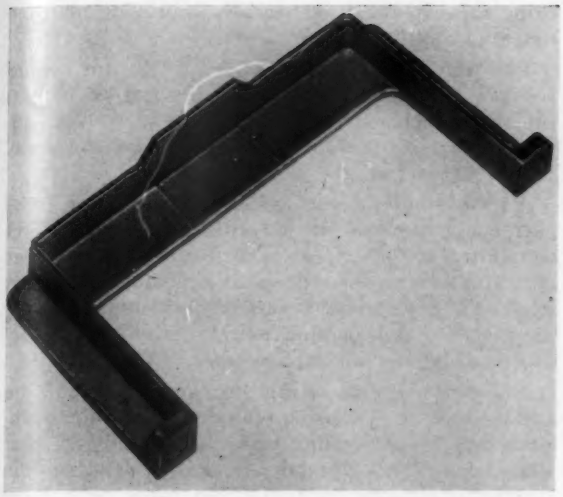


Fig. 3 — Permanent mold alloy magnesium casting.

mate overall dimensions are 2½ in. x 6 in. x ¾ in. and it weighs only 1½ ounces. This part, likewise, was originally fabricated by welding methods. Since the anticipated requirements did not justify the high initial cost of a die, it was cast in a permanent mold. The customer was well pleased wth the cast parts and in addition, the casting eliminated considerable tooling and the added labor of fabrication. The casting cost less than the fabricated part. This can be considered a fine example of new product development.

Malleable Iron

A varied and large tonnage of malleable iron is used in the manufacture of farm implements. But as in other branches of the foundry industry, new applications must be developed to compensate for losses resulting from obsolescence and the design of new machines having entirely different component parts. A fruitful field also exists in the development of castings to replace weldments which have shown excessive field failures. Following are some typical examples of a cooperative effort between the design engineers and the foundryman.

Both the welded and cast versions of a supporting member for a heavy combine are illustrated in Fig. 4. Even though the cost of the malleable casting was approximately the same as the welded fabrication, it was much stronger. Adoption of the casting eliminated field failures of this member which had formerly resulted in the fabricated design due to cracking of the welds in service. Indirect advantages of the casting are the reputation of the implement manufacturer as a producer of trouble-free equipment and the elimination of attendant service expense when structural failures occur in the field.

Similarly, the unsatisfactory field performance of a tractor front axle end, resulted in its conversion from a fabricated design to one piece malleable casting. Field failures of the component, likewise, were due to cracking in the welds. The fabricated and cast designs are shown in Fig. 5.

In addition to the advantages already enumerated, the manufacturing cost of the casting was less than that of the weldment.

Even though many applications, because of service conditions, will not permit such a conversion, Fig. 6 illustrates a pinion and bevel gear blank (as cast and after machining) which are now made of pearlitic malleable iron. These blanks formerly were forgings.

The cost of the pearlitic malleable gears, including final heat treatment, was 36 per cent less; the material cost alone was 56 per cent lower and the labor cost 25 per cent less for the casting than the forging. The flame-hardened pearlitic malleable iron gears have performed satisfactorily in service.

Pearlitic Malleable Iron

Within recent years the applications of pearlitic malleable iron have increased greatly. When the conventional grades of malleable iron do not meet strength requirements or wear resistance, it has been satisfactory, in certain instances, as a substitute for fabrications or forgings. Through the kindness of Carl F. Joseph and the cooperation of the Central Foundry Division, General Motors Corp., some excellent examples of such conversions have been made available. Special mention should be made of the fact that in each case, there was a saving in the cost



Fig. 4 — Casting (left) supplanted weldment (right) which failed in service by cracking of welds.

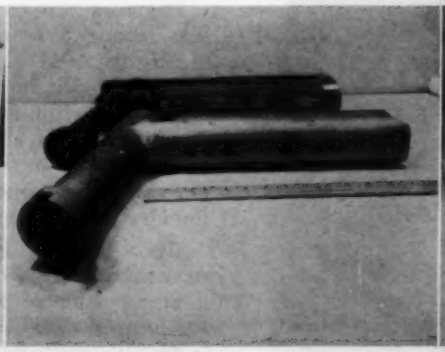


Fig. 5 — Tractor front axle as-cast (below) supplanted as-fabricated axle (top) which failed in service by cracking of welds.



Fig. 6 — Cast gear blanks and machined gears made of pearlitic malleable iron were formerly made from forgings.

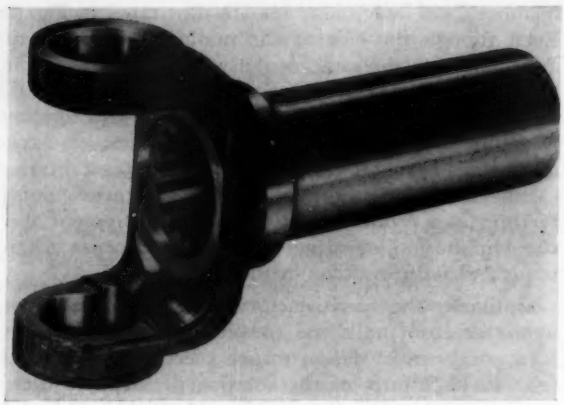


Fig. 7 — Universal joint yoke made of pearlitic malleable iron. This part was formerly made from a forging.

of the rough casting and in addition, a reduced machining cost – a dual economy to the purchaser.

A universal joint yoke made of pearlitic malleable iron is shown in Fig. 7. This part was formerly forged of an SAE-1145, a free-cutting high-sulphur steel. Substantial savings in the cost of the rough casting and machining of the part were achieved. Typically, the internal spline, which is broached, under comparative tests gave a broach life of only 9,000-10,000 pieces on the forgings; while the pearlitic malleable castings yielded a broach life of 15,000-20,000 pieces. When the part was forged, it was necessary to resharpen the broach after each run of 800-900 pieces. After the yolk had been converted to pearlitic malleable iron, 1500-1800 pieces could be machined before the broach had to be reground. The carbon spots in pearlitic malleable permit the machine chips to break off readily, improving machine time and tool life.

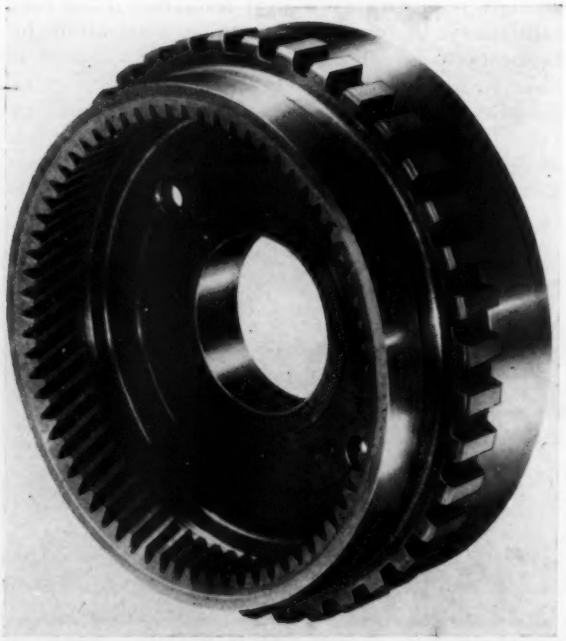


Fig. 8 — This cast reverse internal gear was formerly made from a forging.

The pearlitic malleable iron, of which this casting is made, has an ultimate tensile strength of 100,000 psi and a yield strength of 80,000 psi. When subjected to fatigue cycling tests, the casting stood up as well or better than the steel forgings.

A reverse internal gear is illustrated in Fig. 8. The gear blank formerly was forged from SAE-5140 steel containing 0.40 per cent carbon and 0.95 per cent chromium, commonly known as a medium-chromium steel. To provide the necessary draw angles on the forging die, considerable extra machining was required to remove the surplus stock.

The internal teeth of this gear must withstand the full thrust of the reversing pinion gears and to accomplish this, a surrounding brake band holds the part stationary. To withstand the intense load on the internal teeth, the part must be made of a high-strength material. It is also important that the exterior surface have a high finish to insure smooth operation of the brake band on the friction surface without chatter.

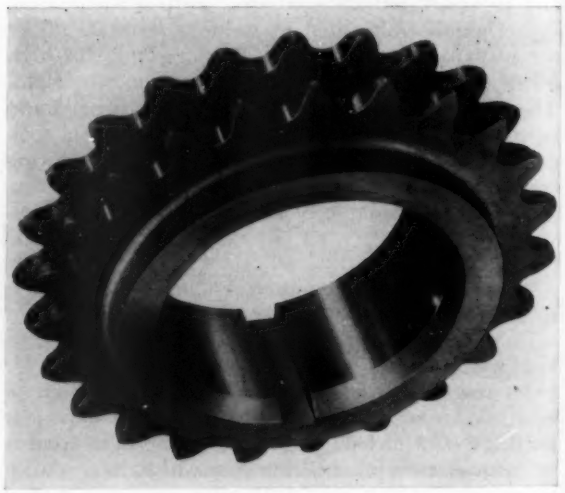


Fig. 9 — This cast crankshaft sprocket was formerly machined from steel bar stock.

The part was converted to a pearlitic malleable iron having a tensile strength of 100,000 psi and brinell hardness of 241-269. The new cast design reduced the amount of metal which had to be removed in machining; the rough casting cost was less than the forging and because of its better machineability resulted in reduced machining cost. In service, the casting performed equally satisfactorily and the foundry has another satisfied customer.

The crankshaft sprocket formerly machined from steel bar stock is shown in Fig. 9. It is apparent that a large portion of the original bar stock is wasted in the form of metal chips.

Pearlitic malleable iron has replaced the bar stock at a lower material cost, fewer machining operations and a greatly reduced total machine time. Following machining, the sprocket teeth are flame hardened by

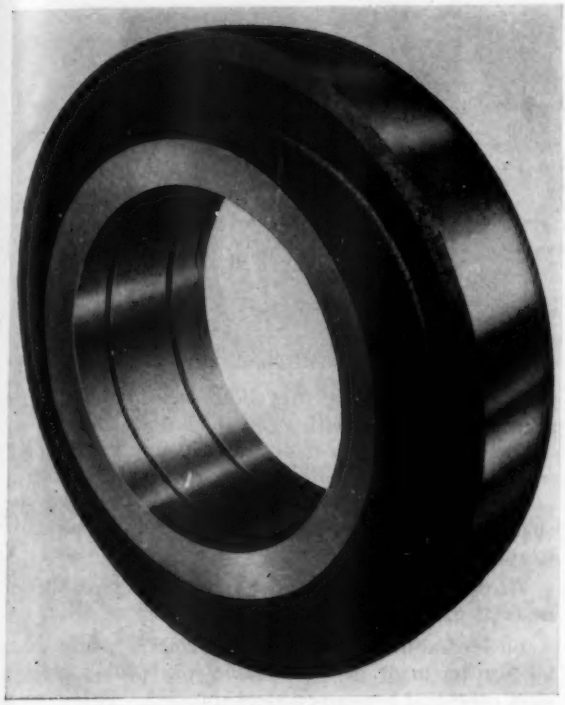


Fig. 10 — This pearlitic malleable iron cast conveyor trolley wheel was formerly made from a forging.

heating for about 8 seconds and quenching in oil held at 130-150 F. The final tooth hardness ranges from Rockwell C55-58. The finished sprocket has strength and wear resistance comparable to the steel bar stock parts.

A typical conveyor trolley wheel is pictured in Fig. 10. In this application, a pearlitic malleable casting is used to replace an SAE-1040 steel forging. The casting has a brinell hardness of 197-241 and the wearing surfaces are induction hardened to a Rockwell hardness ranging from C50-55 to a depth of $\frac{1}{16}$ in. The cast wheels cost less than the steel forgings and because of their excellent machineability, lower machining time and longer tool life result. Based upon accelerated life tests, the customer reports wear resistance comparable to or better than forged steel wheels of like hardness.

The common "C" clamps shown in Fig. 11 were originally steel forgings made from SAE-1040 steel.

Low tool life, high machining and forging cost, plus excessive metal resulted in converting these clamps to pearlitic malleable castings. The conversion resulted in substantial savings but the strength of the forgings was retained.

High Alloy Steel

The application of casting techniques is well portrayed in the field of highly alloyed steel castings which must withstand corrosive media or exhibit relatively high strength at elevated temperatures. Especially noteworthy in the following applications is the wide latitude afforded the design engineer. These parts are so complicated as to be almost impossible of economical manufacture by other methods.

A complicated austenitic 310 stainless steel casting is illustrated in Fig. 12. This part was cast in an investment mold. The complex design of the casting practically rules out economical production by other than foundry techniques and due to being investment cast requires minimum machining.

Another intricate jet engine fitting is shown in Fig. 13. This part was also investment cast of a type 302 stainless steel.

Again, the part was cast because of its intricacy. Note the cast-in holes, recessed areas and varied shape, yet functional design has been accomplished with minimum weight. Machining required only facing and tapping of the threaded holes.

Because of the relatively small quantities required and close dimensional tolerances demanded, the casting in Fig. 14 was investment cast. This gas turbine oil fitting was also cast of a type 302 stainless steel.

Although this casting appears simple, its shape is relatively intricate since all holes were cored and the angled bevel was cast so that the only machining required was tapping the threads.

Nodular Iron and Shell Molding

The development of nodular iron combined with shell molding has opened a new field for the foundry industry. The inherent mechanical properties of nodular iron made it suitable for the manufacture of structural members where relatively high strength combined with ductility are important. Close dimensional tolerances together with excellent casting finish, characteristic of the shell molding process, have materially lowered machining costs — a strong selling argument for the casting salesman.

As a result of the fine cooperation and assistance

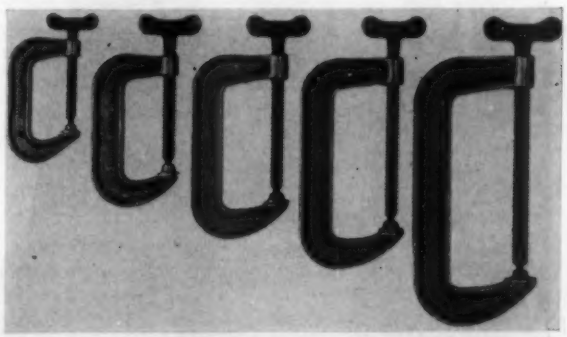


Fig. 11 — These cast pearlitic malleable iron "C" clamps were formerly made from forgings.

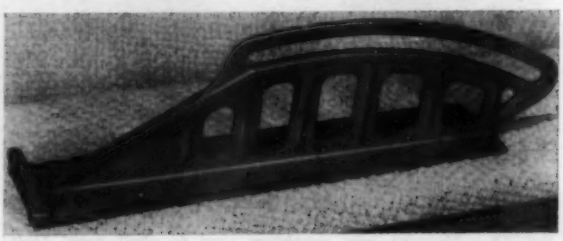


Fig. 12 — Austenitic 310 stainless steel castings made by the investment process.





Fig. 13 (Left) — Intricate jet engine fitting and, Fig. 14 (Right) — Gas turbine oil fitting both cast of a type 302 stainless steel by the investment process.

of Mr. T. W. Curry, Lynchburg Foundry Co., some excellent examples in this field have been made available.

A typical crankarm is illustrated in Fig. 15, originally fabricated as a weldment and then redesigned for shell molding followed by casting of nodular iron. It is interesting to note that the arm is poured in a vertical position and that both bearings must be risered to obtain solidity in these locations. Two castings are made in a mold and the gate and runner enter the risers. In addition to economy of manufacture, field experience shows that the bearing characteristics of the nodular iron casting are superior to that of the weldment.

A combination pinion gear and sprocket are shown in Fig. 16.

This combination gear was formerly made as a fabrication made of a malleable iron pinion gear, a steel tube and a stamped steel sprocket. This part is now cast of nodular iron in a shell mold as an integral unit. The space between the pinion and sprocket is formed by a conventional foundry core with the balance of the casting in a shell mold. The gear teeth require no machining and are hard and ductile

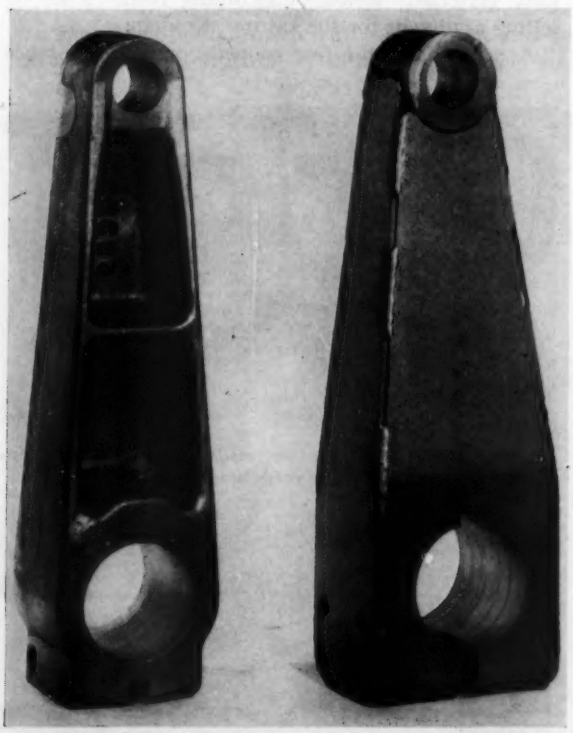


Fig. 15 — Nodular iron crankarm cast in shell mold (left) supplants crankarm fabricated as a weldment (right).

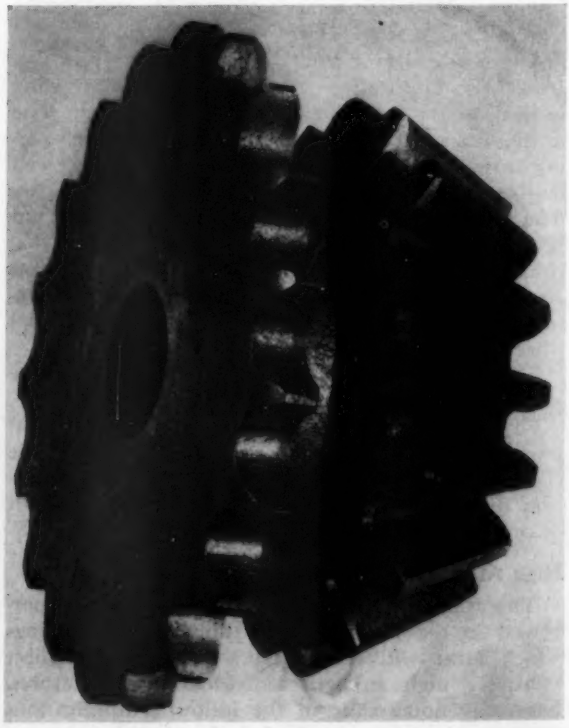


Fig. 16 — Nodular iron combination pinion gear and sprocket cast in shell mold supplants a fabrication from three individual parts.

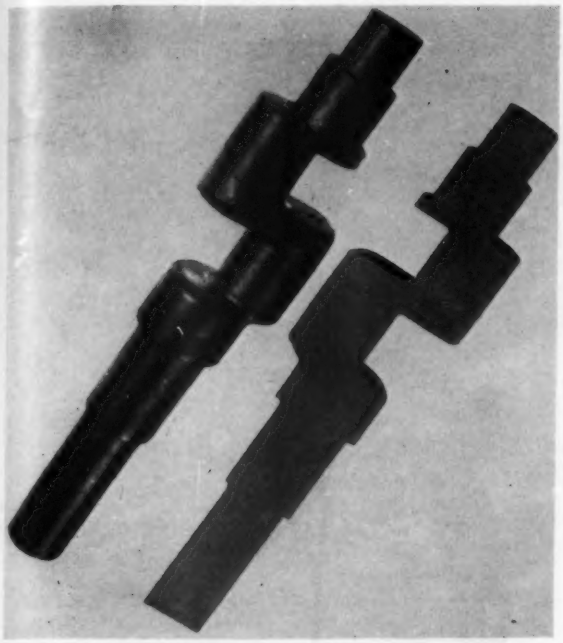


Fig. 17 — This crankshaft is a casting, formerly a forging.

enough in the as-cast condition to be ideal for the application involved. Little thought is required to appreciate the economic advantages involved in this conversion.

A typical crankshaft is illustrated in Fig. 17.

Almost unlimited possibilities are available for the conversion of forged crankshafts to nodular iron, in many instances cast in a shell mold to minimize machining cost. The cast crankshaft offers broader latitude to the design engineer and equally important, at a lower final cost. At least in the smaller sizes, the nodular iron crankshaft can be used either in the as-cast state or with some form of subsequent heat treatment. The full potentiality of the foundry industry in this field is emphasized by the announcement of the General Motors Corp. that the Pontiac crankshaft is now to be cast in shell molds of pearlitic malleable iron.

All three of these applications were cast of nodular iron having the following properties:

Ultimate Tensile Strength,	90,000-115,000 psi
Yield Strength	65,000- 85,000 psi
Elongation	2½% - 5%
Brinell Hardness	200-269

This iron is melted in an acid-lined cupola, desulphurized in a forehearth by injecting calcium carbide and then treating the iron with a combination of Mg-Ce-Si and Ni-Mg-Si alloys followed by a late ferro-silicon inoculation before pouring.

Gray Iron

Our most versatile cast metal is gray iron and likewise, represents the largest tonnage. A large variety of cast irons having a broad range in mechanical properties afford the design engineer wide latitude in structural design combined with economy of manufacture, service performance and ease in machineability. The excellent vibration dampening capacity of gray iron, for example, in the case of engine and machine tool components results in quiet operation and excellent finish on the machined parts since harmonic vibrations are at a minimum when the machine is in operation.

The outstanding casting characteristics of gray iron make possible contours and complexity of design impractical by other methods of manufacture and often at substantial savings in cost. Hence, even in our modern times cast iron still continues as one of our most important structural materials.

The Gray Iron Founders' Society sponsors a contest which encourages gray iron foundrymen to develop ingenious designs for castings to replace other methods of manufacture resulting in a saving in production cost and better service performance. Through the cooperation of the society, several excellent examples of such gray iron applications have been made available.

A machine tool spindle housing is shown in Fig. 18, originally fabricated as a weldment from a number of steel components. The same housing is shown after it was redesigned and cast of gray iron. The manu-

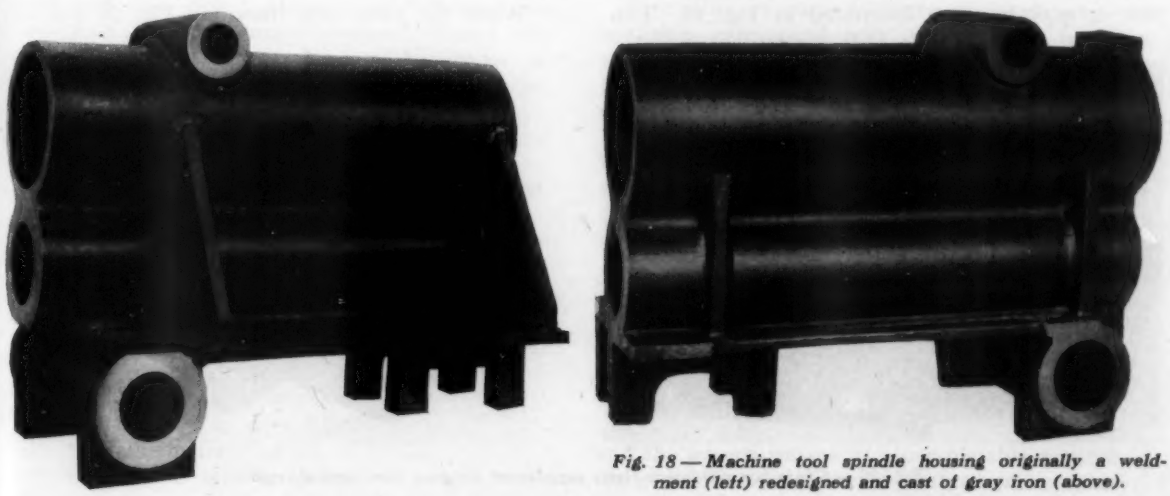


Fig. 18 — Machine tool spindle housing originally a weldment (left) redesigned and cast of gray iron (above).

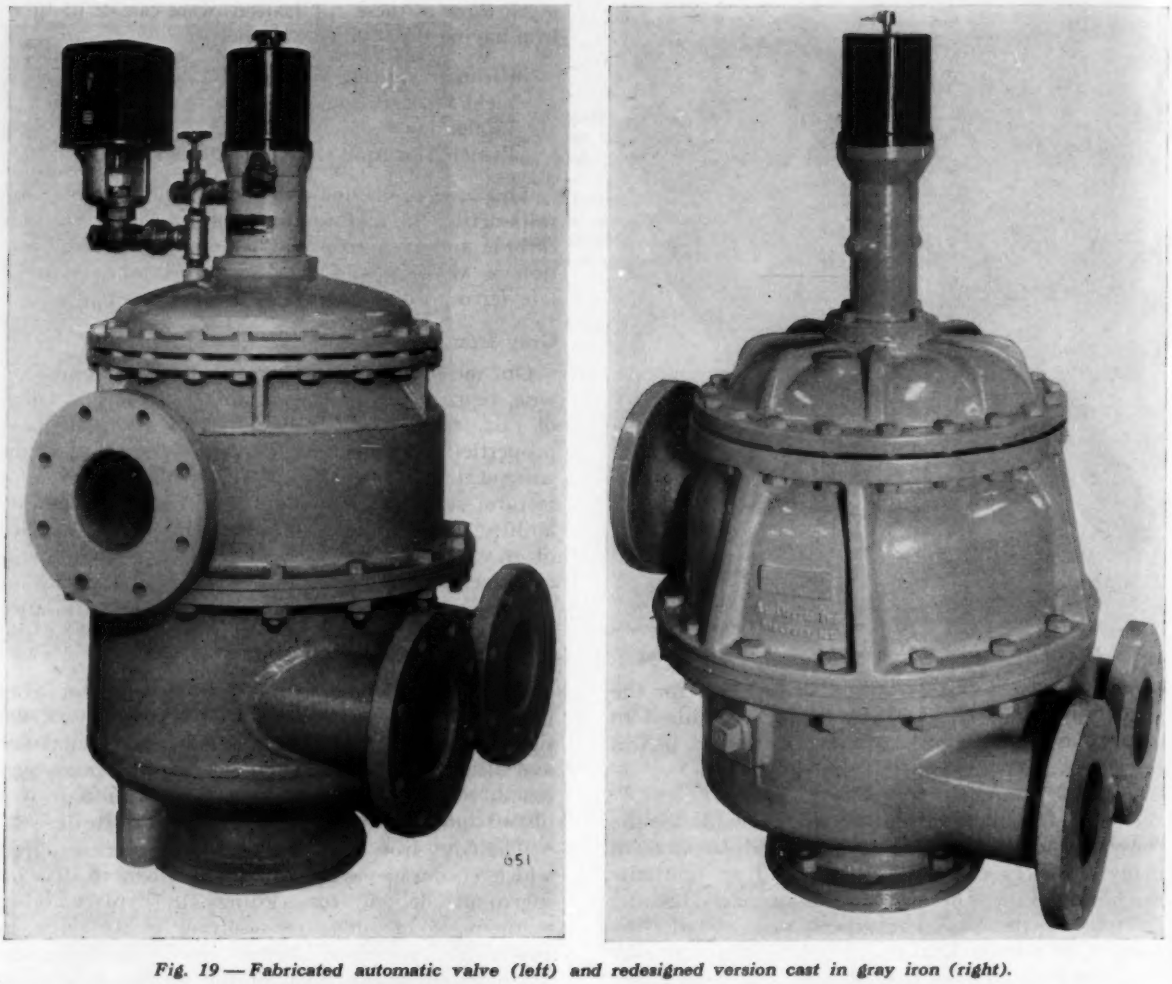


Fig. 19 — Fabricated automatic valve (left) and redesigned version cast in gray iron (right).

facturing cost of the weldment was $218.57 while the gray iron casting cost only $109.67 — a saving of $108.90 per unit, or a reduction of slightly less than 50 per cent. In addition, no one will deny that the casting has considerably improved eye appeal.

A fabricated automatic valve and a redesigned version in gray iron are illustrated in Fig. 19. This conversion not only resulted in reducing the production cost from $338.42 to $91.71 — a saving of approximately 70 per cent — but also eliminated an annoying service problem caused by leakage in the fabricated assembly. Why shouldn't this have created another satisfied customer for the foundry industry?

When the chain box shown in Fig. 20 was con-

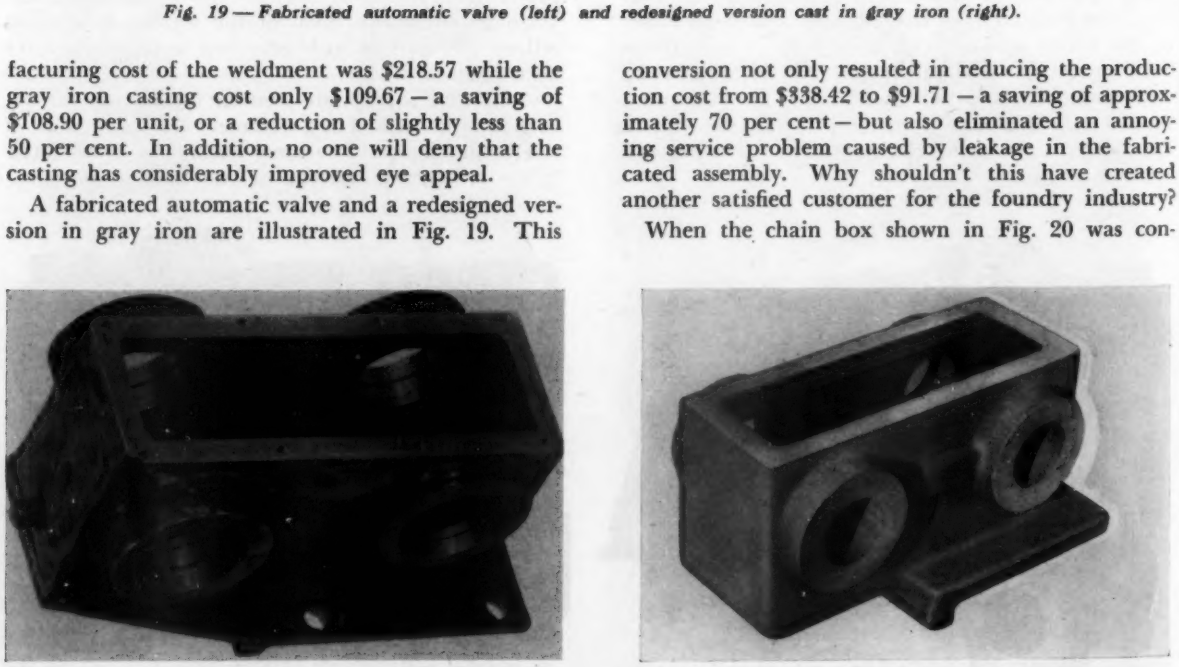


Fig. 20 — Chain box weldment (left) supplanted by gray iron casting (right).

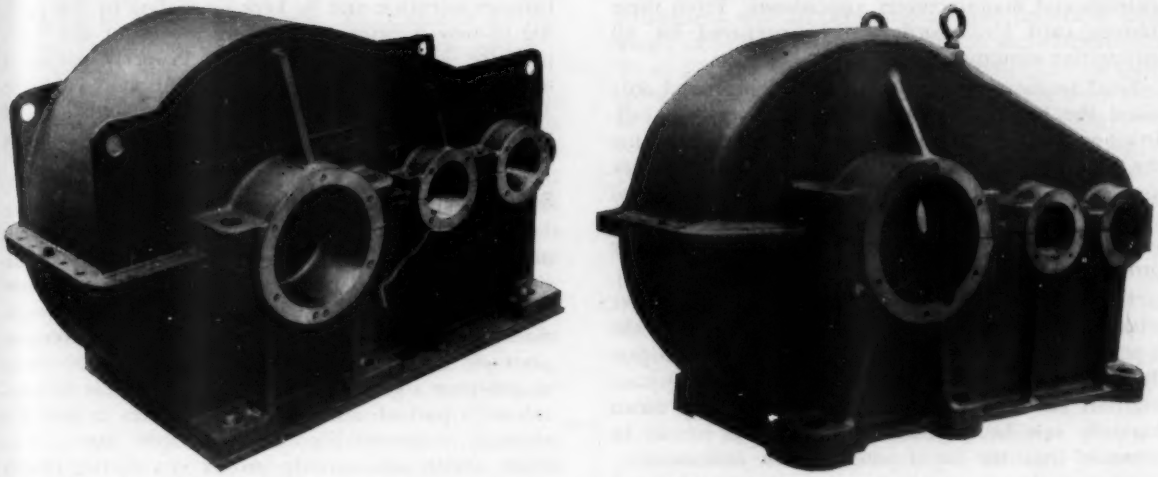


Fig. 21 — Speed reducer housing (above) fabricated by welding together 69 parts and (right) redesigned version made from only two gray iron castings.

verted from a weldment to a gray iron casting the initial cost was lowered from $21.00 to $9.22 and the machining cost reduced from $23.50 to $12.50 – or a total reduction in cost of the finished part from $44.50 to $21.72. Thus the finished casting was produced at an overall saving of about 51 per cent. What casting purchaser wouldn't be agreeable to such a saving?

The speed reducer housing illustrated in Fig. 21 was fabricated of 69 parts; after modification in design it was cast of only two gray iron parts. This conversion resulted in a 73 per cent saving in cost of unmachined pieces and 20 per cent saving in machining cost. The performance of the gray iron casting is superior because of its ability to absorb vibration and operate more quietly. The savings on the first three units covered the difference between pattern and templet costs. Again, in appearance, there can be no question as to which looks better.

These are only a few typical examples of new applications for castings and many more are available. Each attempt requires a careful analysis of the customer needs, service requirements, the economics involved and the closest cooperation between the foundryman and the design engineer. Sound engineering combined with a thorough knowledge of castings and their advantageous application are required.

Elements of Marketing Castings

The several phases of marketing castings are briefed in the following:

A. Planning
   1. Market analysis.
   2. New plant facilities and processes.
   3. Complete customer and prospect records.

B. Sales Effort
   1. Salesman training.
   2. Scheduling of sales calls.
   3. Manufacturers' agents.
   4. Complete sales records.

C. Advertising
   1. Budget.
   2. Trade publications and newspapers.
   3. Engineering letters – direct mail.
   4. Complete brochures or catalogs.
   5. Open house.
   6. Exhibits at industrial shows.

D. New Product Development
   1. Cooperative engineering with customer.
      (a) Improved performance.
      (b) Weight reduction.
      (c) Lower machining or assembly cost.
      (d) Improved metallurgy.
      (e) Lower initial cost.
   2. New applications for castings in newly developed products.

Planning

Market analyses are, in many respects, not nearly as difficult as most foundrymen presume. A considerable number of reliable sources of information are readily available at a comparatively low cost. The actual job is in most respects a clerical one supplemented by a knowledge of equipment on which castings are used. The latter, many times, is apparent in a picture in an advertisement.

Directories of manufacturers of many areas are available, categorized into three general headings: alphabetical order, product and geographical location. These directories frequently include the names of key personnel and total number of employees. The information obtained in this manner is helpful in preparing a list of prospects and a direct mailing list. Although they lack some desired information, classified telephone directories are also an added source.

Excellent additional sources are publications such as:

Mac Rae's Blue Book,
Poor's Register of Directors and Executives,
Thomas Register of American Manufacturers,
Reference Book of Dun & Bradstreet, Inc.,

and the advertisements in a considerable number of trade publications, as well as those of engineering

societies and manufacturers associations. From these sources, card file records can be prepared for all prospective customers.

Final verification can usually be accomplished only when the foundry salesman makes a personal call. During such a visit, the salesman should determine the kind of castings, approximate volume, if present sources are considered satisfactory, general quality of castings they are presently buying, an estimate of the size and character of the company and its ultimate product.

The salesman should acquaint the casting buyer with the manufacturing facilities of the foundry, the type of castings it is unusually well qualified to produce, and get an opportunity to quote on some castings. If the company does not buy castings which logically can be produced economically, it can be removed from the list of potential new customers.

The salesman should be held responsible for a complete report on every call so the sales manager can make final decisions on all matters of policy. If the prospect looks promising, subsequent calls should be made to emphasize the desire of the foundry to serve the customer. Frequently several calls must be made before the first sale is made. The reasons are quite evident because of the peculiar nature of the foundry business.

It is presumptive to think that a new prospective customer will call in a pattern already located in another foundry and place an order with a new foundry unless the presently procured castings are unsatisfactory in quality and delivery or there is a marked price advantage. Even when conditions favor a change in casting source, the buyer knows from past experience that a new source will require time to make sample castings and adjust to the manufacturing technique before castings are delivered which are wholly satisfactory.

These steps cost time and money, even if we ignore the fact that the new foundry must by actual manufacture prove that it is capable of making acceptable castings at a lower price and adhere to required delivery schedules. The sale of castings differs from selling an item already manufactured and in warehouse stock ready for shipment, or a standard product continuously in production involving appropriate production schedules to meet the customer's needs.

In the manner described, the sales organization should call on each of the prospective customers included in the lists prepared from the sources enumerated. Utilizing the salesman's visitation report you can eliminate from your list those firms which either do not use castings or are of a kind, size and quantity unsuited for economical manufacture in your foundry. For reasons of economy, it is quite evident that the original names should be grouped geographically to minimize time and travel expense.

Eventually, the foundry will, by an elimination process, develop a list of prospects which comprises firms whose casting purchases justify intensive sales effort. The list also provides names for a direct mail advertising campaign, the details of which will be described later. The prospect list must have continuous attention and be kept up-to-date by the posting of new information relating to any of the firms plus the addition of new ones. Properly prepared and maintained, it becomes one of the most valuable assets of the sales department. It represents one of the first steps in a systematic sales effort to explore new markets.

Sales Program

The word "salesmanship" is comparatively new in our vocabulary. It is not even found in the 1900 edition of Webster's Unabridged Dictionary. In a few words, we may define "salesmanship" as any means, direct or indirect, by which it is sought to induce or persuade a given individual or group of individuals to purchase a given article. Advertising may be considered a part of salesmanship but differs in that the element of personality does not enter into it. In other words, salesmanship proper is a matter chiefly of the personal relationship involved in selling.

The latter is stressed because selling of castings is a personalized influence rather than through the creation of a market as a result of large scale advertising and a big distributing organization. Each casting requires individual attention if it is best to serve the customer.

The foundry salesman is the medium by which casting requirements are transmitted to the foundry operating personnel. Only through such intimate communication can the foundry produce the casting so that it will meet the customer's needs at the lowest cost. This emphasizes the necessity for adequate training of the foundry salesman, the primary responsibility of the sales manager, supplemented by the technical staff of the foundry.

Selling castings is an engineering job. In many instances, the salesman should be accompanied by a member of the foundry technical staff when working out the engineering details with the customer's design engineers. Matters such as the most desirable pattern equipment, design of the casting, metal sections, dimensional tolerances distortion, critical areas, finish allowances and the required physical and mechanical properties of the metal are essential to the successful manufacture of a casting.

For reasons which do not require repetition, the primary contact with any customer is the purchasing agent or castings buyer; but it is equally important to have intimate contact with the design engineer. He is the man who knows the ultimate needs of the casting and the one who can cooperate in making the design a practical one from the casting standpoint. The design engineer, likewise, is the man who is interested in the engineering properties of the metal so that its qualities may be used to best advantage. Engineers welcome information, providing it is factual and presented by a trained engineering salesman.

Sales calls should be systematically planned, dependent upon the character and volume of business available from any given customer. Up-to-date records should be maintained showing date of the last call, extent of sales, service calls – including reason and results – date when a customer stopped buying

and why, as well as new prospects together with estimate of potential business. Calls should be frequent enough so that the customer feels you are interested in his business.

At intervals, the foundry president should call on the more important customers. A close relationship once established between the customers' executive officers and the foundry president serves to cement their mutual relations in a manner unattainable in any other way. Be on the job to service his needs better than your competitor and he will favor you with his business.

A survey was made of what happened to 100 customers of a firm over a 10-year period. The results were quite startling. Of a total of 100, 15 customers quit during the first year; 13 during the second year; 11 during the third year – and so on until at the end of the 10-year period only 19 active customers remained. Of the former 100 customers who could be located, 68 per cent had withdrawn patronage because of discourtesies, poor service, or indifference; another 14 per cent had gone elsewhere because of unadjusted grievances. Price cutting or buying for less had lost only 9 per cent. In other words, 82 per cent of the original customers had been lost purely through neglect.

Old customers are more valuable than new ones. They are familiar with you and have confidence in your company's products, services and personnel. It pays to keep them. Retention of old customers requires only half the effort necessary to acquire new ones.

There are instances where a small foundry cannot support a large sales force or a large foundry may want to supplement its sales effort by employing the services of a manufacturers' representative. He is an independent operator selling on a commission basis, frequently profitably used in areas where the total potential volume of business is not sufficient to justify employment of a full-time salesman. No financial investment is involved because the manufacturers' representative is compensated only after the sale has been made and the castings shipped to the customer.

Since such an agent will represent the foundry in the eyes of the customer it is important that he be selected with the same care used in hiring a new salesman. It is also essential that he is well informed on the manufacturing facilities of the foundry, the class of castings it is best qualified to produce and its overall policies. The foundry must be prepared to support him with an adequate promotional and advertising program as well as technical assistance and prompt preparation of quotations when requested.

Advertising

Advertising is a form of mass communication which keeps the name of the foundry before the eyes of its existing and potential customers. It also informs the public of the types of castings the foundry produces, the merits of its product and the uses for the castings. Advertising helps reduce the cost of selling by salesmen.

According to a survey conducted by the Sales Executive Club of New York by contacting 228 representative manufacturers of industrial materials, equipment and services, the average cost of a salesman's call on a prospect was $17.24 and ranged from $5 to $30 per call. This clearly indicates how judicious advertising is a means for lowering sales costs by providing inquiries which result in a higher percentage of profitable sales calls, to say nothing of potential customers which might be totally overlooked due to a lack of knowledge of their needs. Conversely, frequently it is the practice of a purchasing agent to file advertisements for future reference as possible sources for specific items.

An annual advertising budget should be established both in total amount to be spent for this purpose as well as a breakdown of the manner in which it is to be used. Several types of advertising may be considered appropriate and profitable in the foundry industry such as trade or professional societies, publications, newspapers, direct mailing, complete brochures, catalogs, open house at the plant and casting exhibits at industrial shows or conventions.

Each serves a definite purpose, the value of which must be determined by foundry management in terms of the market they wish to stimulate and what is to be accomplished.

With some few exceptions, foundries have seldom advertised in daily newspapers. Exceptions are occasionally observed in which a foundry is publicizing open capacity for the production of a specific class of casting but such advertising is spotty and does not constitute a consistent effort in general practice. The author has been unable to learn if this kind of advertising has proven profitable to the industry and hence would hardly be considered an important form of publicity for the foundry. Its value is restricted and fails to inform the purchaser of the skill and manufacturing facilities of the foundry.

Trade publications by comparison provide a valuable form of advertising for the industry, especially if the primary theme concentrates on subjects which will appeal to the design engineer, such as new adaptations, ingenious designs combined with broad flexibility and typical applications by prominent manufacturers. In brief, such advertisements should stimulate the interest of the design engineer in the many advantages of casting over other methods of manufacture, either because of lower original cost, machining or assembly expense.

The desirable physical characteristics of the casting in terms of ultimate service performance should also be emphasized. An abundance of factual engineering information is available emphasizing the multiple advantages of castings but the industry, for the most part, has neglected to use it as one of its most effective selling arguments. The casting process has inherent characteristics which cannot be equaled, in many instances, by other methods of manufacture. We must capitalize on these advantages by making our customers fully aware of them.

Closely associated with this segment of available advertising, the magazines published by the various professional groups, such as those of the civil, mechanical, automotive, electrical, chemical, etc., societies should be included. These publications are

produced by and for engineers – the men primarily responsible for selecting the method of manufacture and therefore must be made aware of the attributes of castings.

Well prepared technical papers presented by foundry technologists before the same engineering groups constitute a valuable form of advertising which is direct and enduring. Such papers, when reproduced in the publications of the engineering societies, result in a highly selective national coverage. Since these same publications are available to engineering students in our colleges, information on the application and merits of castings is simultaneously educating the future design engineers, our customers of tomorrow.

The Foundry Educational Foundation is serving the foundry industry effectively in this direction by not only providing foundries with well schooled technologists but also with what may be considered a valuable advertising campaign which cannot help but pay dividends in the future.

Another form of advertising directed to design engineers, purchasing agents and castings buyers is found in industrial catalogs. Their usefulness is increased because of their great length of life when compared to other forms of advertising. Frequently, they serve as the buyers' first source of information for comparing competing products and deciding which companies should be asked to quote on their requirements.

As an excellent example, the *Mechanical Catalog* published by the American Society of Mechanical Engineers, is typical of several in this group. This book, published annually, is available to some 20,000 members of the society. It is divided into three sections: display advertisements, a directory of manufacturers classified by products, and a directory of trade names.

Direct Mail Type of Advertising

Having prepared a carefully screened list of prospective customers, the direct mail type of advertising in the form of concise, factual engineering letters have proven highly productive based upon actual customer response and interest. A letter of this kind sent once a month is an effective means of keeping the foundry's name before the customer and at the same time furnishing useful information to his engineering staff. Since it contains data of value, it is more apt to be filed away for future reference. The same mailing may include a reprint of a recently published advertisement which portrays the application of a casting by some prominent manufacturer already an established customer of the foundry. A highly respected clientele is one of the best recommendations any company can possess. Direct mail advertising should also be sent to present customers – particularly if it consists of engineering information of value to any casting user.

Numerous progressive foundries in the past have prepared complete brochures covering their manufacturing facilities, extensive illustrations portraying a variety of castings they have made or in current production and specifications of the cast metals they produce. It is the author's opinion that this form of advertising constitutes one of the most profitable for the castings industry.

Appropriate is the old Chinese proverb "one picture is worth more than 10,000 words." A brochure carefully prepared is almost indispensable to the foundry salesman. Good illustrations convey a story almost impossible to tell in words and leave a permanent reference with the customer's purchasing agent and engineering staff. While the original cost is high and requires considerable work and thought, its long term value fully justifies the expense and effort. For a direct mail campaign, it has no substitute.

"Open House" As a Form of Advertising

It has often been said "seeing is believing" and it certainly applies to the foundryman's customer. The casting buyer or design engineer seldom refuses an invitation to a foundry "Open House." He is interested in your facilities and how his castings are made. A large percentage of progressive purchasing agents insist on seeing your industry before they are willing to give you an order. Can you blame him?

In many respects, the foundry reflects the kind and character of the castings he may expect from your shop. Personally seeing the quality and type of castings ready to be shipped provides the casting buyer with excellent evidence of your ability as a foundryman. But, when you do have an "open house" be sure you put your best foot forward. A thorough clean-up and some paint provide the eye appeal many foundrymen have overlooked in the past.

Casting exhibits provide an excellent medium by which the casting producer can make his product known. It is a type of personalized advertising. Until now, a few foundries have exhibited at industrial shows such as the National Metal Congress but these efforts have been sparse. Is there any better way to personally contact a large number of potential customers and especially their design engineers?

An exhibit of typical castings and large pictures of the foundry and its facilities are at least one form of concrete evidence to convince your potential customer that you are qualified to serve him. It goes without saying that such an exhibit must be manned by foundry personnel qualified to meet with purchasing agents and engineers to discuss their needs and to convince them of the foundry's talent, process control and facilities available to serve them.

Recently it has been suggested that the exhibits at the Annual Casting Congress and Show, sponsored by the American Foundrymen's Society, be expanded to include foundry castings exhibits in addition to foundry equipment and supplies. By expanding the technical program to include sessions which will appeal to casting users as well as the foundryman, the Congress can profitably serve a dual effort and thereby advance the status of the foundryman as well as his customer – a wonderful opportunity for closer liaison so essential to the successful manufacture of castings. When plans have been completed by the American Foundrymen's Society for such an integrated Casting Congress and Show, foundry management should prepare to support it.

Conclusions

Every foundry should have some form of marketing program. The basic principles have been outlined and the extent of such a program can be tailored to suit the needs of every foundry regardless of size.

Most important of all is to know your product; explore and energetically sell it based upon its merits; understand its deficiencies and don't mis-apply them, for to do so hurts the future of your business as well as all other foundries. The need for sound castings having high metallurgical quality is evident and its importance will be even greater in the future. Poor design, low strength material not meeting the customer's requirements and especially defects in highly stressed areas are the surest way of losing business.

Management must take an active interest in a sound marketing program and recognize the need for personal contact with existing and potential customers.

The industry needs engineering talent to control established foundry materials and processes as well as develop new improved methods and castings to meet increasing competition.

THE CARBON-DIOXIDE PROCESS

By

D. V. Atterton*

Introduction

In this paper, a survey is presented of the technical and practical developments of the carbon-dioxide process in Great Britain. When descriptions of a process for hardening molds and cores with carbon dioxide were published in various German and Eastern European journals, the majority of British foundrymen were, to say the least, very skeptical. After an extensive laboratory investigation, we carried out the first major foundry trials in Great Britain with the process at Glanmor Foundry, South Wales, in February 1954 in conjunction with Mr. Ivor Rees and his staff.

Mr. Rees, along with a few other foundrymen such as Mr. Andy Talbot of Western Foundries, London, soon realized the great potentialities of the process and made such rapid strides with its development that, within a matter of a few months, the majority of their molds and cores were being hardened by carbon dioxide. From then onward, the process has been rapidly adopted by many foundries for castings ranging from light alloy to alloy steel and from a few pounds to more than 20 tons in weight. In fact, it is evident that developments with the process have surpassed even the most ardent enthusiast's expectations.

The principle of the process, as used today, was described in a British Patent of 1898 taken out in the names of two Lancashire chemists, Hargreaves and Poulson; however, probably because all foundry sands used at that time for both molds and cores were naturally bonded, the idea of a synthetically-bonded sand involved such a radical change of outlook that apparently no foundry trials were even carried out.

During the last five years or so, numerous articles have been published describing practical foundrymen's experience with the process; the early papers came from the Soviet bloc of countries and suggest that the process has been used for probably the last decade. The more important of these papers[1,2,3,4,5] have been summarized and reviewed elsewhere[6] and many of the theoretical aspects and practical techniques have been superseded by later developments in this country.

Recent articles[7,8,9] and the discussions at various branch meetings of the Institute of British Foundrymen have made it clear that the process is already accepted as a production technique and that interest is now focused more on extending its sphere of application.

There can be no doubt that even greater benefits of the process have yet to be realized and much further progress will be made with developments of new binders and gassing techniques. However, it must be emphasized that the process is in no way a remedy for bad foundry practice nor is it adaptable to universal application. The author's main recommendation to all foundrymen is to try the process, after which there is little doubt that they will at once appreciate its great potentialities.

Advantages and Limitations

The actual advantages of the carbon-dioxide process to a particular foundry depend to a large extent on many variables such as type of metal being cast, size of casting being produced, limitations on output, and on various local factors such as availability of skilled labor. The following list of advantages is not intended to be comprehensive and is based primarily on information supplied by a wide range of foundrymen using the process; it is considered that most of the benefits are generally applicable to foundries casting metals ranging from light alloys to steel.

1. There is no necessity to stove the mold or core if a suitable composition binder is employed. The elimination of this time delay means that much better delivery can be given to meet urgent requirements of customers, and it is not uncommon for the pattern for quite moderate size castings to be received in the foundry in the morning and for the casting to be dispatched on the same day. Also, the problem of synchronizing the production of core shop and casting bay are considerably simplified since cores that used to take many hours and often days to produce, are now ready in a few hours or even minutes for casting. For example, in one iron foundry, a sizeable

*Foundry Services, Ltd., Birmingham, England. Official Exchange Paper from the Institute of British Foundrymen to the American Foundrymen's Society.

core weighing 50 lb was damaged just prior to assembling the mold for immediate casting; within two minutes a new core was blown on a core blower and hardened by CO_2 gas and, within five minutes, it was assembled in the mold ready for casting.

Other advantages resulting from the absence of stoving are the elimination of binder and core stove fumes, of doubts as to whether cores are under or overbaked and of bottlenecks in production due to limited stove capacity.

2. There is no necessity for core carriers when the core is hardened by carbon dioxide in the box. In addition, the use of strengtheners and irons in molds and cores can be eliminated when these are normally employed to impart rigidity in the green state. Thus, as an example, one steel foundry is now regularly producing a large core measuring 16 ft x 2½ ft x 3 ft without any core irons; previously, it was made with six irons and had to be turned out on to a machined plate for stoving.

However, when strengtheners are added to give support in the dry state, such as in the neck of a barrel core, these should not be removed or reduced.

The process also results in greater dimensional accuracy since the distortion that can occur with conventionally bonded sands in the green state or during stoving is eliminated. Apart from this greater accuracy, there is an additional benefit in that cores no longer have to be rubbed down to fit into prints.

3. Cores and molds made by the process can be stored almost indefinitely as long as due care is taken with the sand mixture and with the method of gassing, as outlined later in the paper.

4. A unit sand may be used throughout the foundry for both mold and core production which, of course, is not normally possible. A large iron foundry producing a wide range of castings now uses a unit mixture of 3½ per cent binder, 1 per cent pelletted pitch in a sand of AFS Fineness number 75 for both cores and molds with the addition of 1 per cent of a proprietary clay binder to the mix on those occasions when green strength is required. Talbot[7] states that in his foundry the same mix is used for heavy or light work in iron, gunmetal and brass.

5. Transport costs are markedly reduced with the process primarily because there is no longer any need for stoving and also because cores can often be prepared adjacent to the mold. It is fairly general in foundries producing heavy cores that about six crane lifts are saved per core. Instead of seven or eight lifts for such operations as grids into core box—plate on to rammed core—turn over—on and off bogey for transport into core stove—in and out of stove—location in mold—only one lift is now required for lifting the core and placing directly into the mold. Thus, one foundry producing daily twenty 500-lb castings has found a saving of 140 crane lifts and consequently less time wasted in waiting for cranes.

6. There is a considerable increase in casting output from available floor space. Instead of a mold occupying floor space for days while cores are being prepared and stoved, the fully cored-up mold can be completed in a much shorter space of time and the job cast. For smaller castings, book-type molds are readily prepared and no molding boxes are required; the technique of stacking or clamping book molds between fixed divisions on the foundry floor is already being practised.

7. There is considerably less danger of blowing with CO_2 molds and cores except in the case of light alloys. This is due to appreciably higher permeability and the much lower volatile content of CO_2 sands compared to oil or resin sand mixtures. The lower volatile content also results in a marked reduction of obnoxious fumes in the foundry during casting with consequent better working conditions.

Thus, one foundry doing a large amount of core assembly work had to evacuate the foundry after casting when using oil-bonded sands and, although this was reduced to a certain extent by using resins, the management of the foundry were forced to consider the installation of an extensive fume-extractor system. By adopting the carbon-dioxide process, fumes were reduced to such an extent that the need for fume extractors no longer existed.

8. On steel castings, in particular, it does seem that there is a reduced incidence of both hot tears and scabs. The main reason for this is probably the high plasticity of the surface layers of sand at elevated temperatures so that the sand yields under expansive and compressive stresses and thus prevents the formation of scabs and hot tears respectively. There is also moderate evidence to show that contraction is much less restricted with CO_2 sands and that castings are consequently much more free of internal stresses; if this work is substantiated, patternmakers' allowances for steel castings might have to be modified.

9. After a certain amount of experience has been gained, cores and molds can be prepared by semi-skilled labor since there is no longer any need, as with normal sand mixtures, to sprig the sand carefully, vary the degree of packing etc. It is, in many ways, fair to say that the process has enabled much of the humbug to be taken out of coremaking.

Also, cores previously made in two halves and jointed after drying can now frequently be made in one piece. For example, an S-shaped core for an aluminum casting, measuring about 30 inches long and 6 inches diameter, was previously made in two halves which were jointed after stoving; the total production time, including ramming, shaping and inserting strengtheners, strickling the two halves, jointing and coating the cores etc, but excluding stoving time, was between 20 and 25 minutes, whereas the total time is now less than 10 minutes to produce a core ready for casting.

10. One of the most attractive features of the process is the small capital outlay involved for carrying out experimental trials and introducing the process into a foundry. Also, the cost of sand mixes is often substantially lower than oil or resin sand, and the other advantages of the process make it possible to effect considerable savings in the production costs of castings.

When the process was first introduced into Great

Britain, various obstacles had to be overcome before it became a sound production method. None of these were insuperable and it would now appear that there are no major limitations of the process. However, it is perhaps of interest to mention these initial obstacles and indicate how they were overcome.

a) Sands mixed with the silicate-base binders do air harden to varying degrees. This has been surmounted by either storing the sand in closed hoppers or drums, or alternatively covering it with moist sacking. If these simple precautions are taken, sands may be stored for lengthy periods.

b) When the sand is hardened in the core box or with the pattern in position, it is considerably more difficult to strip, particularly when there is a deep draw. Much of this difficulty may be overcome by rapping the core box or pattern prior to hardening the sand, as with conventional practice. Preferably, all boxes with a deep draw should be collapsible and patterns should have a generous taper, with no undercuts.

Early difficulties with stripping, often using poor pattern equipment, led to the popular but fallacious belief that the sand actually expanded on gassing. However, dilatometric experiments failed to show any expansion and the sand appears to be volume stable on hardening.

Where a poor quality pattern has to be used, it is preferable to use a modified silicate binder producing good green strength so that the pattern may be stripped in the green state and the sand then hardened.

c) Difficulties were experienced with the freezing-up of valves on CO_2 cylinders and liquefiers. This problem was actively tackled by the various carbon-dioxide suppliers and has now been overcome by the use of a variety of valves and reservoir systems.

d) Inferior collapsibility was obtained quite frequently in early experiments but laboratory investigations and practical trials showed that by means of modified binders, additions to the sand, hollow cores etc., satisfactory collapsibility is possible with the majority of castings.

The one main question that remains to be solved is that of sand reclamation and of introducing CO_2 sand into a green sand system; preliminary information suggests that green sand may tolerate quite moderate proportions of CO_2 sand without any deterioration in properties. In addition, Schumacher[8] has recently described an investigation in a large German gray-iron foundry where the old CO_2 sand has been added into the green sand system for the last six months; the soda content of the sand has varied between 0.04 per cent and 0.25 per cent and the strength and general properties of the sand have shown no alteration. He also mentions that old CO_2 sand is even being re-used, after mixing with about equal proportions of new sand, for core production.

A preliminary examination of sand reclamation by wet methods is described later in this paper.

Theory of the Process

It has been generally accepted that the bond in the carbon-dioxide process is silica gel produced by the decomposition of the alkaline silicate. Chemical equations, such as

$$Na_2SiO_3 \text{ aq.} + CO_2 = SiO_2 \text{ aq.} + Na_2CO_3 ,$$

have been used to represent the reaction but there is now considerable evidence to show that they are very much an over-simplification and, in many ways, incorrect.

An extensive investigation of the mechanism of the bonding reaction has been carried out by my colleagues, Mr. R. F. Boddey and Mr. F. Ravault, and this has led to the development of the binder used in the experimental work described in this paper. The theory on which the composition of the binder is primarily based is outlined briefly below.

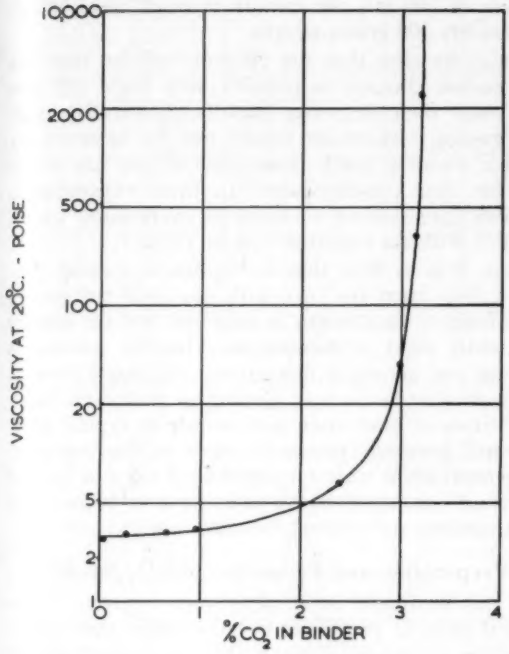
In most alkaline silicates, the main constituents are probably the metasilicate and disilicate. At normal concentrations, these silicates are ionized into Na^+, H^+ and SiO_3^{--} ions. The silicate ions are probably large and complex and their complexity increases as the silica content of the silicate increases; there is much evidence to show that they exist in the form of micelles which are aggregates of a large number of simple ions together with an appreciable number of water molecules.

Chemistry of Reactions in CO_2 Process

There are two main schools of thought on the chemistry of the reaction of these alkaline silicates with carbon dioxide. One theory suggests that reaction with carbon dioxide results in the elimination of water molecules by condensation between molecules of silicic acid with the consequent formation of chains or filaments of siliceous materials which link and enmesh together producing a rigid gel.

The other theory maintains that an aggregation of siliceous particles occurs with the formation of a network which is essentially hydrated cristobalite. The evidence for this latter theory seems more conclusive in that X-ray diffraction patterns show that freshly formed silica gels have centers consisting of crystalline cristobalite. It is further suggested that these centers cannot grow because the initial formation of these crystals increases the viscosity to such an extent that the free migration of ions or molecules is prevented. Consequently, it may be deduced that, although the decomposition of silicates is initially rapid, the rate of reaction decreases rapidly; experimental evidence to support this is presented later in this section.

The characteristic increase in viscosity of silicates on reaction with carbon dioxide is illustrated in Fig. 1. The results shown in this graph were obtained by passing diluted carbon dioxide through the binder and it is interesting to observe that initially silicate ions are stable to small quantities of weak electrolyte, such as carbonic acid. However, at a critical electrolyte content, the solvated layer surrounding silicate ions is removed partially so that discharge of the silicate ions by carbon dioxide occurs with the formation of the silica gel. The longer the time of treatment with carbon dioxide, the greater is the amount of gel formation until the viscosity increase is so great as to prevent further reaction.



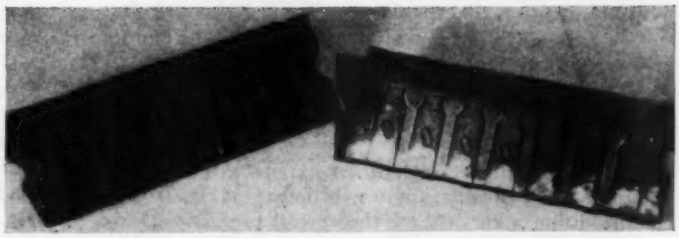


Fig. 1 (Left) — Variation of the viscosity of the binder with carbon dioxide. Fig. 2 (Above) — Showing whitish deposit produced by overgassing.

If the hardening of sands bonded with silicate-based binders and carbon dioxide is now considered in the light of these theories, several interesting points arise. The initial rapid hardening of sand compacts on treatment with carbon dioxide is due to the thickening of the binder consequent upon the formation of silica gel. The viscosities obtained are such that the film of binder may be considered as solid rather than liquid.

Analysis of sand compacts shows that, at this stage, a little over 50 per cent of the binder has reacted with the gas; further treatment with even excessive quantities of gas only increases slightly the amount of decomposed binder due to the high viscosity inhibiting ion migration and reaction. However, this rapid reaction of a proportion of the binder in the early stages imparts sufficient strength to the sand so that it may be easily handled and used for casting immediately.

Investigations in both laboratory and foundry have shown that excessive treatment with carbon dioxide results in deterioration of the strength of sand cores, particularly on standing; this phenomenon is known as "overgassing" and is generally characterized by the appearance of whitish crystals on the core surface (Fig. 2). Chemical analysis has shown that these crystals are sodium bicarbonate but it is not yet clear whether the formation of this material is responsible for the ill effects of "overgassing" or is simply coincidental. However, it forms a most useful indicator for the onset of "overgassing."

In all probability, the cause of "overgassing" is a breakdown of the silica gel which has been produced by initial treatment of the binder with carbon dioxide. The strength and plasticity of silica gels is markedly dependent on their water content, moisture-free gel being very brittle and having very little strength; experimental work by ourselves and also by other workers suggests that a molecule of silica should be associated with approximately six molecules of water for maximum strength properties. Thus, in all probability, "overgassing" results in the loss of water from the gel, possibly by one or more of the following three mechanisms:

a) loss of moisture by evaporation which is accelerated by flow of gas over the gel,

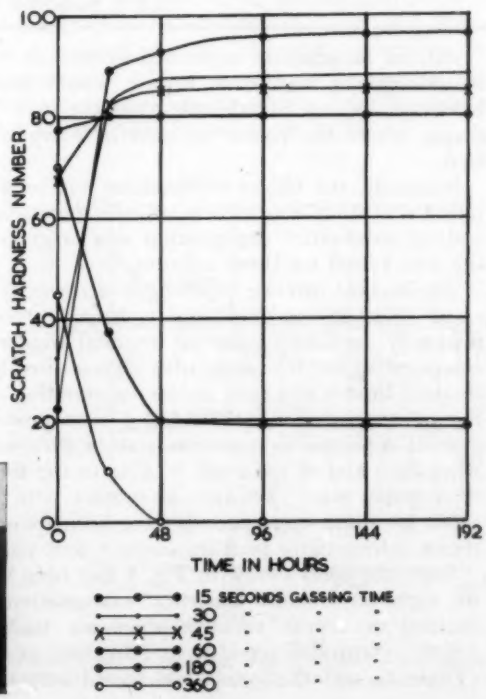


Fig. 3 (Right) — Influence of gassing time on scratch hardness of cores on standing. Fig. 4 (Below) — Cores gassed for varying periods of times after standing for 8 days.



b) hydration of crystals of sodium carbonate that are precipitated as decomposition of the binder proceeds,

c) formation of crystals of sodium bicarbonate, the necessary water being obtained by dehydration of the silica gel.

The following results of an experimental investigation into "overgassing" will help to clarify and substantiate some of these theoretical predictions. Six cores from the same sand mix were gassed for 15, 30, 45, 60, 180 and 360 seconds; the scratch hardness number of these cores was determined immediately after gassing and then at daily intervals. It may be seen from Fig. 3 that, for the cores gassed up to one minute, the initial hardness values increase rapidly but then markedly decrease with further gassing up to three and six minutes.

Similarly, the first four cores all increase in hardness on standing, the maximum hardness being obtained with the shortest gassing time whereas the hardness of those cores grossly "overgassed" deteriorate on standing, to a scratch hardness number of zero in the case of the six-minute gassed core after two days. The appearance of these cores after eight days standing is illustrated in Fig. 4.

Chemical analysis of these cores shows a very rapid initial increase in the carbon dioxide content of the first core followed by only a further slight increase with the remainder (Table 1).

TABLE 1 — CHEMICAL ANALYSIS OF CORES GASSED FOR VARYING PERIODS OF TIME

Gassing time, sec	0	15	30	45	60	180	360
CO_2 content, %	0.01	0.350	0.338	0.357	0.388	0.450	0.412*
Na_2CO_3, %	—	0.640	0.660	0.615	0.570	0.490	0.380
$NaHCO_3$, %	—	0.050	0.100	0.170	0.220	0.350	0.390
% binder reacted.	—	57%	62%	62%	65%	65%	65%

*N.B. % binder low in sand.

Sodium bicarbonate is formed even with only 15 seconds gassing and, from this, it would seem that moderate sodium bicarbonate contents may be tolerated before the commencement of overgassing effects.

Naturally, the binder composition has been so adjusted that there is moderate latitude to gassing times and no substantial depreciation was observed until the core gassed for three minutes.

The marked increase in strength of correctly gassed cores (Fig. 3) on standing is, in all probability, primarily associated with the residual binder undecomposed by the treatment with carbon dioxide. Such residual binder sets even harder on standing or stoving at temperatures of 212-392 F due probably to general moisture evaporation and to dissolution of silica sand and of silica gel, leading to the formation of a glassy bond. Silicates in contact with normal sands for quite short periods have been found to increase substantially in silica content and viscosity.

Since the cores shown in Fig. 4 had been standing for eight days before analytical examination, it was decided to repeat these experiments under more strictly controlled conditions. In the next series (Table 2), carbon dioxide was passed very slowly at the rate of 200 mls per minute through cores of approximately 100 grams weight.

It may be seen that the efficiency of the reaction with carbon dioxide is initially very high (98 per cent) but then decreases rapidly. Naturally, such high gassing efficiencies would not be achieved in practice where a much faster flow of gas has to be used for time considerations. In these experiments, the fifth core showed evidence of overgassing which is in line with the results shown in Table 1.

Thus, it is evident that the optimum gassing procedure is to treat the core with gas until the initial rapid increase in strength is obtained and the core is sufficiently rigid to handle; any further gassing is wasteful and generally deleterious. Although there is a considerable latitude in gassing so that even treatment times of more than one minute at typical flow rates still generally produce a very satisfactory core, the general 20/20 rule suggested by Talbot of having 20 seconds gassing at 20 psi pressure is to be strongly recommended for general foundry work.

Preparation and Properties of CO_2 Sands

In this section, a detailed account is given of the general physical properties of CO_2 sands; these properties are, in general, comparable with those of conventional oil or resin sands but certain properties, such as permeability and gas evolution, do show marked differences. In addition, an account is given of the collapsibility and surface finish obtained with the process.

So far, the process has been found satisfactory for bonding a wide range of silica sands and also zircon and chamotte/silica sand mixes; it is preferable for the sand to be clean, cold, dry and free from active clays and, unless otherwise stated, the properties quoted here have been obtained with such a sand.

Preparation of sand. The introduction of the binder into the sand may be treated in exactly the same way as with oil or resin binders. During the trial and adoption of the process by many foundries in Great Britain, a wide range of sand mixing equipment has been satisfactorily employed; the general technique is to mix the dry sand and any powder additions, such as coal dust or pelletted pitch, for about one-half to one minute and then add the viscous binder and mix for a further two to four minutes.

As long as the mixing is thorough, the type of plant employed is not critical; essentially, only a simple mixing operation is required as it is generally not necessary to develop any green strength in the mix. The gassed strength of the sand varies slightly depend-

TABLE 2 — CHEMICAL ANALYSIS OF CORES GASSED WITH VARYING AMOUNTS OF CARBON DIOXIDE

	1	2	3	4	5	6
Wt. CO_2 passed, grams	0.208	0.288	0.384	0.702	1.877	3.645
Initial wt. core, grams	137.1	135.7	128.7	131.9	127.2	125.0
% CO_2 absorbed by core	98.1	97.9	96.5	62.0	35.3	19.6
% Na_2CO_3	0.359	0.501	0.694	0.795	0.603	0.485
% $NaHCO_3$	nil	nil	nil	nil	0.519	0.706

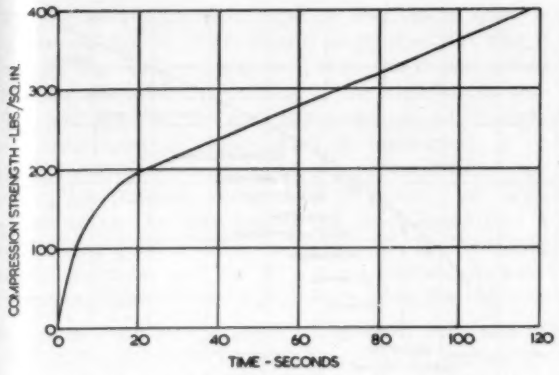


Fig. 5 — Variation of compressive strength of cores with time of gassing.

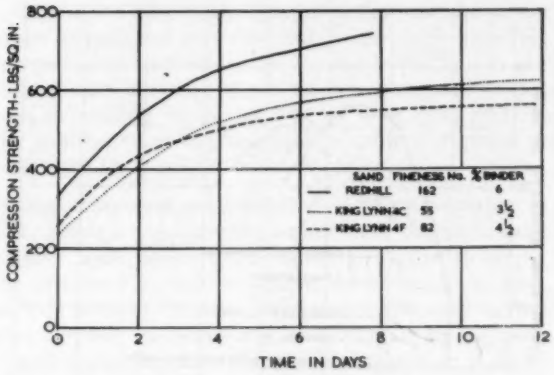


Fig. 6 — Variation of compressive strength of gassed cores on standing.

ing on the type of mixer or mill employed; for example, a simple horizontal or vertical blade-type mixer gives higher strength than edge-runner mills. This is probably due to a certain amount of premature hardening of the binder by reaction with carbon dioxide in the atmosphere during vigorous milling.

Hardened Strength. The compressive strength obtained with most sands immediately after gassing is of the order of 200 to 300 psi and this increases under favorable conditions to about 500 to 700 psi on standing in a normal foundry atmosphere for several days. However, it must be emphasized that cores and molds are perfectly suitable for casting immediately after gassing; for example, the gear-box casing core shown in Fig. 33 has been used in a casting in less than three minutes after production. The strength figures obtained are certainly less than has generally been accepted for many cores bonded with oils and resins where compressive strengths as high as 1200 psi are not exceptional.

However, except for such applications as centrifugal casting, it is generally agreed that such strengths are unnecessarily high and the carbon-dioxide process has been instrumental in establishing this point. It should be added that strengths well in excess of 1000 psi may be obtained if CO_2 sands are given a slight stoving at temperatures in excess of 230 F.

The actual gassed strength depends upon a large number of variables but, assuming that a dry, cold, clay-free sand is used, compressive strengths of between 200 to 300 psi are obtained in less than one minute gassing time; this is illustrated in Fig. 5 which shows the fast initial rate of hardening of the sand. With the particular gassing method used in these tests, the compressive strength values continued to rise after comparatively lengthy treatments with gas but the other properties of the cores showed a general deterioration. This is due to the phenomenon of overgassing discussed in detail in the previous section on the theory of the process. Essentially, prolonged treatment with carbon dioxide results in an embrittlement of the bond and in a rapid deterioration of the core or mold on standing. See Figs. 3 and 4.

Overgassing was not appreciated in the early days of the process when it was quite common but fallacious practice to be lavish with gas application if it was desired to obtain a particularly hard core. However, an overgassed core can be readily recognized by the appearance of whitish crystals or powder on the sand surface; cores or molds showing such a white deposit are invariably found to be friable.

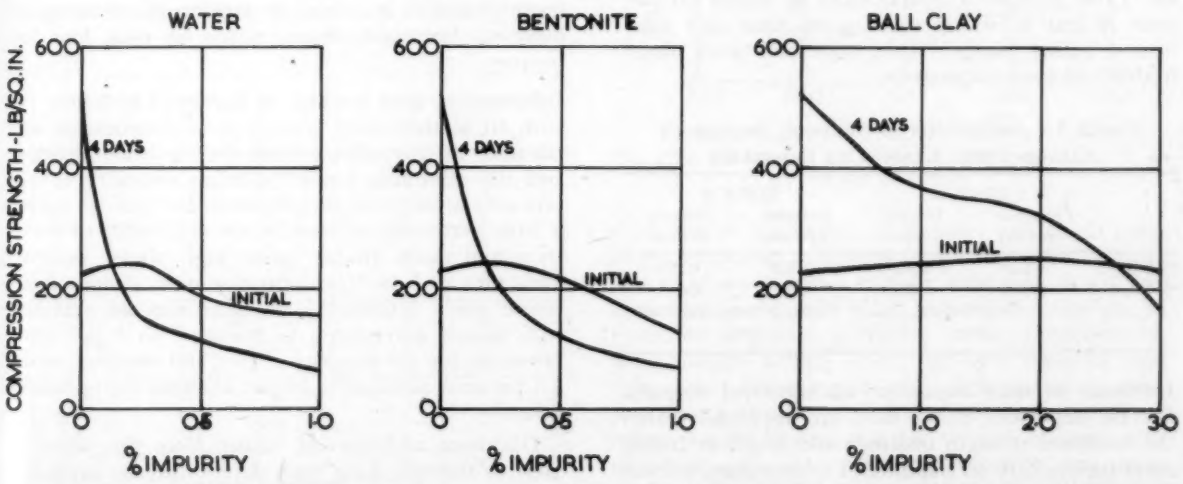


Fig. 7 (Left) — Influence of water on strength of CO_2 cores on standing. Fig. 8 (Center) — Influence of bentonite on strength of CO_2 cores on standing. Fig. 9 (Right) — Influence of ball-clay on strength of CO_2 cores on standing.

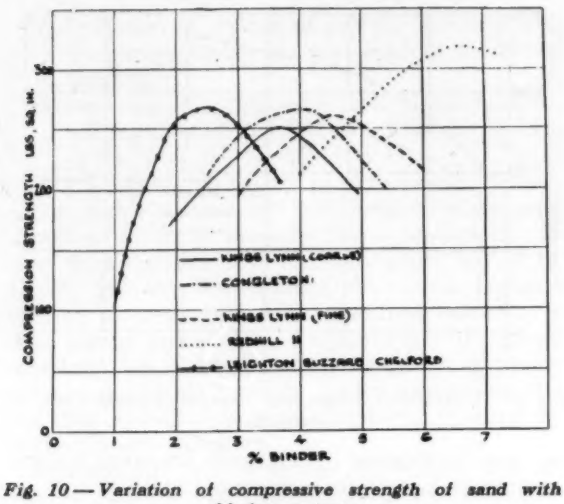


Fig. 10 — Variation of compressive strength of sand with binder content.

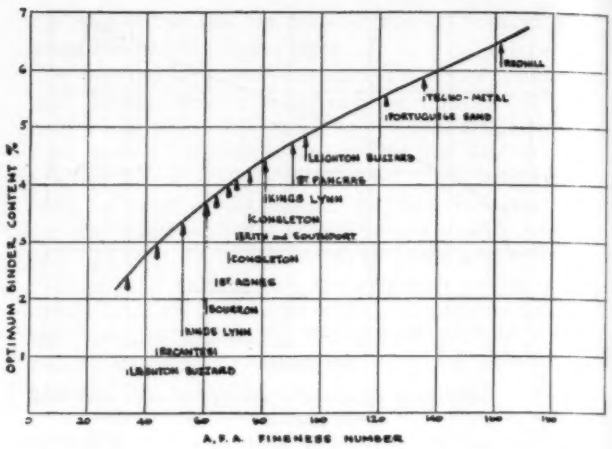


Fig. 11 — Variation of optimum binder content with AFS fineness number of sand.

On standing, cores that have been correctly gassed to obtain the initial set show an appreciable increase in strength. (Figs. 3 and 6). Thus, in a matter of a few hours, there is a marked increase in compressive strength and scratch hardness, followed by a falling off in the rate of increase but it is significant that the cores continued to increase in strength properties even after 14 days; this is confirmation of the advantage claimed with regard to core storage.

Certain cases have arisen where cores that have been correctly gassed have markedly deteriorated on standing. This has occurred primarily with cores that have been stored in sulphurous atmospheres and analytical investigations have shown that such friability is due to the absorption of sulphur dioxide from the atmosphere. Although the exact mechanism of the reaction involved is not clear, sulphurous acid, being a much stronger acid than carbonic acid, is able to displace carbon dioxide from the gassed core and also react with undecomposed binder. The resultant sodium sulphite produced is quite rapidly oxidized to sodium sulphate during subsequent exposure to air. Table 3 shows a typical series of results on two cores A and B, which were stored near and away from a rather pungent core stove and were found friable and hard respectively.

TABLE 3 — ABSORPTION OF SULPHUR DIOXIDE BY GASSED CORES LEADING TO FRIABILITY

	CORE A		CORE B	
	Exterior Surface	Interior Surface	Exterior Surface	Interior Surface
% CO_2	0.215	0.300	0.300	0.290
% SO_4	0.315	0.025	0.015	0.015
	Very friable core.		Hard and very satisfactory core.	

Influence of sand impurities on hardened strength. Certain impurities in the sand can appreciably alter the hardened strength obtained and result in friable cores particularly on standing. The two main deleterious properties are moisture and active clays such as bentonites and, to a lesser extent, ball clays.

Figure 7 shows the influence of moisture in the sand on the initial hardened strengths and on the strength on standing. It may be seen that the initial set strength of the sand is not drastically reduced by the presence of small amounts of moisture but that the strengths on standing decrease rapidly causing the cores to become friable and unusable. It is of academic interest that the cores do eventually, after about a week, begin to pick up again in strength; this no doubt occurs after the water present in the core has evaporated and the residual binder present begins to thicken up.

In addition, the presence of water in the sand increases the danger of blowing defects on casting.

Similar deteriorations in strength properties are obtained with the presence of active clays in the sand. Figure 8 shows the very detrimental effect of small quantities of bentonite and similar results, although much less severe, are obtained with clays of the ball- and fireclay-type (Fig. 9). Consequently, the incorporation of small quantities of ball clay in the sand up to about one per cent is moderately satisfactory when it is desired to develop green strength; however, bentonite should never be used for this purpose.

Influence of sand grading on hardened strength. As with all binders, sand grading is of considerable importance in determining both the hardened strength and the minimum binder addition necessary to obtain satisfactory strength properties for foundry usage; a large percentage of fines in the sand tends to make cores and molds friable unless high binder percentages, such as 5 to 7 per cent, are used whereas with coarse sands, satisfactory strengths may be obtained with binder percentages as low as 2 to 3 per cent. However, for the majority of core and molding sands, a 4 per cent addition of binder is about the optimum quantity.

Optimum additions of binder from the point of view of strength have been determined by investigating the variation in strength with binder content when gassing AFS cores for a constant time. A typical

series of results for fine, medium and coarse sands is illustrated in Fig. 10. It should be pointed out that if gassing was further continued, then the curves would not show a strength reduction at the high binder percentages; however, the gassing time employed throughout this series of experiments is considered to be representative of the general gassing conditions prevailing in practice. More than 50 sands have been examined in this way and it has been found that a moderately good relation exists between the optimum binder contents and the AFS grain fineness number of the sand; relationship is illustrated in Fig. 11.

Effect of sand temperature. The main influence of hot sand is to cause a certain amount of premature hardening so that the sand tends to develop a certain amount of green strength and becomes unworkable. In addition, hot sand mixes tend to air-harden rapidly and it is for these reasons that cold sand is recommended.

Green strength, flowability and shatter index. When the core or mold is hardened in the box or with the pattern in position, green strength is of little or no importance. The green strength of sands bonded with binders for use in the carbon-dioxide process is low, usually varying between ½ to 1 psi and depending upon the milling time, type of sand and on sand temperature. However, for certain methods of gas application and when using poor quality pattern equipment, it is desirable to strip green and for such cases, green strength CO_2 binders have been developed; such binders give green strengths varying between 2½ to 4 psi. Alternatively, materials such as ball clay or dextrin may be milled into the sand in quantities up to about one per cent with no appreciable deterioration in gassing strength.

For many of the applications where stripping green is desirable, many foundrymen have used the low green strength binder and have relied on the rapid formation of a hardened skin on the sand surface which prevents distortion.

The flowability of CO_2 sands in the ungassed state is high and values determined by the deformation test on the standard ramming apparatus are in the region of 80 to 90. These figures vary very little with the sand used or with increase of binder content.

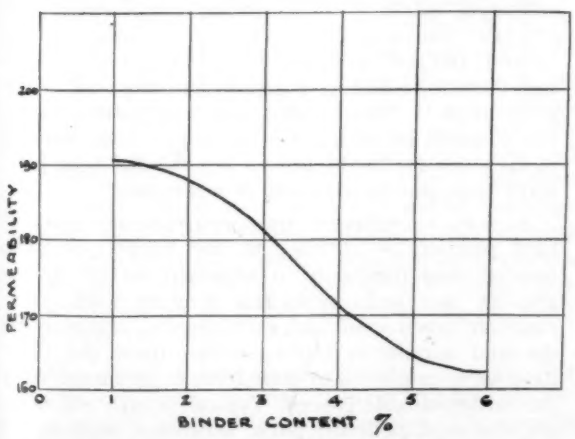


Fig. 12 — Variation of permeability with binder content.

The more informative shatter index test shows the sands to compare very favorably with highly flowable conventional core-sands; the shatter index figures obtained vary between about 60 per cent and 75 per cent depending upon the type of sand used and on the milling time but the binder content appears to have very little influence.

Permeability and gas evolution. The permeability of CO_2 sands is significantly higher than the same sand bonded with conventional binders; Figure 12 shows the influence of binder content on permeability and, as with normal binders, a reduction generally occurs with increase in binder content. Increase in the time of gassing with carbon dioxide also results in a slight increase in permeability. The sand used in the experiments illustrated in Fig. 12 had an AFS fineness number 75 and gave a dry permeability of 135 with a typical oil binder whereas the value with the recommended addition of the CO_2 binder was 172.

In addition to this, the volatile content of CO_2 sands is very low being approximately one-third to one-quarter that evolved from normal core sands. This is illustrated by Fig. 13 which shows that the gas evolved from a CO_2 sand is approximately 7 to 8 cc per gram at 1830 F with a four per cent binder addition. This low volatile content, coupled with the higher permeability, generally results in a marked reduction in the incidence of blowing defects on castings. However, with certain castings, the rate of gas evolution is of greater importance than the total gas content.

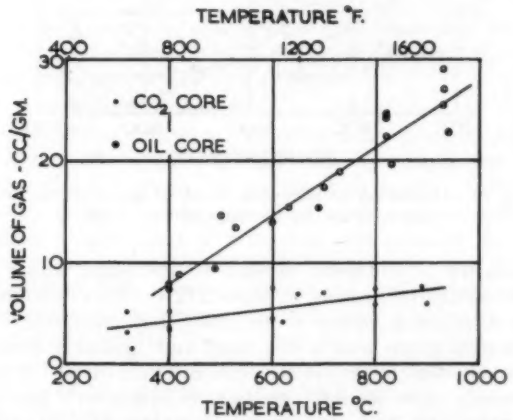


Fig. 13 — Variation of gas evolution with temperature for an oil and CO_2 sand.

Figure 14 is a graph of the start and finish times of gas evolution at various temperatures for both oil and CO_2 sands and it may be seen that, at the low temperatures of light alloy casting, the rate of gas evolution is greater with CO_2 sands. Consequently, with castings having totally enclosed cores or with thin-section diecastings, blowing defects may be more prevalent with CO_2 sands due to the evolution of gas before a solid casting skin has formed whereas, with an oil or resin core, little gas evolution may occur until a substantial thickness of metal has solidified at the metal/core interface.

Although this blowing difficulty has not occurred to any appreciable extent, it should be mentioned that it could be readily eliminated by a short stoving at temperatures around 230 F as the majority of gas evolved is steam.

Thermal Properties. Early work on the carbon-dioxide process suggested that the bonded sands might have a higher chilling power than the same sand bonded with conventional oil or resin binders. Consequently, determinations of the thermal conductivity have been made and the results are illustrated in Fig. 15.

It may be seen that the apparent thermal conductivity of a sand bonded with the CO_2 binder is initially about five times as high as the same sand bonded with oil but that the difference rapidly disappears on heating. This effect is due to the small

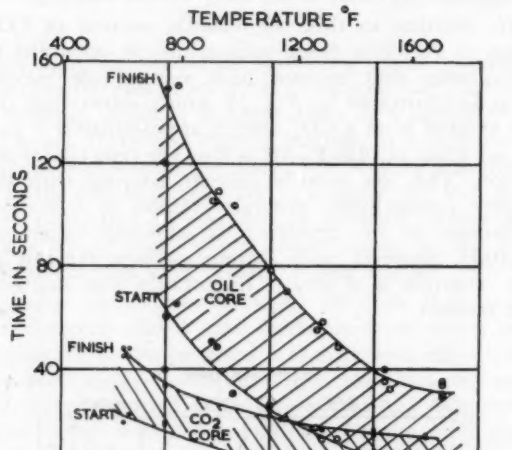


Fig. 14 — Variation of start and finish of gas evolution with temperature for an oil and CO_2 sand.

quantity of moisture present in the sand which is eliminated by heating to about 212 F. This difference in conducting power is no doubt less than that between a green and a dry sand and probably would not be detectable on any castings but those of extremely thin section. In fact, Schumacher[8] has recently carried out temperature measurements on 2-inch diameter cores surrounded by steel and his results show that the conducting properties of a CO_2 sand and a steel core-sand are very similar and slightly less than that of a bentonite-bonded molding sand.

Surface Finish. During the early trials with the carbon-dioxide process on the foundry floor, considerable variation in surface finish of castings was observed. The two main variables, apart from the fineness of the sand, appeared to be the type of metal and the section thickness of casting. Now, with our general increase in knowledge of the process, the reasons for such variations are apparent and have resulted in the development of various techniques and special binders so that with iron and copper-base

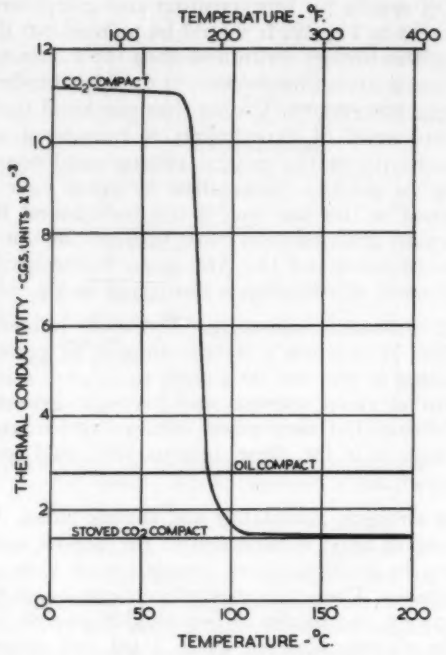


Fig. 15 — Variation of thermal conductivity of oil and CO_2 sands with temperature.

alloys, where the surface finish was initially disappointing, very satisfactory finishes have been obtained as is illustrated by many of the photographs at the end of the paper and by the widespread adoption of the process in foundries casting these metals.

With light alloys, the surface finish is equivalent to that obtained with conventional oil or resin binders. At the temperatures prevailing in light alloy castings, no chemical reaction occurs between the sand and the binder and the surface finish is a direct function of the sand-grain size and the degree of ramming. Where superior surface finish is required, conventional coatings based upon mica, talc etc. have been used most successfully.

At temperatures in excess of about 1475 F, any residual binder and sodium carbonate begins to react and sinter with the sand. Such sintering occurs to an increasing extent with increasing temperature and generally results in an increase in the size of pores between the sand grains at the mold or core surface and thereby results in a greater incidence of metal penetration.[10] Metal penetration and burnt-on sand was observed on many iron and copper-base castings in early trials with the process but this has been generally overcome by a variety of techniques.

As may be expected from conventional foundry sand practice, an increase in the fineness of CO_2 bonded sand results in an approved surface finish; also, in iron and copper-base foundry work, it is common practice to add carbonaceous materials to the sand in order to improve surface finish and these have been employed to considerable advantage with the carbon-dioxide process. Typical of such additions are coal dust, pelletted pitch, woodflour, asphalt, oil and also graphite; all of these have been employed

with CO_2 sands but coal dust and particularly woodflour have the disadvantage of tending to produce friable cores and molds. Pelletted pitch, however, results in slightly improved strength properties particularly with regard to edge hardness of cores.

The improved surface finish obtained with carbonaceous additions is illustrated by Fig. 16 which shows the influence of these materials on the surface finish of an 85/5/5/5 gunmetal and a typical gray iron. The surface finish ratings used in these experiments were obtained by pouring the metal on to the surface of standard AFS compacts to a height of about 8 in. in a standard test casting, and assessing the surface finish visually and by touch against a series of arbitrarily numbered standards ranging from a poor (surface rating 0) to a very good surface finish (surface rating 6). This method, although empirical and susceptible to personal error, has been found to give consistent results and the assessments, carried out usually by six independent observers, do not generally vary by more than half a point. It is perhaps of general interest that this technique is being employed by Sub-Committee T. S. 48 of the Institute of British Foundrymen in an extensive investigation of factors influencing surface finish.

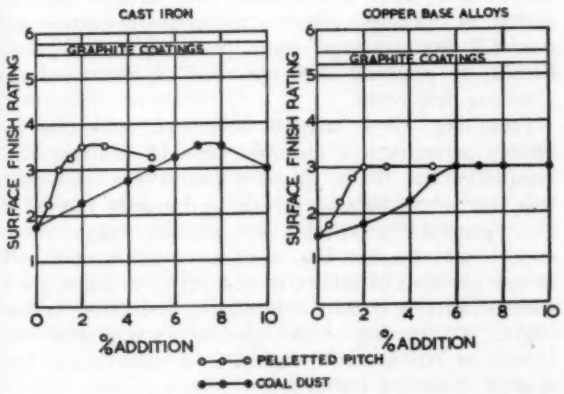


Fig. 16 — Variation of surface finish with coal dust and pelletted pitch additions on iron and gunmetal.

The results obtained with coal dust show that an addition of about 6 per cent is approximately the optimum amount. The metal surface adjacent to the core containing no addition was covered by a firmly adhering layer of sintered sand which was removed only with difficulty. With increasing coal dust contents, the fused layer was reduced in thickness and was eliminated at a content of 5 per cent coal dust. Additions of 6 per cent or more produced a comparatively smooth surface finish which would be more than satisfactory for general production requirements. The deterioration in surface finish with the excessive coal dust content of 10 per cent is due to slight blowing from the core.

The disadvantages of coal dust are that it gives rise to a moderate amount of fume and also encourages surface friability in cores; however, this material is used very successfully and extensively on a production scale as an additive to CO_2 sands.

The tests with pelletted pitch show that it is more efficient on a percentage basis in producing a good surface finish, and an addition of 2 per cent gives a surface comparable to that obtained with 6 per cent coal dust. In addition to the beneficial effect of pitch on sand properties, there is a reduced amount of fume compared to coal dust.

Naturally, the necessary amount of carbonaceous material for satisfactory surface finish depends upon the section thickness of the casting, the pouring temerature and the type of sand; the figures of 2 per cent pelletted pitch and 6 per cent coal dust should be considered as maximum amounts and smaller quantities are frequently employed for thin section work. This is illustrated by the following empirical rule which has been employed successfully by several foundrymen for section thicknesses in iron up to about two inches: "An addition of a percentage of coal dust equal to twice the section thickness or a percentage of pelletted pitch equal to the section thickness is used."

Some foundrymen have used additions of graphite to improve surface finish preferring this material because of the reduced amount of fume; however, although very good finishes have been obtained, the results seem more variable and graphite has the disadvantage of being relatively expensive.

When, as with conventionally bonded sands, a high quality surface finish is required, the use of coatings is essential unless very fine sands are employed. It may be seen from Fig. 16 that graphite coatings, either carried in an inflammable or water medium, result in a considerable improvement in surface finish. Graphitic coatings have been found the most suitable for both copper-base and iron castings.

The surface finish obtained with steel is, in many ways, the converse of orthodox sand practice. With thin castings, up to ¾ to 1 inch in section, the surface finish is generally not so good as that obtained with normal sands. However, with sections greater than 1 inch, there is a marked improvement in surface finish and many steel castings are now being made in CO_2 sands without the use of refractory silica or zircon dressings which were previously considered essential. This is due to the formation of a viscous slag layer at the metal/mold interface. Analysis of this slag shows it to be a ternary mixture of $Na_2O/FeO/SiO_2$ and the extent of its formation increases with both pouring temperature and section thickness of casting. Previous investigations have shown on a laboratory scale that the formation of such slags can be most effective in preventing metal penetration and improving surface finish.[10]

Where metal penetration has occurred with steel castings, conventional paints have been used successfully; in Germany and, to a lesser extent, in Great Britain, the rather surprising but very successful technique of coating molds and cores with thick water-base dressings that are simply allowed to air dry for periods of the order of ten hours has been found most successful. Also, as with conventional sand mixes, the incorporation of fine sands or flours to obtain maximum bulk density combined with minimum permea-

bility has been used to considerable advantage for producing good finish.

With alloy steels, the surface finish is good and there is a reduced incidence of cold-lap formation which is often prevalent with normal sands. It is generally accepted that cold-lap formation is due to the formation of oxide skins on the molten metal surface and it seems that such films are fluxed away with CO_2 sands.

Collapsibility. Contrary to the many theoretical predictions which were made on the collapsibility of CO_2 cores when the process was first publicized, there can now be little doubt that the collapsibility is satisfactory for most applications and for metal ranging from light alloys to alloy steels. Thus, Talbot[7] has stated that, with iron and aluminum, he has not had trouble at any stage with knockout or breakdown for castings ranging in weight from ounces to tons and in section thickness from ⅛ inch in aluminum to 18 inches in cast iron. Also, in addition to many other authors and lecturers on the process, Schumacher,[8] Rees[11] and Parramore[12] have shown how satisfactory collapsibility is obtained on all metals with the use of sand additions in certain instances.

In order to obtain a better understanding of the factors affecting the collapsibility of CO_2 cores, an extensive investigation of the strength of CO_2 sands at varying temperatures has been carried out in the author's laboratories. By heating small, one-inch high and one-inch diameter, cores in an electric furnace up to temperatures of approximately 2190 F, in 212 F intervals, the compressive strength/temperature curves of CO_2 sand, with and without various additions, were determined.

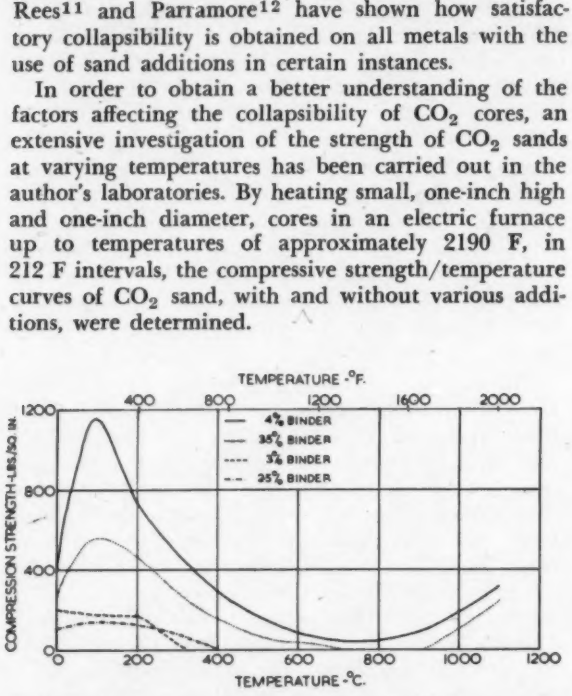


Fig. 17 — Variation of compressive strength with temperature for CO_2 sands of varying binder content.

Figure 17 shows a typical series of curves for sand compacts containing varying contents of the proprietary binder used throughout the investigations described in this paper; it should be mentioned that this binder contains additions added specifically to assist collapsibility. The shape of the curve for the 4 per cent binder addition may be considered typical, as this is the optimum binder addition recommended for good strength and handling properties for this particular sand.

It may be seen that the strength of the sand varies markedly with temperature; on heating to about 212-392 F, there is a marked increase in strength leading to compressive strengths of the order of 1200 psi. If the core is heated to about 212 F for only very short periods of time, there is initially a decrease in strength probably due to loss of water from the silica gel, but this is soon superseded by the considerable strength increase. Further heating above 392 F leads to a steady loss in strength and the sand is completely collapsible by about 1022-1112 F and remains so until heated to about 1560 F when the strength gradually begins to climb due to the partial sintering of the binder. As will be explained later, this increase in strength above 1560 F can be markedly reduced or even eliminated by making certain additions to the sand.

By measuring the temperature attained in varying size cores surrounded by different sections of metal, it has been possible to correlate the strength temperature curves with collapsibility results observed in practice. Thus, the temperatures attained in the center of 2-inch diameter cylindrical cores after casting ⅜-inch and 1¼-inch section thickness of metal around them were 662 F and 842 F respectively, for aluminum at 1328 F pouring temperature and 1292 F and 1652 F respectively for iron cast at 2552 F. From both these examples and the strength temperature curves it may be predicted that slightly better knockout should be obtained with thicker section aluminum castings, whereas the converse should be true for iron castings; these predictions were confirmed by practical shakeout tests on the two-inch diameter test cores.

From Fig. 17, it may be seen that reduction in binder percentage is most effective in assisting collapsibility but, below about 3 per cent binder content, the cores became friable and would not have been generally acceptable for foundry usage. However, it may be seen that even a reduction of a half or one per cent of binder is most effective when good collapsibility is required. Similarly, reduction in the degree of ramming of the core favors improved collapsibility but naturally results in a reduction in the quality of casting finish.

As with surface finish, the addition of carbonaceous material has been found most effective in aiding collapsibility with copper-base alloys and iron; these additions primarily influence the strength curve above 1560 F where sintering of sand and binder occurs and consequently are of little value for light alloy work. Coal dust, pelletted pitch, graphite, asphalt and woodflour have all been employed satisfactorily although there is still the objection to woodflour and, to a lesser extent, to coal dust that they encourage rapid air drying of the mix and core friability.

The actual mechanism of these additions in overcoming sintering is still not clear but it may be seen from Fig. 18 that, with 1 per cent of pelletted pitch, there is no strength increase above 1560 F. The necessary additions of these carbonaceous materials for satisfactory collapsibility are similar to those required for satisfactory surface finish. For example, the optimum addition of pelletted pitch is of the order of one to two per cent; with higher additions, the pitch tends to impair collapsibility particularly in the center of large block cores or with thin section castings

when the pitch is not burnt-out and binds the sand grains together. It has been suggested that additions of fluorspar or calcium fluoride are most beneficial to the breakdown of CO_2 sands; experiments and practical tests have shown that only with high purity calcium fluoride is any appreciable improvement obtained.

Schumacher[8] has extended the type of curve illustrated in Figs. 17 and 18 and has shown that after about 2192 F, the strength of the sand again begins to decrease and he has suggested that this is responsible for the good knockout obtained with steel and certainly no additions have been found generally necessary with this metal.

Thus, to summarize, with light alloys, cores are generally heated to temperatures where sand strengths are very low or zero and consequently the collapsibility is very good and is generally considered to be superior to conventional oil or resin binders. This, in addition to the many other advantages of the process, probably explains the very rapid introduction of the carbon-dioxide process into foundries doing both sand and die castings. When difficulties are experienced with collapsibility it is suggested that the binder percentage should be reduced, or, alternatively a small addition of graphite, of the order of ½ to 1 per cent be made to the sand; this material has been found beneficial with light alloys but its influence appears to be due to its flake structure weakening any residual bond after casting rather than to its chemical composition.

With copper-base and iron castings of any appreciable section thickness, the addition of carbonaceous materials is generally essential and the most popular additions in Great Britain are pelletted pitch in amounts up to about 2 per cent, and coal dust up to 6 per cent.

For steel, no additions are generally made to the sand. Collapsibility is probably on a par with conventional steel-sand binders and the general appearance of a core on shakeout is a sintered or fused layer of sand adjacent to the metal surface with the interior layers of the core collapsing readily either into grains or small lumps of sand.

Methods of Gas Application

During the early stages of the introduction of the carbon-dioxide process into foundries, it became apparent that many of the inherent advantages of the process might be nullified if the method of applying the gas is cumbersome and lengthy. This is likely to be particularly true of small molds and cores in highly mechanized foundries. Consequently, considerable efforts have been made to streamline the application of gas and already much progress has been made.

It is intended in this section to describe present methods of producing cores and molds and, in addition, to outline possible new lines of approach. The desirable features of any gassing technique are:

1) To introduce the gas into the sand in the minimum of time.
2) To introduce the gas into all parts of the sand so that they are equally and effectively hardened.

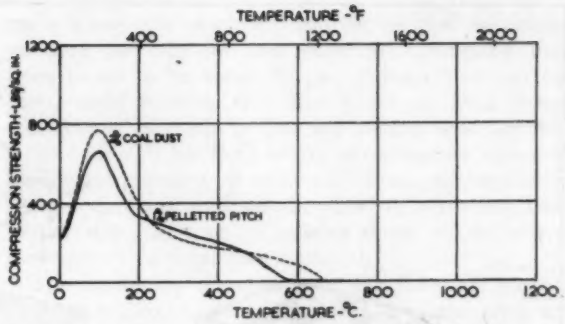


Fig. 18 — Influence of coal dust and pelletted pitch additions on shape of strength-temperature curve for CO_2 sands.

3) To have the minimum consumption of gas.
4) To have the minimum amount of manual work, and
5) To have no restriction on mold- or core-making production.

The importance of these requirements naturally depends on whether the foundry is doing jobbing or repetition work and on the size of mold or core produced. Thus, for a large one-off core, where the time of gassing is likely to be small compared to the total time of producing the core, comparatively primitive gassing techniques are acceptable and will still show substantial advantages over existing core-making methods. For a small core produced in considerable quantities on a core blower, delays due to gassing might result in a substantial reduction in output from the blowing unit.

Although the methods of gas application have been subdivided into those suitable for molds and those suitable for cores, it should be appreciated that some methods may be used advantageously for both classes of work.

Supplies of carbon dioxide. Carbon dioxide is available in Great Britain in three main forms: from cylinders of gas, generally of 28-lb size, from liquefiers of 50 to 100-lb capacity into which blocks of solid carbon dioxide may be inserted, and from low-pressure static tanks of several tons capacity which are filled from mobile tankers with liquid gas. Suppliers of gas have been most cooperative in the whole development, and systems of cylinders and liquefiers are now installed in many foundries so that carbon dioxide at suitable pressure may be piped around the foundry and be available at many points.

Valves for both cylinders and liquefiers did cause some difficulties initially as they tended to freeze up which restricted and even prevented any flow of gas. This freezing up is due to the absorption of heat as the liquid carbon dioxide in the cylinder vaporizes and as the gas expands on passage from the high pressure in the cylinder or liquefier, about 700 psi, to the application pressure of approximately 20 psi. However, this difficulty has now been successfully overcome and many suitable valves are now on the market. These are usually either of the so-called double reduction type whereby the gas is reduced to an intermediate pressure of 200 to 250 psi before

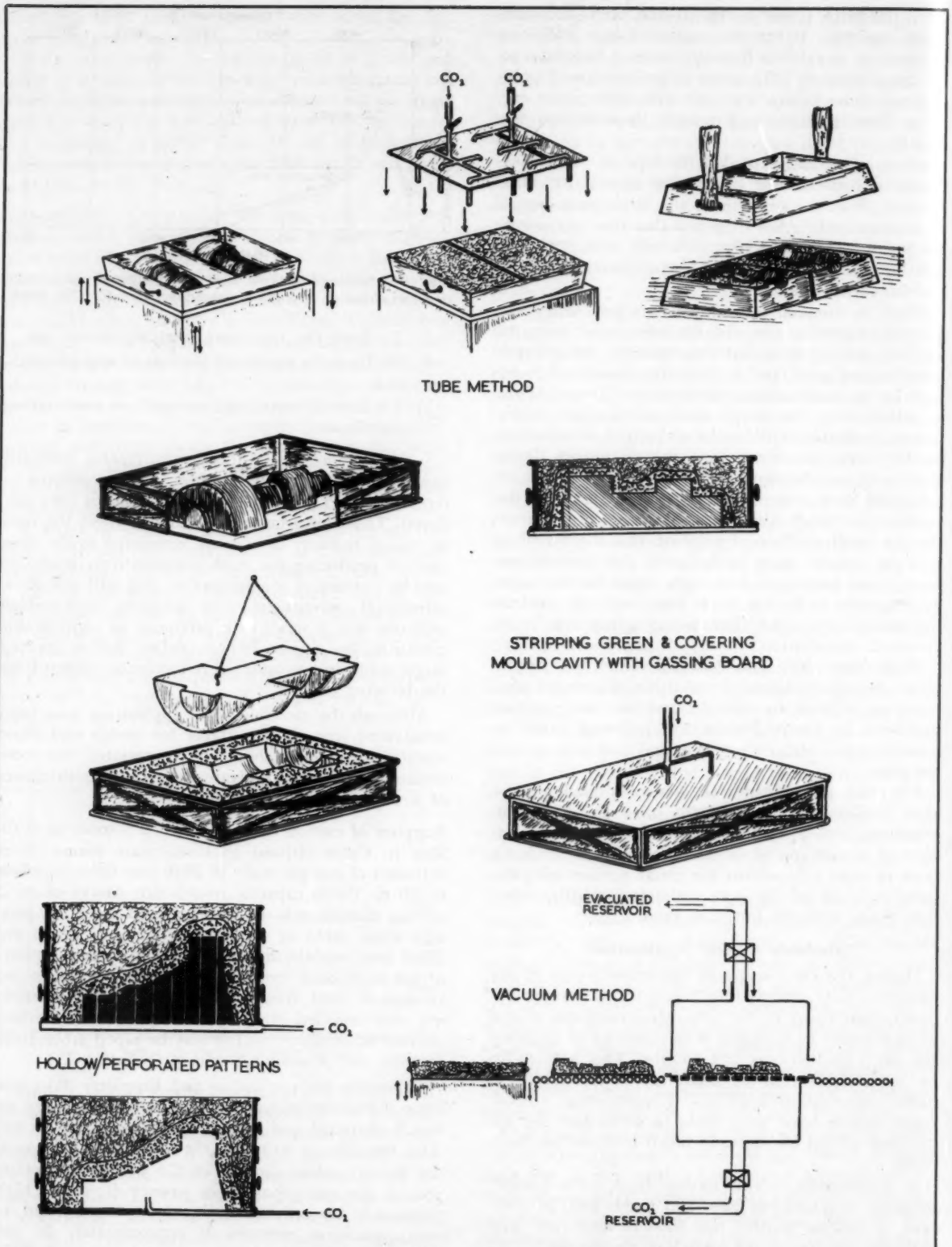


Fig. 19 — Typical methods of mold production by CO_2 process.

application, or valves actually incorporating a heating element to counteract the cooling.

The Production of Molds. When molds are hardened with the pattern in position, due attention must be given to the pattern equipment as the rigidity of the sand makes stripping more difficult than with green sand.

The ease of stripping both molds and cores can be considerably assisted if the following precautions are taken:

a) Any undercuts or dents in the box should be filled with plastic wood or some similar material; this prevents the sand keying into these imperfections which would otherwise hinder stripping and damage the mold or core.

b) The box should be rapped prior to gassing, in the same manner as with normal molding materials. This technique normally eliminates difficulties with stripping unless the pattern has a very deep draw or very little taper. Alternatively, the pattern can be partially stripped while green so as to break any undercuts and then returned to its original position for hardening.

c) Normal paints and varnishes on pattern equipment are of no assistance to stripping and can quite frequently be troublesome due to the softening of the paint by the alkaline binder; lacquers based upon nitro-cellulose are fairly trouble-free, but Talbot recommends using one coat of ordinary varnish and finds this most satisfactory. Alternatively, rubbing the surface of the pattern with graphite has been used to considerable advantage.

A recent development due to Parramore[12] which should have a very wide application is the use of wooden patterns and core boxes sprayed with metal using a simple wire-pistol gun. He has sprayed patterns with both aluminum and copper to a thickness of about 0.010 in. and has found that these coatings adhere better if an initial spray of about 0.003 in. of zinc is applied. Naturally these sprayed layers have to be rubbed down with emery prior to use but the quality of pattern obtained is exceptionally good.

(1) Hardening by Tubes. For initial experiments with the process, the majority of foundries adopt the most elementary method of hardening by introducing the gas into the sand by means of tubes. A typical example of this is illustrated in Fig. 19. The passage of gas down the tubes leads initially to rapid hardening in the region of the tube openings, and further hardening occurs as the gas flows outwards through the sand eventually escaping into the atmosphere. Using this technique, with introduction of the gas at about 20 psi and with an application time of about 15 to 30 seconds, a volume of sand of radius about three to four inches around the tube is effectively hardened. Passage of gas for any longer period of time is unlikely to increase substantially the volume of hardened sand and also increases the possibility of overgassing the sand leading to mold friability.

Apart from these technical considerations, excessive gassing times simply increase the rate of loss of gas to the atmosphere. Similar comments can be applied to the gas pressure employed and the general optimum pressure for efficient and economic hardening has been found to be about 20 psi. However, in certain cases, such as with very fine sands or with some mechanized methods of gas application that are described later, it has been found advantageous to use a higher gas pressure up to about 60 psi; this certainly reduces the time required for effective hardening but reduces the tolerance on gassing times so that overgassing may be more prevalent.

The hardening of molds and also cores by means of ¼-inch diameter tubes inserted into the sand at four- to six-inch spacings has been practised quite extensively on a production scale. The main disadvantages of the method are that

(a) It is wasteful on carbon dioxide.
(b) It is time consuming unless manifold gassing tubes are employed and even then it is not as rapid as other methods.
(c) It tends to cause uneven gassing, particularly as air-pockets are often formed in re-entrant portions of the pattern resulting in ineffective hardening.

However, the tube method is used with marked success, particularly for the production of large molds and cores since the time of gassing in such cases is comparatively unimportant. A semi-mechanized method that is used on a production scale is illustrated in Fig. 19. A heavily tapered molding box is placed over two metal half patterns; the sand is dumped in from an overhead hopper, jolted and then strickled off. A flat metal plate, preferably having a rubber seal at its edge, and containing a series of tubes connected to a carbon dioxide source is suspended above the molding box; the tubes are spaced at about four-inch intervals and they follow the contour of the pattern. The plate is lowered so that the tubes are imbedded in the sand and the plate is flush with the strickled surface. Application of carbon dioxide for about 12 to 15 seconds is sufficient for hardening and the two half molds are then tipped out by manual inversion of the box. The total time for the production of these two molds is between 50 and 70 seconds. Of incidental interest, is the technique of clamping a series of these book molds together which leads to a very considerable saving in both molding boxes and floor space.

(2) Hardening with Hollow Patterns. It was appreciated at a quite early stage that it would be preferable if gassing could take place at the sand/pattern interface; this would overcome the principal objection to the tube method where the gas, finding the path of least resistance to the atmosphere often bypasses pockets of sand, and should also enable increased mechanization of gas application and economies in gas consumption.

Where large number of molds are required, it is economic to make hollow metal patterns and space core-box vents at suitable positions over the working surface. Introduction of the gas into the hollow pattern then hardens the mold surface and gassing times of about 15 seconds are usually sufficient (Fig. 19).

Alternatively, a much more simple technique is to

adapt a wooden pattern by drilling approximately ¼-inch holes vertically down through the pattern, filling the holes on the pattern surface with core-box vents and fixing a gas chamber on to the base of the pattern (Fig. 19); such a chamber may be readily constructed by fixing about ½-1-inch square wooden battens to the bottom of the pattern along with a base board of metal or wood. This chamber should be sealed with rubber strip or other sealing medium so as to prevent gas leakage.

With such a technique, it is very easy to produce shell type molds but naturally of greater thickness than is the case with resin-bonded shells. Such molds can be backed up with green sand of low moisture content. An ejection bench, developed and used in the Billingham Foundry of Imperial Chemical Industries Limited, which incorporates gassing from the pattern surface, is shown in Fig. 19A.

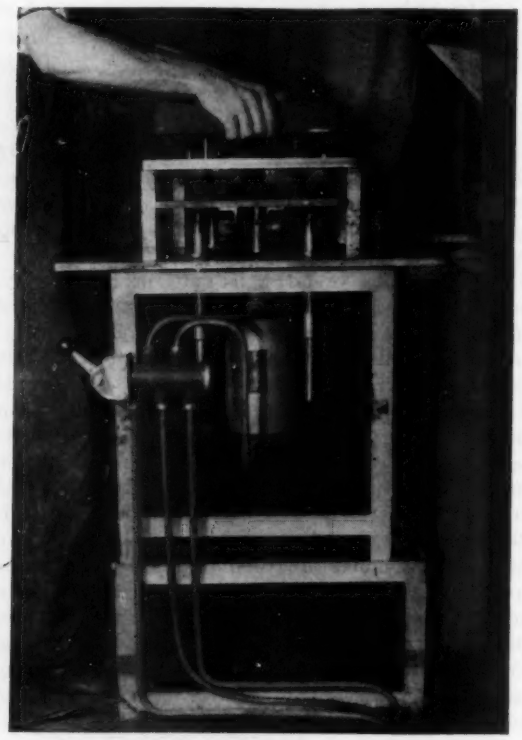


Fig. 19A — Ejection bench for CO_2 molds; gassing by means of vents in the pattern face.

(3) Hardening after Stripping Green. This technique is perhaps the most versatile method of mold production as it can be adapted immediately to existing pattern equipment, and can also be most economical in gas consumption. CO_2 binders have been developed that impart a moderate green strength to the sand, 3 to 4 psi being typical figures. In addition, these binders have a high sag resistance probably due to the rapid production of a hardened skin on the sand surface. Such binders enable the mold to be stripped green and this is essential with poor quality patterns or those with little taper and a deep draw.

After the pattern has been stripped green, a variety of simple techniques may be employed for hardening the mold. For example, as illustrated in Fig. 19, the mold cavity may be covered with a hood or board and gas injected on to it; if the board has a rubber-type seal to fit the periphery of the molding box, then the effectiveness and economies of the method are considerably enhanced.

Alternatively, both cope and drag may be closed together while still green and gas injected into the mold cavity down the sprue (Fig. 20).

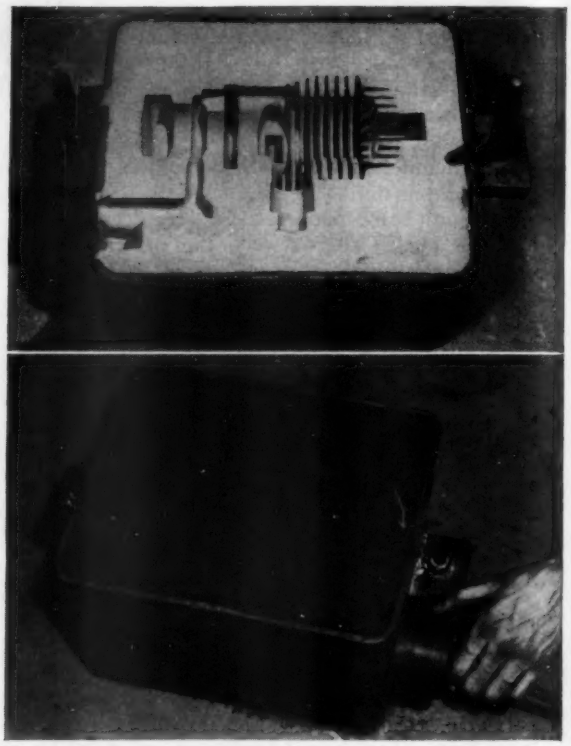


Fig. 20 — Gassing by gas injection down the sprue into mold cavity.

(4) Evacuation Technique. Due to the difficulty with the tube technique of gassing all parts of a mold readily, it was considered by ourselves and also several foundrymen, in particular Mr. Jack Bonney of Bonney-Floyd Foundry, Columbus, Ohio, that it might be possible to reduce the air pressure in a mold or core by evacuation and then harden rapidly and uniformly by application of carbon dioxide. Although this technique might appear a little complex, a prototype piece of equipment has been developed that is functioning satisfactorily for the production of small book molds.

This prototype is shown in Fig. 19; essentially, the half mold is stripped green and turned out on to a perforated plate which is then transferred on to a wire grid. A metal cover is brought down over the mold so as to form a vacuum chamber. This chamber is connected to a large reservoir which is continually evacuated so that, by opening a valve, the pressure in the vacuum chamber can be instantaneously reduced to a low value. The chamber is then sealed from the

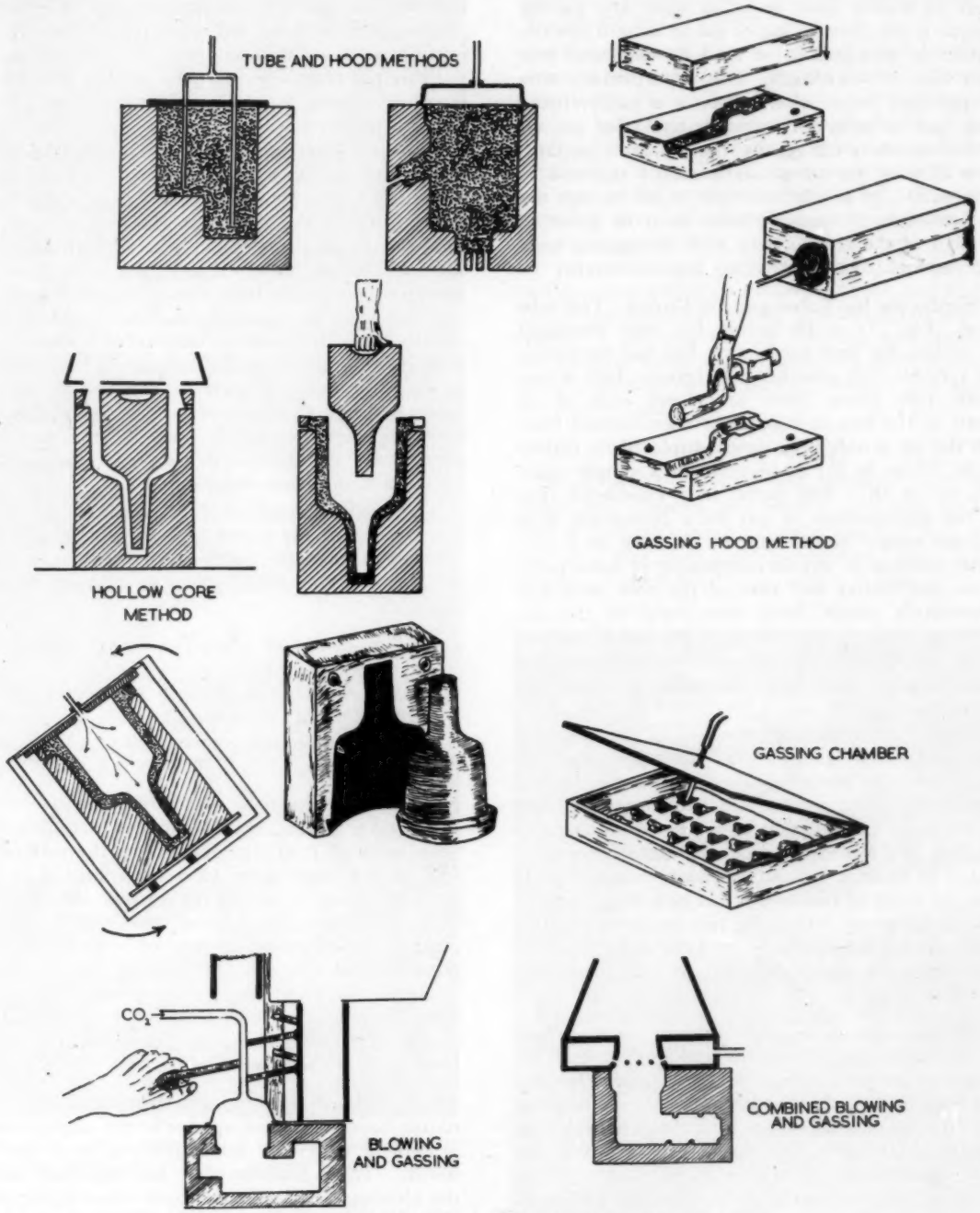


Fig. 21 — Typical methods of core production by CO_2 process.

vacuum reservoir and connected to a carbon-dioxide reservoir. Hardening is extremely rapid and very uniform and leads to the production of higher strengths than are obtained with other gassing methods.

(5) **CO_2 Shell Molds.** A method that might be capable of very great development has been suggested by my colleague, Mr. D. H. Snelson. Similar to the D-process, a CO_2 sand is blown into a cavity forming a shell over the pattern plate. This shell may be hardened by such methods as the simultaneous introduction of gas into the sand/air stream as it enters the mold (see later) or by using a hollow pattern. Such a method, if successful, would enable shells to be produced rapidly without the necessity for heated pattern equipment and with very low binder costs. As will be shown in the section on foundry work, the surface finish obtained with fine CO_2 sands is directly comparable to that obtained by orthodox shell molding.

Production of Cores. As with molds, the main reason for soft or friable cores resulting from any gassing technique is the channelling of gas in certain specific directions so that pockets of sand do not come into contact with carbon dioxide. This is particularly true with split core boxes where there is a badly fitting parting line or with coreboxes designed for use on core blowers where the venting requirements for satisfactory blowing are often diametrically opposed to those required for an efficient flow of gas through the core. Consequently, consideration must be given to the design of the box so that, with the gassing technique employed, gas channelling does not occur.

(1) Hardening by Tubes and by Covers. The tube method (Fig. 21), as for molds, has been employed quite widely for core production but has frequently given variable and unsatisfactory results. This is particularly true where there are closed ends or reentrants in the box so that pockets are formed from which the air is only very slowly displaced by carbon dioxide. Thus, in the production of a simple cylindrical core with a side barrel in a closed-end split box, the introduction of gas by a cylindrical tube down the center of the core for as long as 3 to 4 minutes resulted in serious overgassing of some parts, but the side barrel and nose of the core were still comparatively green. Such were some of the disheartening results experienced in the early days and these detracted from the value of the process until it was realized that they were due solely to faulty gas application.

By cutting vents through the box to the nose and to the side barrel, a clear path for gas through all parts of the core was established and it was decided to apply the gas by means of a rubber gasket, fixed over the print face, as shown in Fig. 21. Using this technique, the hardening time on this core was reduced to 20 seconds. In addition, cores that previously had to be made in two halves can now frequently be made in one piece; either the two halves of the box can be rammed up separately, sprigged and jointed together, and then gassed (Fig. 21), or a unit box may be used.

(2) Hollow core production. Certain cores, such as cylinder blocks, clutch housings, and gear boxes in the automobile industry, are particularly suitable for production as hollow cores since they have a comparatively large print area/weight ratio. Such hollow cores are most advantageous for easy breakdown and also enable considerable economies to be made in unit core costs. It was soon realized that the carbon-dioxide process is eminently suitable for hollow core production and the process is being used to an increasing extent for this purpose on a production scale.

A typical technique is as follows: a plug of metal defining the hollow portion of the core is attached to the blow plate of a vertical core blower. The dimensions of the simple shape plug are such that there is a half-inch annular space between the plug and the inside of the box and the blow holes in the blow plate naturally correspond with this space. On blowing, this plug defines the wall thickness of the hollow core as half-inch which was found by practical tests to be satisfactory for strength, ease of handling and resisting premature breakdown during casting. Naturally, it is essential that the gassing operation takes place after the core has been removed from the box and the box from the blower, as otherwise there will be a bottleneck on production.

Since any core must, at some stage during its production, be turned over on to a core plate, no time is expended if gassing takes place during the roll-over stage. Instead of a normal perforated core plate, a solid plate with a single hole situated centrally for the introduction of carbon-dioxide gas is used; with a stream of gas issuing from this tube, it is placed on to the core box containing the hollow core. During the roll-over, which normally only takes a few seconds, a crust of hardened sand is formed on the inside surface of the core sufficient to prevent the core collapsing; the core, in fact, acts as its own carrier. The core box is immediately stripped and returned to the core blower and the hollow core is left on top of the plate for completion of gassing.

The production of a hollow gear-box core is illustrated later in the paper but the following figures, bearing in mind the lower cost of CO_2 mixes, serve to illustrate the great economic advantages of the method.

Core A: Solid oil core—19.1-lb wt; hollow CO_2 core 12.9-lb wt.

Core B: Solid oil core—44-lb wt; hollow oil core 29-lb wt, which is made in two halves and then joined; hollow CO_2 core 9-lb wt.

(3) Gassing Chamber. Although this method has very great appeal, very little production use has been made of it so far. Essentially, the method consists of inserting green CO_2 cores into a chamber through which a stream of carbon dioxide flows. With large cores, the gassing times will probably be considerably in excess of those of other methods and there may be danger of overgassing the core surface. However, for small cores, gassing times are approximately the same as those with other methods. Figure 21 shows a typical improvized chamber that is being used on a production scale.

(4) Combined blowing and gassing. Various techniques have been developed whereby carbon dioxide is applied to a core immediately after it has been blown. Thus, Saubermann[9] has described recently the blowing of hollow valve cores whereby the carbon dioxide is applied to the sand through the former immediately after blowing. Another method used on a cartridge-type blower incorporates a gassing cup so that the core may be gassed, using a pressure of about 40 to 60 psi, while the cartridge is being filled with sand for the next core (Fig. 21).

However, the ideal is represented by the blowing of a core so that carbon dioxide is applied simultaneously with the admission of sand into the corebox. A prototype of such a blower has been developed by my colleague, Mr. D. H. Snelson (Fig. 21). The method is based on the use of a hollow orifice blow-

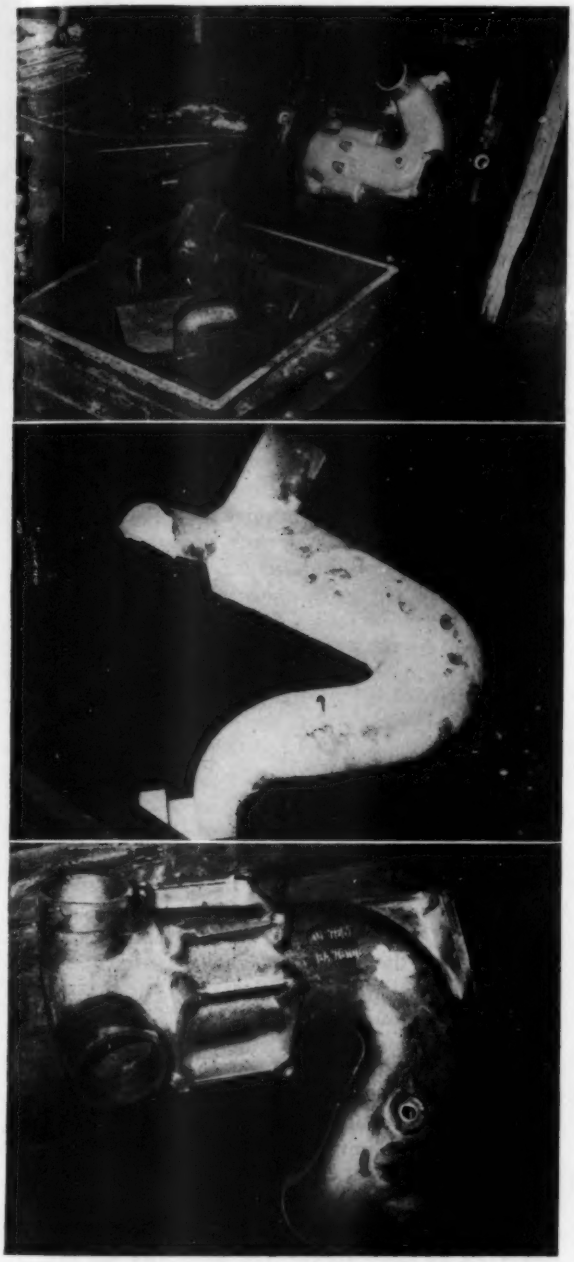


Fig. 22 (Top) — Mold and pattern for elbow pipe casting; (Center) Core assembled in mold; (Bottom) Finished casting.

plate which is shaped in the form of a venturi. Assuming that the air pressure used in the core blower is about 100 psi, carbon dioxide is injected into the blow plate at a pressure of about 70 to 80 psi and, on blowing, carbon dioxide is drawn into the sand air stream but does not penetrate into the sand reservoir or container in the blower due to the pressure differential. Naturally, this development will require considerably more experimental work but there can be little doubt that it will eventually constitute a very satisfactory production method.

There can be little doubt that the techniques employed for introducing the gas into the compacted sand will be the field of greatest development with the process in the future.

Gas Consumption. It is difficult to give a general estimate of the quantity of gas required for treating a given weight of sand as this naturally varies with the gassing technique employed. The quantity of gas actually absorbed by a correctly gassed core, with say a 4 per cent binder addition, is between 0.25 and 0.3 per cent of the weight of sand treated. These figures have not been achieved in practise, but the author knows of one foundry who have achieved a figure of 0.7 per cent, based upon the weight of sand mixed and the weight of gas purchased. On the very high side and using wasteful gassing techniques, figures have reached 5 per cent but the majority of foundries in this country using the process on a production scale are probably using somewhere between one and two per cent of gas.

In the Foundry

In this section, a brief description is given of the techniques adopted and results obtained in a few typical foundries employing the process. It is hoped that this will give more information and a clearer picture of the potentialities of the process.

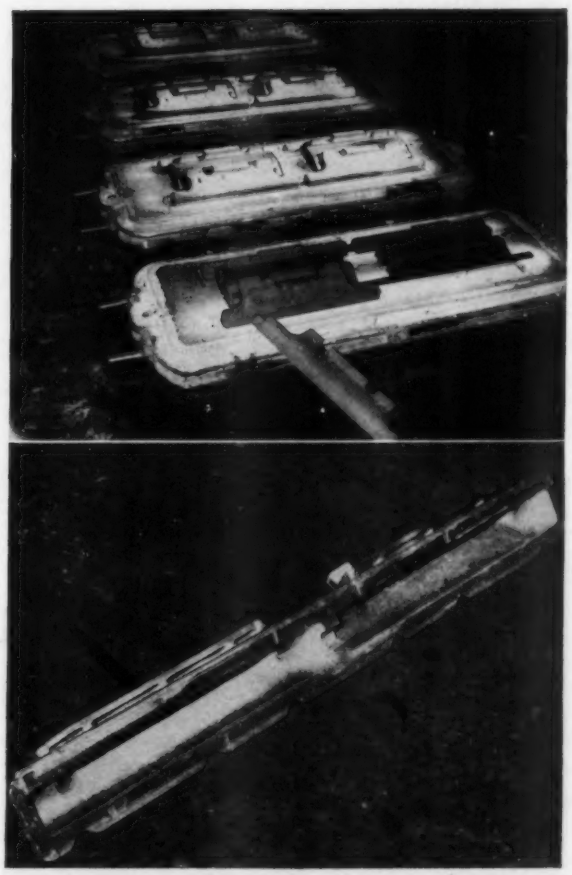


Fig. 23A (Top) — Core and mold assembly for automobile engine manifold. Fig. 23B (Bottom) — Finished manifold casting.

In light alloy foundries, the process has been rapidly adopted, particularly in gravity diecasting foundries. No major difficulties have been experienced with collapsibility and surface finish and many foundrymen have found these properties superior to normal sand mixes. Some difficulties have occurred on castings of very thin section due to blowing off the core and, in these cases, it has been necessary to stove the core; however, in general, the incidence of blowing defects has been reduced. Greater dimensional accuracy has been obtained, sufficient to stimulate praise from such casting consumers as an aircraft producer and an automobile manufacturer.

The appeal of the process to one particular foundry producing magnesium castings for the aircraft industry is the possibility of increasing accuracy so that weight tolerances of less than one per cent can be more readily obtained. The mixture previously employed was 4 per cent bentonite, 5 per cent flowers of sulphur and ½ per cent boric acid in a fineness number 67 sand; the clay binder was found advantageous in reducing hot tearing although it did give a poor quality core. With CO_2 sands, it is not possible to prevent metal/mold reaction in magnesium with normal inhibitors such as boric acid and ammonium bifluoride, since these materials are acidic and react with the alkaline binder immediately on introduction into the mix.

However, it has been established on a laboratory scale that mold reaction can be prevented by the use of inhibiting coatings on a sand mix containing no inhibitors or simply by the introduction of approximately five per cent of sulphur into the sand. Consequently, with the same base sand, the foundry quoted above now uses additions of 3½% proprietary silicate binder and 5 per cent sulphur. Cores of much better quality are obtained; the surface finish is stated to compare very favorably with conventional sand practice with no trace of mold reaction and the collapsibility is superior. Typical of such castings produced in this foundry are a sand-cast landing wheel hub of one-inch section and a gravity die-cast gear box of ¼-inch section in a magnesium/8 per cent aluminum alloy.

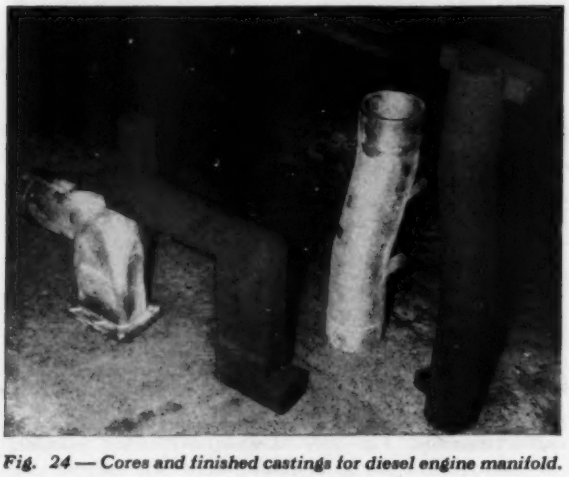


Fig. 24 — Cores and finished castings for diesel engine manifold.



Fig. 25 — Typical CO_2 cores for light alloy and copper castings.

At Birmingham Aluminum Castings (1903) Co. Ltd., who are one of the leading producers of a wide range of castings for the aircraft, automobile, domestic and atomic energy markets, the process has been employed to considerable advantage in both the gravity and sand foundries. In fact, on all prototype work, both molds and cores are produced by the CO_2 process, regardless of the ultimate method of production, in order to check the accuracy of pattern equipment.

The sand employed is one of fineness number 75 mixed with 3¼ per cent binder and ¼ to ½ per cent graphite; this mix is employed throughout and, although many castings are produced in alloys containing up to 10 per cent magnesium, no metal/mold reaction is experienced. Figure 22 shows the stages in the production of an elbow pipe casting. The pattern surface was lubricated with graphite to ease stripping and the mold was gassed with tubes; the core was sprayed with a proprietary white dressing.

Figure 23 shows the core assembly and finished casting of an automobile engine manifold. The cores and finished casting of a diesel engine manifold are shown in Figure 24. It is of interest that the cores shown were produced without strengtheners in a matter of a few minutes for this particular photograph.

A typical range of CO_2 cores used for both light alloy and copper castings by Diecastings Ltd. is shown in Fig. 25. In many copper-base foundries, the process is being employed extensively for core and also mold production. Carbonaceous additions are generally added to all sand mixes and good collapsibility is obtained; for normal sand mixes, coatings are generally necessary in order to obtain a good surface finish on castings of appreciable section. At the Billingham Foundry of Imperial Chemical Industries Ltd. a wide range of castings has been produced by the process and surface finish comparable to shell molding has been obtained by the use of fine sands.

Figure 26 shows an 88/10/2 gunmetal hopper for handling explosives where a surface finish of high standard was required; the molds and cores were produced in a sand of fineness number 162 with 6 per cent of binder and the finish obtained was ex-



Fig. 26 — Molds, cores and finished castings of gunmetal hopper.

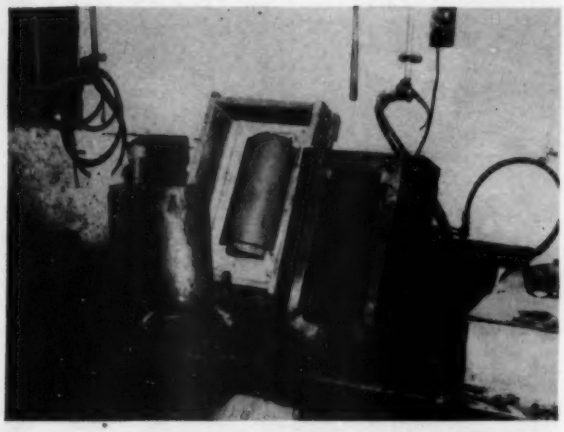


Fig. 27 — Pattern equipment, mold, core and finished casting of leaded bronze bearing.

cellent. Figure 27 shows the pattern equipment, mold, core and finished casting of a leaded-bronze bearing of 80-lb weight. This casting was produced at Western Foundries Ltd., London and measured 14 inches long, 7 inches external diameter and 1-inch section; both mold and core were blacked in the region of the ingate.

An aluminum bronze casting produced with CO_2 cores by Diecastings Ltd. for a Marconi scientific instrument is shown in Fig. 28 and illustrates the good surface finish obtainable without the use of coatings.

Perhaps the most extensive developments with the process have taken place in iron foundries and much of this progress has been described by Mr. A. Talbot, the foundry manager of Western Foundries, in numerous lectures to branch meetings of the Institute of British Foundrymen. In this foundry, after approximately 18 months' experience, the process is now being used for all cores and a high percentage of molds. The standard mix is 3½ to 4 per cent of binder, 2 per cent of coal dust in a fineness number 77 sand. This sand is employed on heavy- and thin-section iron and copper-base castings.

For light alloys, woodflour is substituted for coal dust. The collapsibility in all cases is found to be excellent and, where high standard surface finish is required, a proprietary inflammable blacking is used. It should be emphasized that a high standard surface finish is essential since no shot blasting equipment is yet available in this foundry. Normal patterns are used and boxes are rammed with a flat type rammer; no grids or plates are used for cores. Gassing is generally carried out with tubes or by means of hoods placed over the mold cavity after stripping green.

Mr. Talbot feels that the main advantages of the process are the continuity of accuracy obtained and the very reduced handling. In addition, the process is considerably cheaper than conventional methods as the following figures quoted by Mr. Talbot indicate; for both large and medium cores, the costs, exclusive of sand and labor, are now 50/- per ton ($7.00) compared to 84/- per ton ($11.75) before the advent of the CO_2 process.

The following photographs give some indication of the wide range of work undertaken at Western Foundries. Figure 29 shows the core for an oil-well pump gearbox casing; this is produced in two halves, the base weighing 12½ cwts (Fig. 30) and the top 8-cwts. An inflammable blacking was used on both molds and cores and an excellent finish was obtained. Figures 31 and 32 show the core box equipment and core for a lathe bed plate. Previously, this core had to be made in seven pieces and was heavily sprigged; with the CO_2 process only two straight rods are used down the length of the unit core.

In the foundry of R. J. Hunt & Son Ltd., Birmingham, England, who produce castings for the auto-

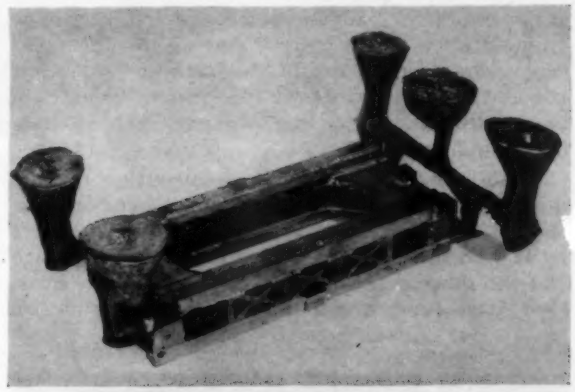


Fig. 28 — Aluminum bronze casting.



Fig. 29 — Core for oil-well pump gearbox.

mobile industry, the application of the process has been confined to their mechanized core shop where they have thoroughly investigated the use of the process on a production scale for one particular gearbox core; the results with this core have been so encouraging that two more cores are shortly to be turned over to the process.

The method used for producing the gearbox core is illustrated in Fig. 33. The method may be described as the hollow-core method and is particularly suitable for cores with a large area of print to mass of sand such as occurs with cylinder blocks, clutch housings and gearboxes. The advantages of such hollow cores, apart from extreme ease of gassing, are excellent collapsibility and a considerable reduction in the amount of sand used per core.

Using this technique, a serious bottleneck on core stoves has been eliminated, these cores being used at the rate of 240 per day. In addition, there has been a reduction in core distortion with a reduced amount of rejects. The sand mix is fineness number 73 sand with 3 to 4 per cent binder and 1 per cent pelletted pitch; the good finish obtained is illustrated in Fig. 33 (h).

This particular core weighed 58 lb in oil sand whereas the CO_2 hollow core weighed only 39 lb which is equivalent to a saving of nearly 60 per cent in core cost alone at an oil sand price of ½d per lb ($.06 per lb) compared to a CO_2 sand cost of 0.34d per lb ($.04 per lb).

In the large jobbing and semi-mechanized meehanite foundry of Ashmore, Benson, Pease and Co., considerable strides have been made with the process and, at the end of 1955, approximately one year after their initial trials with the process, they are producing 90 per cent of their cores and a substantial proportion of molds by the process. At the present time, only limited trials have been carried out in their mechanized foundry.

In this jobbing foundry it is interesting to note the rapid decline in the use of core stoves; of four large mold stoves, approximately 16 feet by 24 feet by 12 feet, three are now closed and used for storage and the other copes with dry sand and loam cores. One Acme stove is still used but a battery of five stoves, measuring 5 feet by 6 feet by 6 feet, are completely closed.

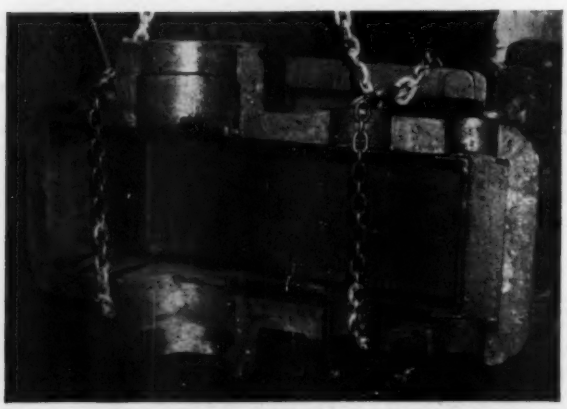


Fig. 30 — Base of oil-well pump gearbox.

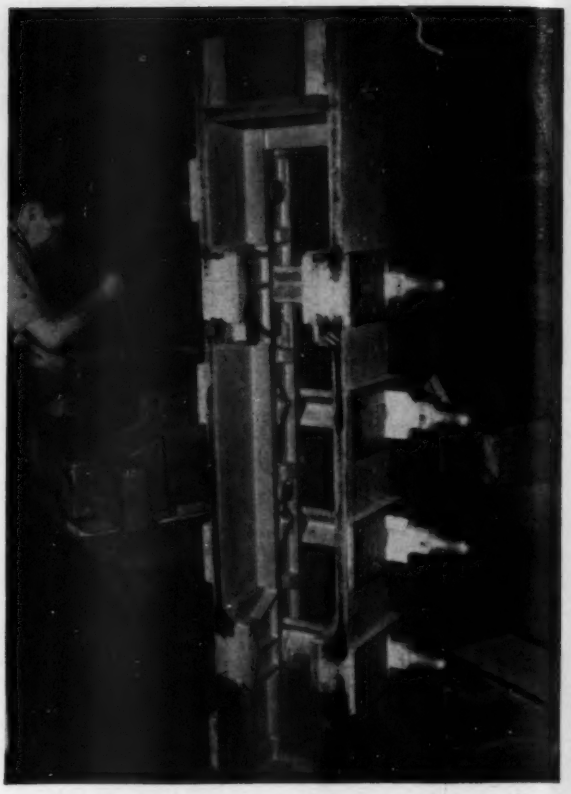


Fig. 31 — Corebox equipment for a lathe bed plate.

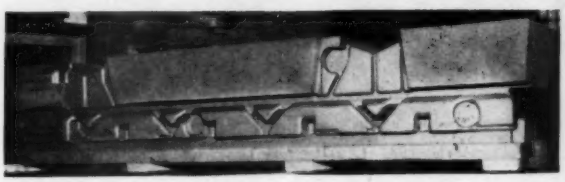


Fig. 32 — Core for lathe bed plate.

The carbon dioxide is at the moment supplied from four liquefiers of 100-lb capacity and the gas is piped the full length of the foundry so as to be available at numerous points; a static bulk liquid tank of 5-tons capacity is being installed at the moment. The sand is prepared in 675-lb batches and the following unit sand is employed; fineness number 75 sand with 3½ per cent of binder and 1 per cent pelletted pitch with the addition of 1 per cent ball clay for those mixes requiring green strength. The methods of gassing molds and cores are generally those of stripping green and gassing with a sealed board over the mold cavity and of manifold tubes respectively. Cores and molds are coated with an inflammable graphite dressing where necessary.

Typical jobs produced in this foundry are illustrated by the following photographs. Figure 34 and Fig. 35 show the cores and molds for a diesel cylinder frame of finished weight 4.5 tons. Figure 36

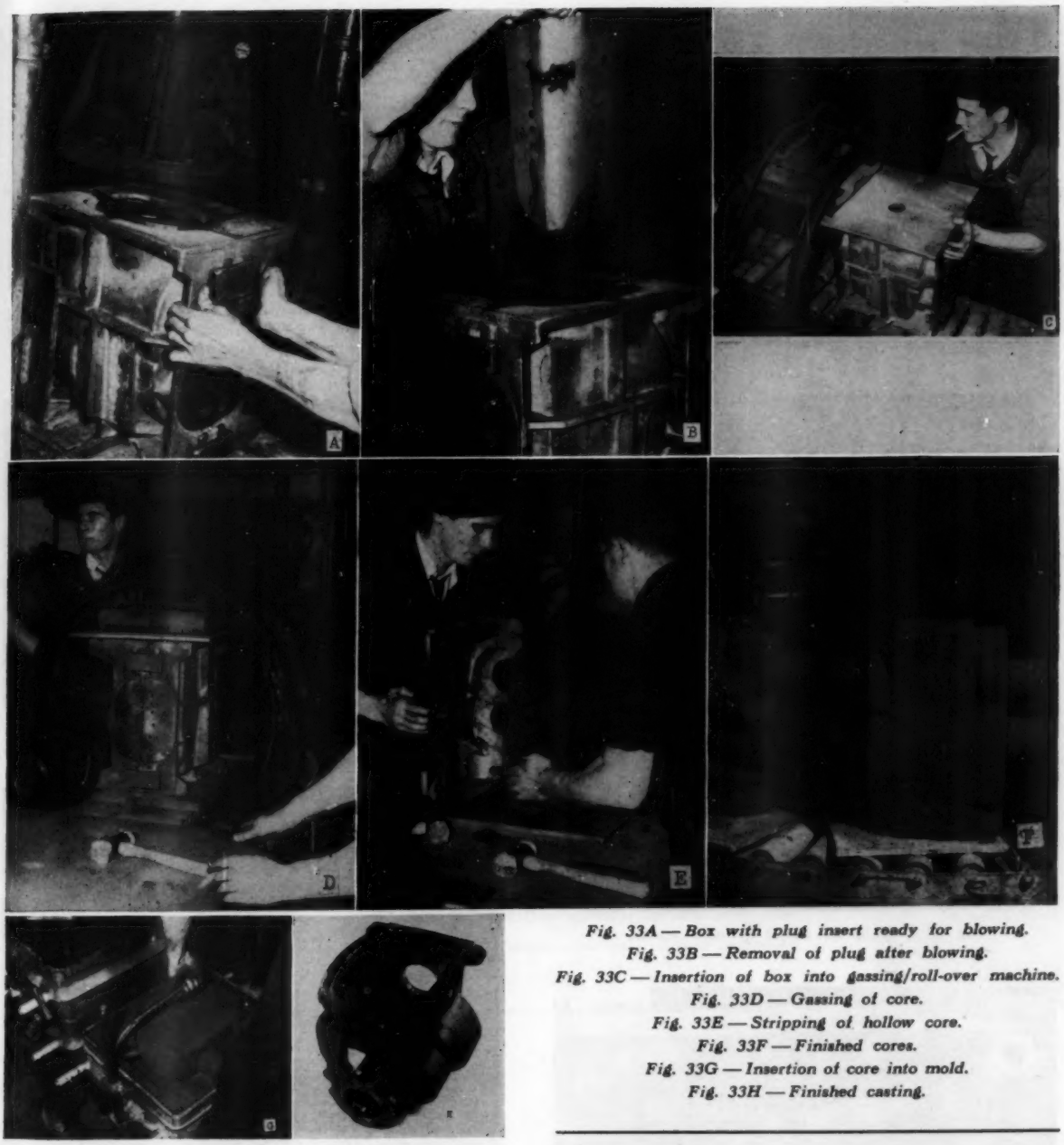


Fig. 33A — Box with plug insert ready for blowing.
Fig. 33B — Removal of plug after blowing.
Fig. 33C — Insertion of box into gassing/roll-over machine.
Fig. 33D — Gassing of core.
Fig. 33E — Stripping of hollow core.
Fig. 33F — Finished cores.
Fig. 33G — Insertion of core into mold.
Fig. 33H — Finished casting.

shows a similar diesel frame of 5 tons weight where both cores and molds were produced by the process.

Figure 37 shows the stages in the production of a flywheel weighing just over 600 lb which was previously made in dry sand; this is produced on a roll-over machine with a facing of CO_2 sand over the pattern backed up by green sand.

Figure 38 shows an intricate core for a tube support casting for the oil refinery industry which was previously made in oil sand.

Finally, the base plate for a water gas generator is shown in Figure 39; this casting is produced in a core assembly of diameter 16 ft, 8 in. and weighs 5 tons.

At William Cross and Son Ltd., Birmingham, who produce mainly castings for the electrical industry, the process has eliminated a serious bottleneck on stove capacity for large cores. Figure 40 illustrates the various stages in the production of a pillar for a weighing machine using a sand of fineness number 75 with 4 per cent binder and ½ per cent pelletted pitch; the good finish obtained is illustrated by the photograph of the finished casting.

Figure 41 shows an attachment for blowing and gassing cores that has been developed in this foundry; a pressure of 40 lb of carbon dioxide is used in this set up and a gassing time of 3 seconds for the small mincer core shown in the photograph.



Fig. 34 — Cores for diesel cylinder frame.



Fig. 35 — Mold for diesel cylinder frame.

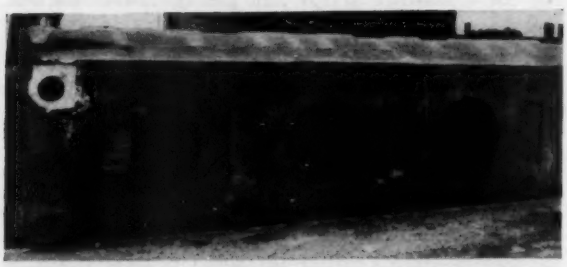


Fig. 36 — Diesel engine cylinder frame weighing 5 tons.



Fig. 37 — Stages in production of a 600-lb. flywheel showing CO_2 sand facing on pattern, gassing with cover over mold cavity, finished mold.



Fig. 38 — Intricate core for tube support casting.



Fig. 39 — Core for base plate of water gas generator.

The gassing operation is carried out while filling the cartridge for the next core.

The process has been very extensively employed in Glanmor Foundry and Thomas and Clement foundry by Mr. Ivor Rees for steel and iron castings. There can be little doubt that the foresight and ability of Mr. Rees have resulted in much of the progress that has taken place in Great Britain, with the carbon-dioxide process. Steel castings ranging in weight from several pounds up to many tons have been produced and Glanmor foundry have probably established a record by producing a 5-ton CO_2 core which was used for a slag shute. The sand mix employed is based upon a local sand of fineness number 60 and usually 4 per cent of the binder is employed. The finish obtained on heavy castings without the use of coatings is of exceptional standard but an inflammable zircon dressing is used for certain applications. Figure 42 shows the strip of sand occurring from a steel spiked roll crusher weighing 600 lb on knockout; this type of strip is characteristic of the CO_2 process on steel castings.

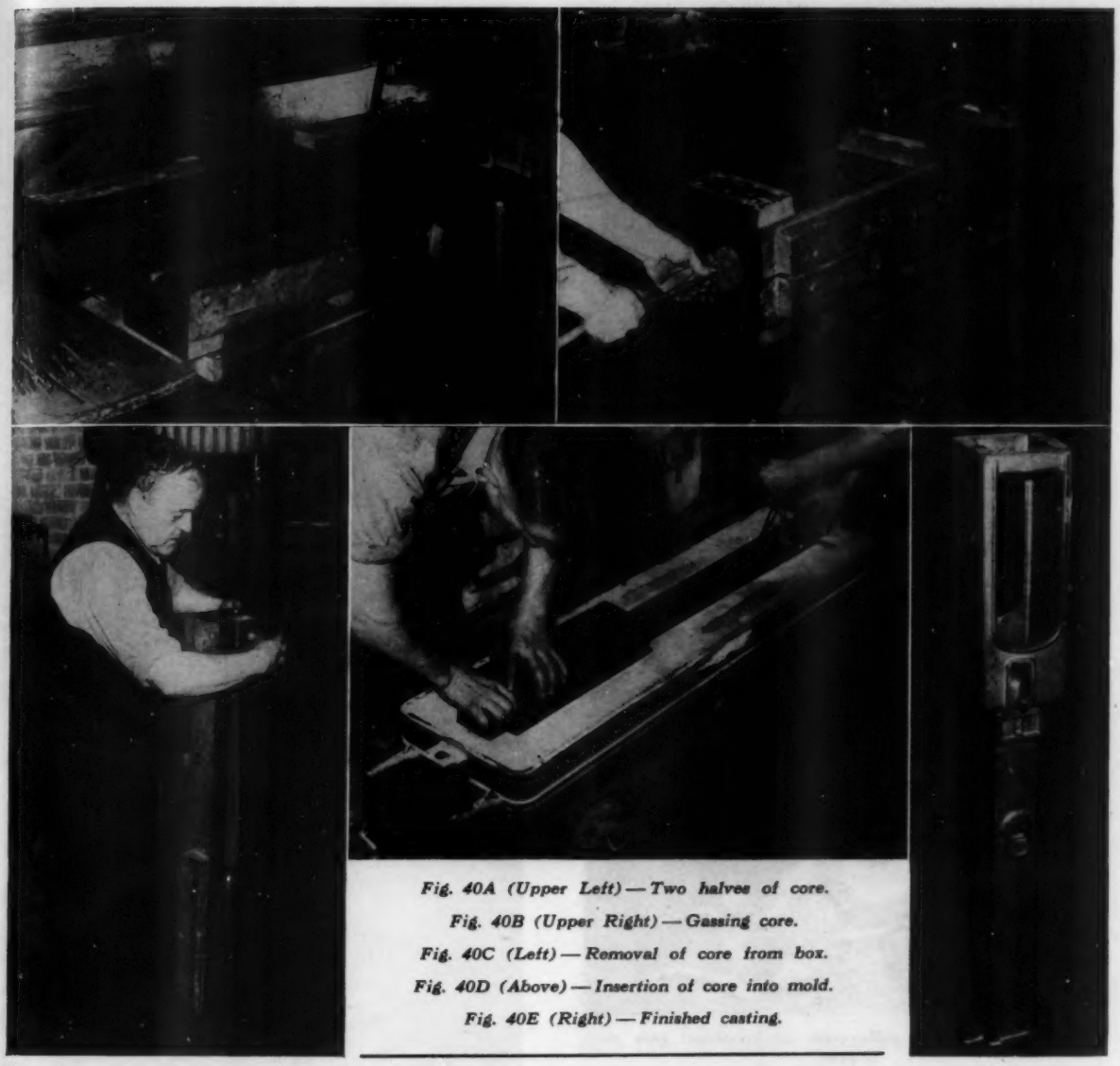


Fig. 40A (Upper Left) — Two halves of core.

Fig. 40B (Upper Right) — Gassing core.

Fig. 40C (Left) — Removal of core from box.

Fig. 40D (Above) — Insertion of core into mold.

Fig. 40E (Right) — Finished casting.

Figure 43 shows a double plated steel flywheel weighing 6 tons after fettling and this casting was produced in 36 hours from receipt of drawing to pouring the casting.

A recent casting produced at Glanmor foundry is a cast-iron launder holder (Fig. 44) for a blast furnace weighing 16 tons 700 lb fettled. In this case, the pattern was made and the job cast within 14 days of the receipt of the drawing. The mold was produced by making a flat bed in CO_2 sand on a bed of ashes and the internal core was struck up with a contoured pattern on the gassed bed. The ten-side cores (Fig. 45) and the five-cover core were all prepared in CO_2 sand and all were dressed with a graphite spirit blacking. The very good strip obtained is shown by the layers of sand (Fig. 46) that peeled cleanly away from the casting surface.

In the stainless steel foundry of the A.P.V. Paramount Ltd., the process is being used on a considerable scale for the production of molds and, to a lesser extent, cores. Many castings that were previously made in green sand are now produced by the CO_2 process due primarily to advantages in quality and accuracy and in freedom from slag and sand inclusions.

The base mix is a mixture of two parts and one part of sands of AFS fineness numbers 76 and 67 respectively with 5 per cent of the proprietary green strength binder referred to earlier.

The majority of molds are gassed by injection of carbon dioxide from the pattern face, similar in principle to the hollow pattern technique. Figure 47 shows the various stages in the production of book molds on a vibrator/roll-over machine for stainless steel valve castings. These book molds are clamped between fixed partitions on the foundry floor or on mobile racks (Fig. 48). Typical castings produced at this foundry are illustrated in Fig. 49.

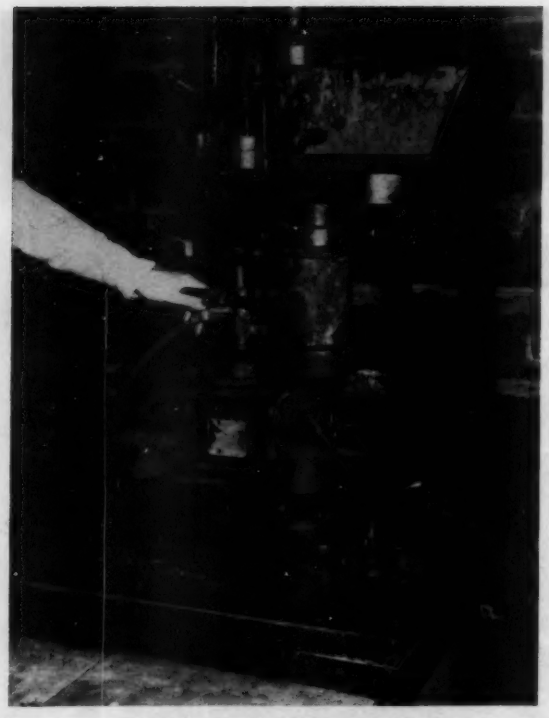


Fig. 41 — Core-blower attachment for gassing cores in situ.

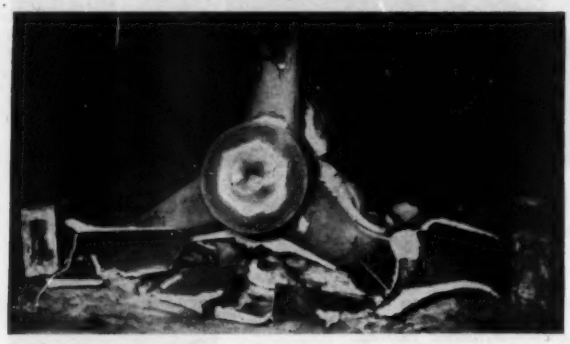
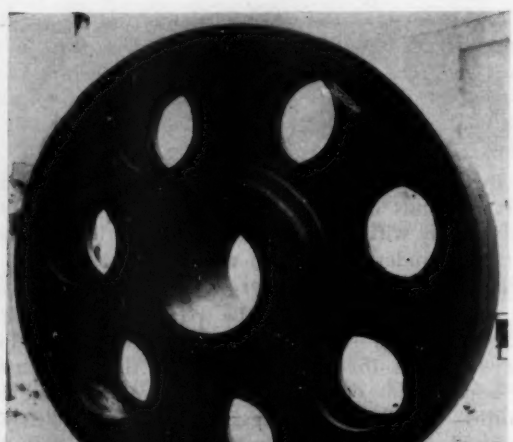


Fig. 42 — Steel spiked roll-crusher on knockout; note strip of sand.



(Left)
Fig. 43 — Double plated steel flywheel weighing 6 tons.



Fig. 44 — Cast iron launder holder weighing 16 tons, 700 lbs.

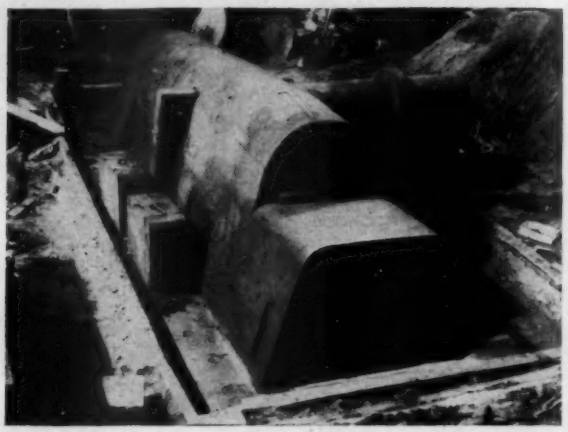


Fig. 45 — Strickled internal core for launder holder.



Fig. 46 (Right) — Layer of clean peeling sand removed from launder holder.

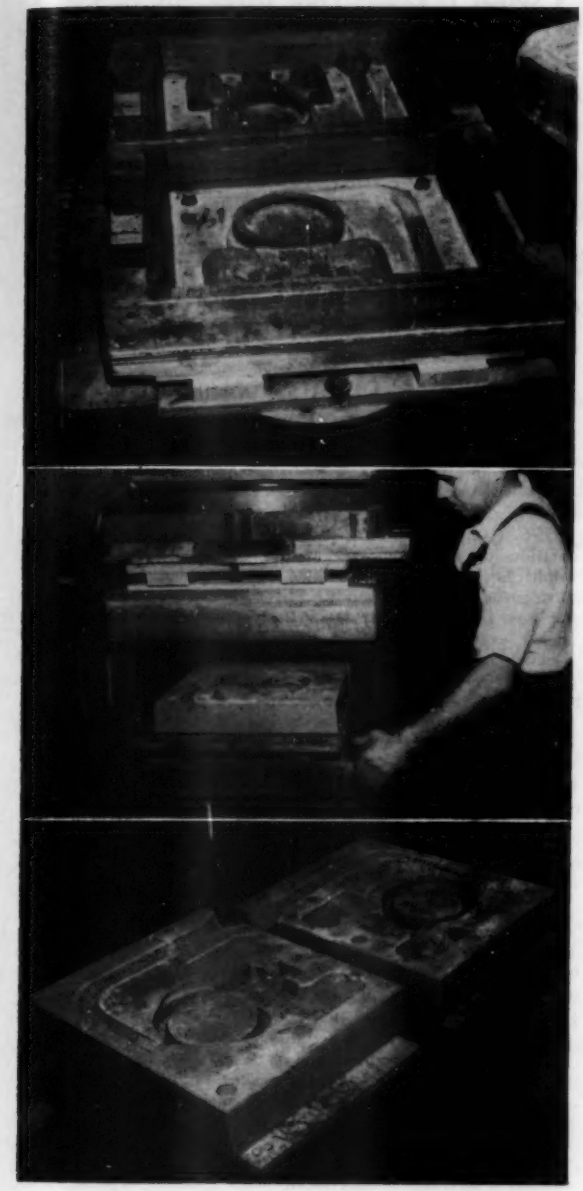


Fig. 47A (Top) — Double pattern for stainless steel valve on vibrator-roll-over machine; Fig. 47B (Center) — Two half molds after roll-over; Fig. 47C (Bottom) — Two half molds.

Sand Reclamation

Since the major portion of the binder material is water soluble, it was considered in the early stages of the process that reclamation of used sand by a simple washing process should be a production possibility. Preliminary trials on a laboratory scale showed that all of the residual binder material is removed by boiling water and 80 to 85 per cent by cold water treatment.

A pilot plant treatment of four tons of CO_2 sand that had been employed for steel casting production was carried out in a "log washing" plant by kind permission of the Research Laboratories of the United Steel Companies Ltd., Rotherham and in collaboration with the British Steel Casting Research Association. The log-washer is based upon improvised equipment employed by gold-miners for removing gold dust from sand and clay and consists essentially of a helical bladed shaft mounted parallel to the base of a concrete-lined trough which is inclined at an angle of 1 to 12 to the horizontal. Four rows of blades are welded to the shaft at an angle of 70 degrees and it is driven at a rate of 10 to 14 rpm with an hourly through-put of about 1.5 to 2.0 tons on this experimental plant. Water is supplied from a spray pipe fitted longitudinally above the trough and the water level is controlled by a weir at the end of the trough.



Fig. 48 — Assembled book molds.



Fig. 49 — Typical stainless steel castings.

The used CO_2 sand was initially fed through a jaw crusher and was charged into the lower end of the trough by a conveyor belt. The sand is driven up the trough by the rotating shaft and, after being exposed to the water spray and a grinding action, it is discharged at the upper end. The sand on discharge is naturally very wet and contained approximately 10 per cent of water.

Samples for analysis were taken from the following points:

a) Before washing.
b) Half way along the trough.
c) After washing.

The results obtained are given in Table 4.

TABLE 4

	% Before Washing	% During Washing	% After Washing
Total Na as Na_2O	0.540	0.221	0.203
Water soluble alkali as Na_2O	0.437	0.118	0.100
Water insoluble Na_2O	0.103	0.103	0.103
Water soluble CO_2	0.218	0.050	0.041
Na_2CO_3	0.525	0.120	0.099

The total sodium oxide content of the sand is of the same order as the theoretical concentration obtained by a 4 per cent addition of the binder. Of the water soluble alkali present, nearly 80 per cent has been removed by the log-washer. An even higher efficiency could have been obtained by reducing the distance between the blades and the trough so that small lumps of sand are completely broken up and thoroughly washed.

The physical properties of the washed and dried sand when bonded with the normal addition of binder were found to compare very favorably with new sand. However, this aspect of the process has attracted very little attention to date and, to the best of the author's knowledge, there is only one foundry reclaiming sand by wet methods on a production scale.

Acknowledgments

The author would like to acknowledge the invaluable assistance of the many foundrymen who have so generously assisted in the preparation of this paper by making freely available details of the techniques employed in their particular foundries. In particular, he would like to thank the Directors and staff of Birmingham Aluminium Castings (1903) Co. Ltd., Diecastings Ltd., Imperial Chemical Industries Ltd., Western Foundries Ltd., R. J. Hunt & Son Ltd., Ashmore Benson, Pease & Co., William Cross and Son Ltd., Glanmor Foundry Co. Ltd., A.P.V. Paramount Ltd.; also the Carbon Dioxide Co. Ltd. In addition, he would like to thank the Directors of Foundry Services Ltd. for their constant encouragement and his many colleagues there, particularly Mr. F. Pursall and Mr. D. H. Snelson, who have been actively engaged in the development of the process.

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DISCUSSION

Chairman: DOUGLAS JAMES, Urick Foundry, Erie, Pa.

Co-Chairman: L. J. PEDICINI, General Motors Corp., Detroit, Mich.

Secretary: WILLIAM RUPRECHT, ACF Industries, Inc., Berwick, Pa.

D. C. CAUDRON[1]: This is a very excellent and thorough presentation of the CO_2 process. We find no statements that would be contrary to our own experience with the process during the past eight months. In fact, the graphs and tables are very helpful in explaining and verifying our own results. As an example, "shelf-life" of an occasional extra core has varied considerably and has been puzzling. The information given on the effects of excess moisture in the sand-binder mix before gassing, and the effect of over-gassing, readily explains these apparent anomalies. We have sample cores made last November and stored away that are in perfectly usable shape today. These have been through a six-weeks period of almost continuous rainy weather with very high humidity day after day.

As regards additions to the sand-binder mix, we have found iron oxide up to 1 per cent to be helpful in facilitating handling of the cores, as well as improving the casting-finish. Addition of up to 1 per cent woodflour is our standard practice as an aid to shake-out. When adding woodflour we also add from ½ to an equal weight of water. It seems that woodflour removes water from the binder, shortening its bench life and generally weakening the resulting core, if compensation water is not added.

It is our experience that veining does not occur when using CO_2 cores (with or without iron oxide). This we have proved to our own satisfaction on an 11½-ft diameter butterfly valve seat ring of 1600-lb. weight, and cross section of approximately 3x5 in., made of gun metal. Side by side, the CO_2 core showed no veining, while the resin-bonded core necessitated considerable chipping. This lack of veining is particalarly helpful in the case of double shrouded pump runners where any veining means more cleaning expense either for the foundry or customer.

As regards gassing time, we have found a solenoid valve in conjunction with an electric timer adjustable from 0-60 seconds to be essential to uniformity of gassing. This timer is actuated by a push-button at the end of the CO_2 hose.

1. Pacific Brass Foundry, San Francisco, Calif.

CAUSES AND EFFECTS OF GRAIN SIZE IN PEARLITIC GRAY CAST IRON

By

Michel Ferry* and Jean-Claude Margerie**

The outline of a microstructural grain, which is obvious in solid solutions, can be readily delineated in a eutectic or a normal eutectoid. This is also true of gray cast irons in which the two phases of the eutectic do not exist side by side according to a definite pattern. Due to this irregularity the "eutectic cells", the existence of which has been proven by Boyles,[1] can often be only indirectly proven to exist, either by the presence of a phosphide eutectic or of segregations which are revealed by the application of special etching reagents.

In a previous paper on the grain size of cast iron[2] the authors pointed out a practical metallographic technique as devised by Gevers,[3] an etch similar to the Stead-LeChatelier reagent reveals the eutectic grain; one may then measure the average size of this grain by counting the number of eutectic cells through which a straight line passes or an arc of 1 cm true length in the plane of the micrographic section. For practical purposes, this number varies from a few units up to about 60. The higher numbers correspond to the finer grains. The authors have shown that this number is inversely proportional to the average diameter of the cells.

Observations published by the authors in the preceding paper[2] have been applied to a number of castings with various phosphorus contents. Some simple geometric shapes, poured at different temperatures, have been studied. The authors have noted the changes of the grain size in its relation to the shear strength. This method has permitted them to show the influence of certain factors, and they could confirm the effect of the cooling rate, V/S, the volume-surface area ratio. The authors have also noted certain anomalies which they shall explain.

The authors have undertaken experiments with samples of crucible-melted pearlitic gray cast iron to outline and develop this method of grain size measurement. Their idea was to develop the main factors which have a noticeable influence on the dimensions of the eutectic cells, and to establish to what extent the usual mechanical properties depend thereon. The authors do not claim that they arrived at a final solution at this time but state that they have confirmed some of their earlier conclusions relative to the influence of the law of cooling from the pouring of the liquid metal into the mold up to its complete solidification. Chemical composition, and more especially the carbon content, has an influence on the "granulometric distribution" throughout the cast piece. Finally the authors have taken advantage of recent work on the mechanical properties of gray cast iron[4] in order to understand the relations as found—some of them for a long time—within the stated limits, the appearance of the fracture and the metallographic examination.

I. Law of Cooling vs Grain Size

Notwithstanding the complex nature of gray iron, especially because of the effect of various elements in the cast iron, the eutectic cell shows one of the fundamental properties of the crystalline grain per se. It develops in the specimen under study, starting from the center, with a certain rate; but contrary to what one knows from the majority of the crystals, e.g. the austenite, the rate of development of the eutectic cells of the gray cast iron is more or less the same in all directions, so that the cells are more or less spherical, at least during the beginning of growth. A paper by A. Hultgren, Y. Lindblom and E. Rudberg[5] verifies this. This paper also shows how this eutectic solidification is accompanied by characteristic temperature variations.

The laws governing the appearance and development of the crystal nuclei determine the grain size, and the authors will outline these laws for cast irons.

1. Combined Influence of Position of the Point Under Study in the Piece, and of the Cooling Rate

A. General Rules

A casting, as a rule, is not homogeneous as evidenced by examining the grain size. The law of cool-

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This is the Official French Exchange Paper to the American Foundrymen's Society.

EDITOR'S NOTE: The manuscript was written in French and translated therefrom.

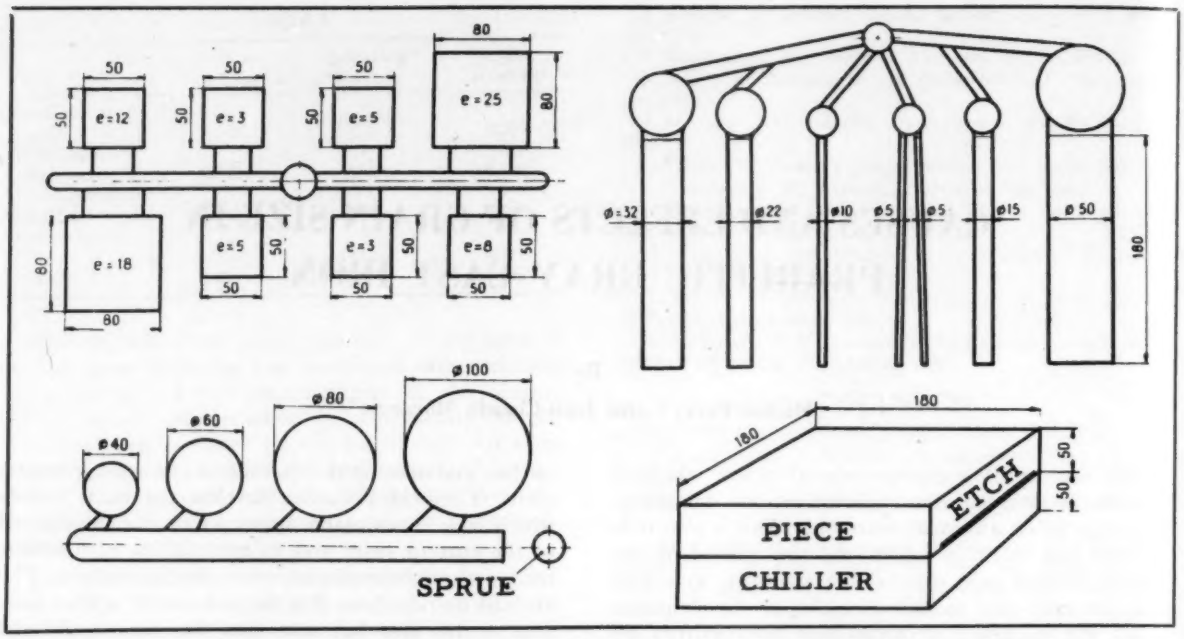


Fig. 1 — Test pieces of simple geometric form showing dimensions in mm and gating arrangement.

ing varies from one point to the other in the grain. The grain of a cast piece is finer the closer it is to the casting skin, or the thinner the section, i.e. the faster the solidification. This has been pointed out by the authors in a previous paper[2] when they worked on crucible-melted cast iron, always using the same raw materials in the charge. The cast irons were roughly eutectic (composition A) as follows: TC.. 3.38-3.48 per cent; Si, 2.60-2.76 per cent; Mn, 0.50-0.60 per cent; S, 0.05-0.06 per cent.

They also examined a definitely hypo-eutectic iron, by holding within the composition shown above except that Si was lower (composition B) as follows: T.C., 3.22-3.40 per cent; Si, 1.51-1.79 per cent; Mn, 0.56-0.62 per cent; S, 0.07-009 per cent.

The phosphorus content was varied from 0.10 to 0.60 per cent. In the series of compositions studied, these phosphorus analyses showed their influence on compositions A and B. This means that we can now disregard the other analyses and consider only the average results as if phosphorus were the only variable.

The authors poured a series of cylinders and plates according to a procedure described in their previous paper[2] (Fig. 1). The pouring temperatures were 1500 C (2730 F), 1400 C (2550 F) and 1300 C (2370 F). Test results of these castings are shown in Tables 1-4, as examples, because the findings on cast irons of different compositions have never contradicted the following rules:

1. Metallographic examination of castings of simple geometric form (plates, cylinders, spheres) will show that grain size in the center of these sections is independent of their shape, all other factors, such as cooling rate, analysis, and pouring conditions being constant.

In log-log coordinates, the average curve showing the variation of the number of cells per centimeter as a function of the cooling rate (V/S) slopes down. Figure 2 shows measurement results for the center of spheres, cylinders and plates poured at 1500 C

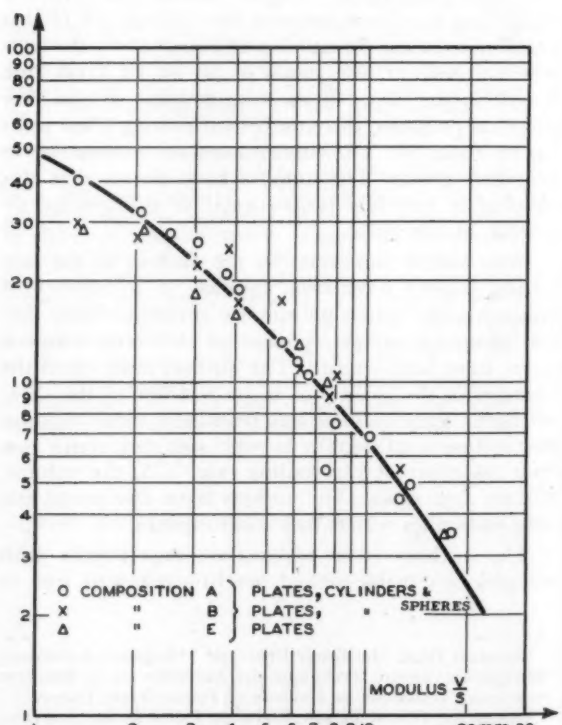


Fig. 2 — Relationship of number of cells per centimeter in the center as a function of the ratio V/S (poured at 1500 C or 2730 F).

Table 1 – Chemical Composition of Cast Iron Used in This Study on Influence of Cooling Conditions

No. of Casting	Pouring Temp., C	Composition, % T.C.	Si	Mn	S	P
		Composition A				
1285	1500	3.46	2.69	0.55	0.06	0.10
1286	—	3.42	2.76	0.55	0.05	0.20
1287	—	3.40	2.70	0.57	0.06	0.27
1288	—	3.38	2.66	0.54	0.06	0.42
1289	—	3.38	2.70	0.54	0.06	0.52
1415	1400	3.44	2.62	0.52	0.05	0.20
1416	1300	3.48	2.76	0.52	0.06	0.20
1417	1400	3.39	2.61	0.48	0.04	0.54
1418	1300	3.46	2.60	0.52	0.05	0.51
		Composition B				
1510	1500	3.40	1.69	0.62	0.08	0.14
1513	—	3.22	1.67	0.62	0.08	0.38
1528	—	3.47	1.79	0.61	0.07	0.56
1529	1400	3.36	1.69	0.62	0.07	0.26
1533	1300	3.27	1.60	0.58	0.06	0.26
1532	1400	3.25	1.59	0.59	0.08	0.59
1536	1300	3.39	1.51	0.56	0.09	0.60
		Composition E				
1821	1500	3.56	1.45	0.61	0.04	0.34

(2730 F) (Composition A, B and E).

2. Metallographic examination of cast samples of simple geometric forms shows that the number of cells or grains per centimeter decreases from the casting surface toward the center of the piece.

3. If one studies simultaneously the influence of the cooling rate and of the location of the point under investigation, one is led to trace a family of curves (Fig. 3) for each series of geometrically similar castings which have been poured under identical conditions from the same iron. As a rule these curves slope down and do not intersect.

4. These three above mentioned rules show exceptions which appear systematically.

Table 3 – Number of Cells Per Centimeter at Surface and Center of Pieces of Simple Shape Poured at 2550 F

Shape of Piece	Diameter, in.	Average Number on Surface Composition A	B	E	Average Number at Center Composition A	B	E
Plate	1/8	48	(32)	(35)	46	(33)	(38)
Plate	3/16	40	35	32	34.5	35	31
Plate	5/16	35	30	30	26	24	23
Plate	1/2	30	22	26	15.5	18.5	21
Plate	11/16	26.5	17	18	9.5	12	15
Plate	1	22	15.5	17	7.5	10.5	11
Cube	4	—	—	9	—	—	4
Cylinder	7/16	28.5	—	—	24.5	—	—
Cylinder	9/16	24.5	28	—	17	23	—
Cylinder	7/8	26	28	—	12.5	16.5	—
Cylinder	1/4	23.5	21	—	10	11	—
Cylinder	2	19	16	—	6	6.5	—
Sphere	19/16	30	—	—	12	—	—
Sphere	23/8	25	—	—	7	—	—
Sphere	31/8	22.5	—	—	7	—	—
Sphere	4	18.5	—	—	5	—	—

(For Depth of Chill See Table 2)

Table 2 – Number of Cells Per Centimeter at Surface and Center of Pieces of Simple Form Poured at 2730 F

Shape of Piece	Diameter in.	Average Number on Surface Composition A	B	E	Average Number at Center Composition A	B	E	Depth of Chill, in.
Plate	1/8	40	(28)	(28)	40	(29.5)	(28)	1/16
Plate	3/16	35	26.5	30	32	27.5	28	3/32
Plate	5/16	30.5	25	22	26	22	18	1/8
Plate	1/2	23.5	20	18	18.5	17.5	16	3/16
Plate	11/16	18.5	15	15	11.5	11	13	1/4
Plate	1	15.5	14	13	5.5	9	10	5/16
Cube	4	—	—	7	—	—	3.5	21/32
Cylinder	7/16	36.5	—	—	27.5	—	—	3/32
Cylinder	9/16	30.5	26.5	—	21	25	—	5/32
Cylinder	7/8	23.5	23	—	13	16.5	—	7/32
Cylinder	1/4	20.5	18	—	7.5	10.5	—	5/16
Cylinder	2	16.5	14	—	4.5	5.5	—	1/2
Sphere	19/16	25.5	—	—	11	—	—	9/32
Sphere	23/8	17.5	—	—	7	—	—	13/32
Sphere	31/8	17	—	—	5	—	—	17/32
Sphere	4	12.5	—	—	3.5	—	—	21/32

B. Exceptions to the General Rules

In the specimens the authors studied, they have found two exceptions to these general rules, both of which are equally interesting.

1. First of all they have encountered, distributed mainly in the center of the pieces, discontinuous areas, which are well outlined by concave contours where the grain is much finer in the area, or where on the contrary its size is normal as would be expected. If the authors in their grain size measurements did not exclude these areas, the figures obtained would have been up to 10 times larger (Fig. 4).

Their positions and distribution suggest that one is dealing with areas which were last to solidify. Their

Table 4 – Number of Cells Per Centimeter at Surface and Center of Pieces of Simple Shape Poured at 2370 F

Shape of Piece	Diameter, in.	Average Number on Surface Composition A	B	E	Average Number at Center Composition A	B	E
Plate	1/8	60	—	(24)	60	—	(24)
Plate	3/16	52	(31)	36	48	(31)	36
Plate	5/16	50	35	33	29.5	29	28
Plate	1/2	49	30	29	22	22	21
Plate	11/16	42	26	23	14.5	17	14
Plate	1	39.5	20	20	12.5	12	11
Cube	4	—	—	12	—	—	4.5
Cylinder	7/16	(31.5)	—	—	(32)	—	—
Cylinder	9/16	29.5	(20)	—	20	(26)	—
Cylinder	7/8	32	27	—	18	19	—
Cylinder	1/4	34.5	25	—	12.5	14.5	—
Cylinder	2	28	21	—	8	10	—
Sphere	19/16	39.5	—	—	14.5	—	—
Sphere	23/8	38.5	—	—	11.5	—	—
Sphere	31/8	29	—	—	9.5	—	—
Sphere	4	28	—	—	6	—	—

(For Depth of Chill See Table 2)

microstructural appearance shows that the cast iron has undergone a noticeable undercooling in these areas (fine graphite distributed around wedged-in dendrites). Finally in similar but much smaller pieces, the regions of a similar location are made up of white cast iron with a very fine microstructure (Fig. 5).

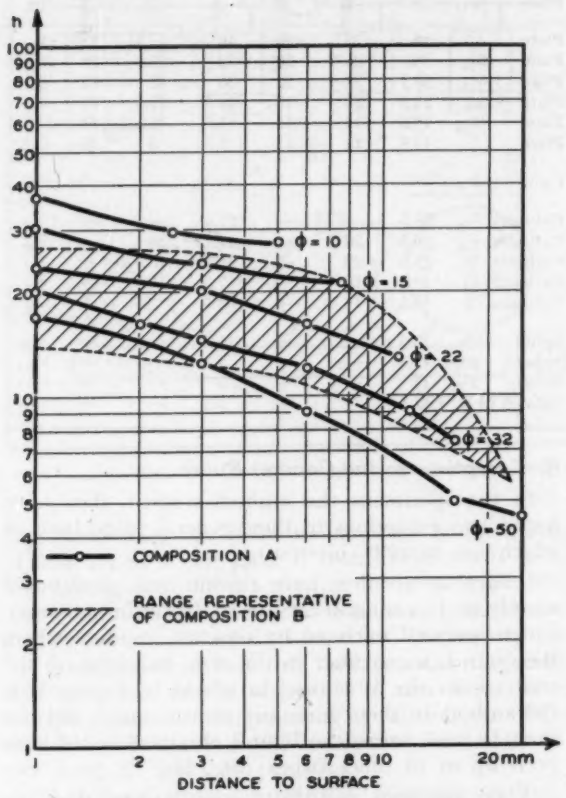


Fig. 3 — Grain size in cylinder specimens poured at 1500 C (2730 F). (Number of cells per centimeter)

(Below)

Fig. 5 — Structure of plate castings 11/16 in. thick (left) and 5/16 in. thick (right). Mag. — 10X. Etch — Stead-LeChatelier reagent.

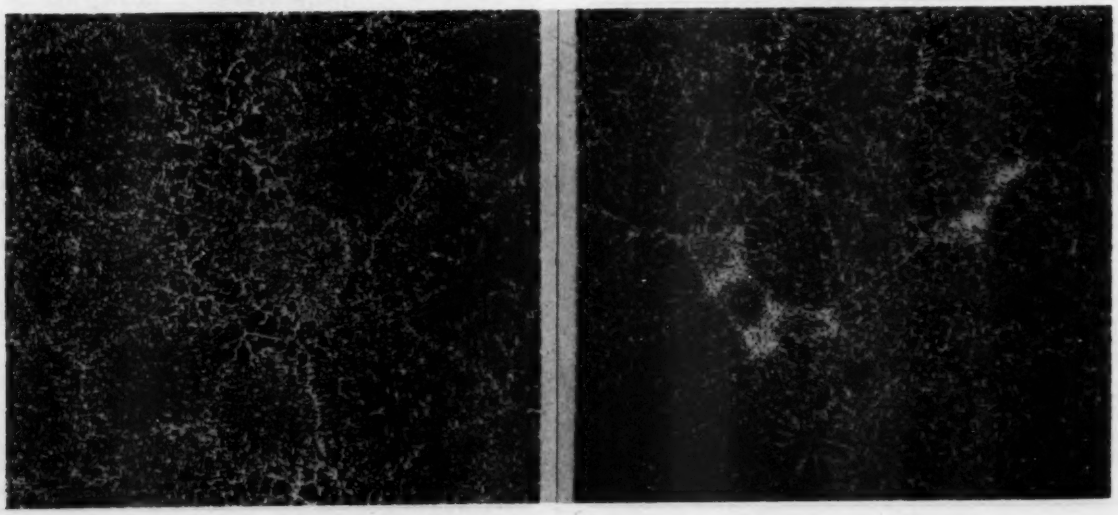
The authors therefore feel that these observations prove that this anomaly is due to a noticeable supercooling, the reason for which the authors do not know, and which thus apparently is an example which contradicts the general rules, as the solidification is more rapid, the supercooling is more pronounced.

2. As a second exception to the general rules one notes in the cases where the cooling is very rapid without, however, resulting in a chilled iron, the occurrence of a larger grain, which one would not expect from the results of the experiments performed with the same iron under different conditions. We shall see this anomaly occurring again when we discuss the influence of temperature on pouring. This phenomenon was observed in tests on casting B (Tables 3 and 4) but it is more evident in the photomicrographs in Fig. 6.

This is a sample of chilled cast iron. The photomicrograph at the right in Fig. 6 represents the surface poured against the sand opposite to the chill.



Fig. 4 — Structure of 1-in. thick plate. (Casting No. 1285). First type anomaly of grain size. Mag. — 5X. Etch — Stead-LeChatelier reagent.



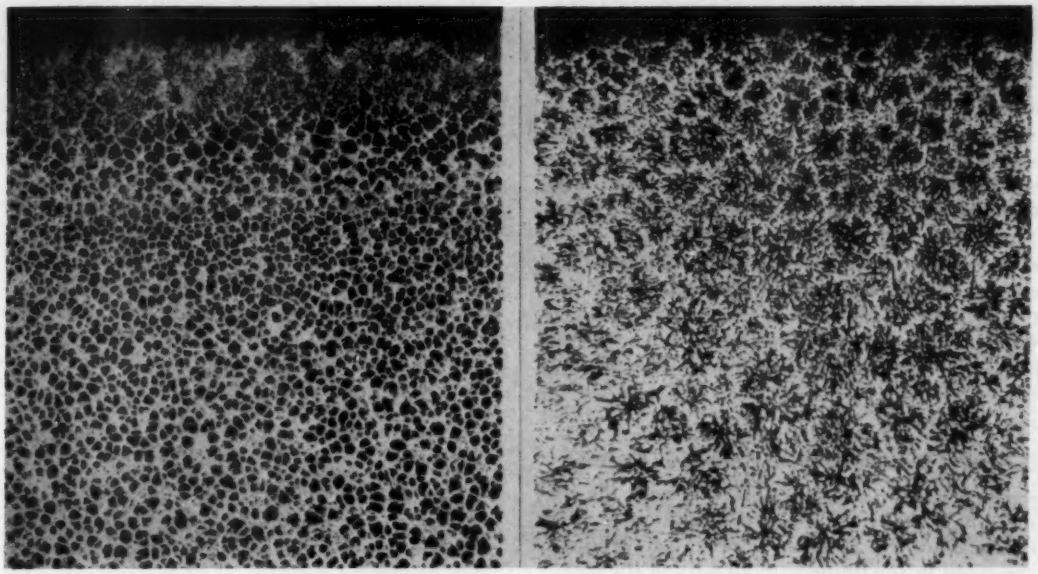


Fig. 6 — Grain structure of chill-cast piece. Left photomicrograph shows surface on the chill side (second type anomaly) and the right photomicrograph shows the surface on the sand side. Mag. — 10X. Etch — Stead-LeChatelier reagent.

The cells become larger and larger proportionately to the distance from the sand. Contrariwise, the photomicrograph at the left in Fig. 6 representing the surface poured against the chill shows clearly the outline of a zone where the eutectic cells are much larger, a zone which separates a mottled zone near the chill from a region where the cells are particularly fine.

These findings are most astonishing especially in view of the experiments by Tammann on nucleation. The number of nuclei formed per unit of time volume is a function of degree of supercooling which itself increases with the cooling rate. For a very slight supercooling the number of nuclei is small, which normally gives rise to large crystals; this is what the authors observed in the center of thick sections for which the rate of cooling was small.

Likewise, as the degree of supercooling increases, the number of nuclei also increases, which is expressed by a much smaller grain (case of the first type of anomaly). But when the supercooling becomes large the number of nuclei decreases and a large grain results; this seems to be what happens for the abnormal results just indicated, as shown in Fig. 6 in the area of a chill. It seems that in the cast irons this stage of grain growth, when it can be observed, is produced by such a degree of supercooling that a slight increase of such supercooling leads to the production of white iron.

In other words, one notices the production of cementite crystals in the cast irons when the nucleation and growth of the eutectic cells does not occur rapidly enough; the heat developed by such solidification is not sufficient now to compensate the loss in the mold, the temperature drops and the degree of supercooling becomes great enough for cementite crystals to appear. This explains why one observes in the mottled cast irons eutectic cells which did not have the time to become completely developed and are floating around in cementite.[6]

Likewise, as the degree of supercooling, i.e. the rate of cooling increases, one often notices the succession of the following structures for a cast iron of a given composition: (a) larger eutectic cells, thicker graphite flakes; (b) fine eutectic cells, medium-sized graphtie flakes; (c) very fine eutectic cells, graphite in rosette form, or in supercooling; (d) larger eutectic cells supercooled graphite; (e) eutectic cells surrounded by cementite; (f) some eutectic cells immersed in cementite; (g) white cast iron.

These observations are shown in Fig. 7.

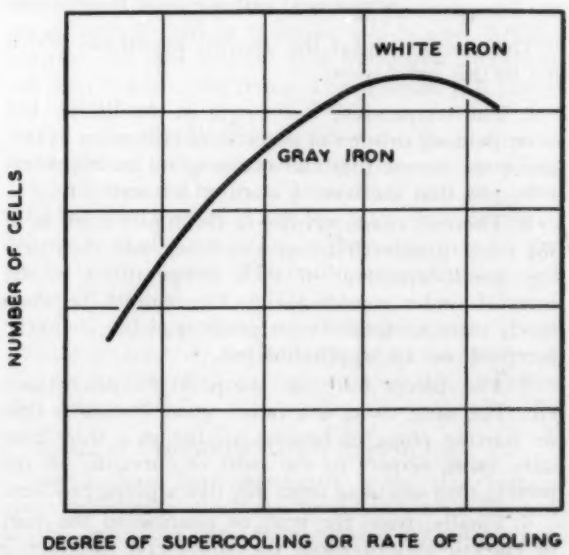


Fig. 7 — Relationship of number of cells vs degree of chill or rate of cooling.

2. Influence of Pouring Temperature

Let us refer once more to the grain size measurements in Tables 2, 3, 4. In Section I.1 of this paper the authors used them as they refer to changes in compositions A and B. The authors also performed tests on an iron with higher carbon content, composition E, which has the following analysis: T.C. 3.56 per cent; Si, 1.45 per cent; Mn, 0.61 per cent; S, 0.04 per cent; and P, 0.34 per cent.

A series of plates was poured at temperatures approximately 1500 C (2730 F), 1400 C (2550 F) and 1300 C (2370 F). The authors counted the average number of cells at the surface and in the center of each piece; the influence of pouring temperature has been shown to exist, depending on the location of the area examined.

A. Influence of Grain Size Near Surface of Casting

Theoretical Considerations — Starting with the basic hypothesis by Fourier on the propagation of heat, one can calculate the absolute temperature gradient, G, in the mold in the immediate vicinity of the metal-sand interface at the instant eutectic solidification sets in.

The authors show in the Appendix that the following formula is valid:

$$G = \frac{2\,(T_1 - T_o)^2}{V/S\,\Delta T)} \quad \frac{c\,\theta}{c'\,\theta'}$$

where:

- V is volume of the piece
- S is area of the metal-sand interface
- T_1 is temperature of eutectic solidification
- T_o is temperature of mold prior to pouring, i.e. room temperature
- ΔT is $T - T_1$, the difference between pouring temperature and solidification or supercooling temperature
- c, θ, c', θ' are specific heat and densities of the sand and the iron, respectively.

The authors repeat the applied hypotheses which led to this conclusion:

1. The temperature is uniform in the liquid iron from pouring to start of eutectic solidification. Thus, one must assume that convection plays an important role, and that the iron is more or less eutectic.

2. Thermal characteristics of the liquid iron, as of the mold (conductivity, specific heat), and their density are independent of their temperatures in the interval under consideration. This cannot be absolutely correct, especially for green sand but it can be accepted as an approximation.

3. The pieces have no sharp edges, projections. etc. The sand mold is a rather good insulator, thus its starting phase of heating applies to a thin layer only (with respect to the radii of curvature of the piece), thus one may treat this like a plane problem.

4. Finally, from the start of pouring to the start of eutectic solidification, the sand-metal interface is kept at a constant temperature (at exactly the temperature of eutectic solidification) by the presence of a very thin layer of solid cast iron which by becoming thicker or thinner, serves as a thermal regulator.

This hypothesis is largely suggested by practical experience; it is often the base of the theoretical calculations in the solution of the problem of how the liquid cools in the mold.

It is also logical (and it is even the validity of this hypothesis which the authors wish to compare with the experimental results) that the value of G determines the eutectic grain size at the surface because it establishes the amount of heat lost per unit of time and surface at the instant of formation of cellular nuclei.

This heat loss per unit is in accordance with Fourier's principle

$$q = kG = A\ (T_1 - T_o)^2\ \frac{S}{V\,\Delta T}$$

where A is a constant for the type of iron and the mold (conductivity, specific heat, specific gravity). Thus from our hypotheses one is led to conclude that if one always uses the same composition iron, the same sand mixture which is subjected to the same mixing technique, then on pouring at certain temperatures, of pieces of various shapes, the number of surface cells found at locations where there are no sharp projections, etc., will be a function of $V/S\ \Delta T$ only, $n = \frac{f\,(V\,\Delta T)}{S}$.

We shall now examine the experimental results to determine to what extent our hypotheses and approximations are justified in their entirety.

Experimental Results — (a) For practical purposes the authors adopted for cast iron A the formula $\Delta T = T - 1200$, where T is the pouring temperature.

As an example, the authors present in Fig. 8 results obtained with iron of composition A poured as small plates; the correlation between the number of cells and $V/S\ \Delta T$ is excellent. The experimental values for the lower pouring temperature there is greater scatter.

In the case of cylinders, there were a number of abnormal findings. This is due to the fact that the

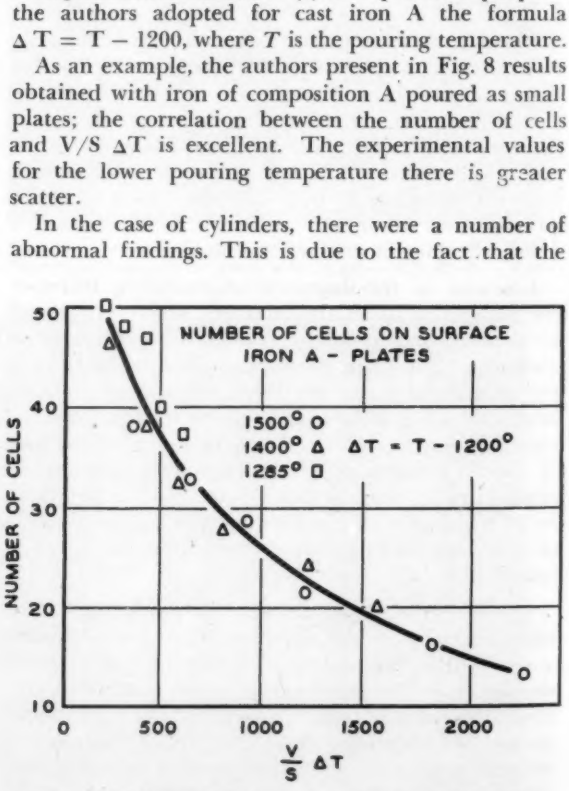


Fig. 8 — Relationship of number of cells vs $V/S\,\Delta T$.

pieces in question, although they are not mottled, are almost surface-tempered. The authors have already pointed out that in such case the number of cells will decrease noticeably. The authors elaborate on this point more fully.

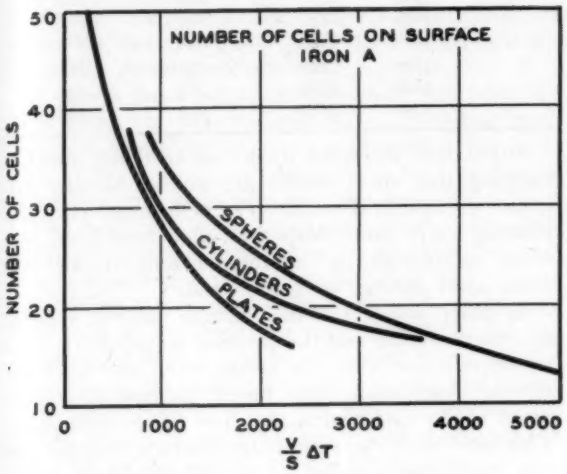


Fig. 9 — Relationship of number of cells vs $V/S\,\Delta T$.

If one plots on one graph the average curves obtained for plates, cylinders, and spheres for the case of one iron of composition A, one sees that they are displaced with respect to each other (Fig. 9), the cylinders are between the two other curves. This is because the volume of sand active in the dissipation of calories varies per unit of metal-sand contact surface with the shape of the pieces, notwithstanding our hypotheses.

(b) These results are confirmed for two more irons, B and E, but here it is better to use the equation $\Delta T = T - 1150$, which takes the supercooling into account.

(c) The experimental results permit the use of the expression $V/S\ (T - \theta)$, as a variable, where T is pouring temperature, and θ is a temperature selected for the specific iron, which may be quite different from the theoretical value T_1 (solidification temperature of the austenite-graphite eutectic). Why this value is so different will be discussed subsequently.

At the instant of eutectic solidification near the interface there exists, in spite of our first hypotheses, a temperature gradient in the liquid, so that the heat absorbed by the mold is much lower than the calculated value.

Thus in the case of our eutectic cast iron (composition A) the temperature θ as found experimentally is about 50 degrees higher than the theoretical value T_1.

For the case of non-eutectic cast iron, the formation of cells is preceded by the precipitation of primary crystals; they will liberate an amount of supplementary heat which has not been taken into consideration in the calculation.

It is understandable now that the effects of the temperature gradient in the liquid and the primary solidification can compensate each other; thus one may find for θ the calculated value, as one does for the case of irons B and E which are definitely hypoeutectic.

B. Influence of Grain Size in the Center of the Pieces

1. In this case it is not possible to reason in the same way as in the case which made it possible to derive the law of variations of the number of cells in the surface.

In order to obtain this result one must determine the temperature gradient in the center of the piece at the instant of eutectic solidification. But for this we would have to know when the inside is partially solidified, temperature distribution, and this can be done only by an intricate and cumbersome experimental set-up.

2. Thus we have chosen to use the empirical method. By using a logarithmic scale for the cooling rate, when tracing the curves for the variations of the number of cells in the center of the piece as a function of V/S, the authors found that the characteristics for the various pouring temperatures can be obtained from one curve, if one shifts it.

This now means again that all the experimental points pertaining to one casting and to a certain cast iron must be found on one curve only, if one uses a corrected cooling rate $K \cdot V/S$, where K varies with the pouring temperature.

For all the irons studied, independent of the geometry of the pieces, it became obvious as an estimate that the value of K depended only upon the pouring temperature. Thus we were led to assign

$K = 1$ for pouring temperature 1300 C (2370 F)
$K = 1.15$ for pouring temperature 1400 C (2550 F)
$K = 1.3$ for pouring temperature 1500 C (2730 F)

The experimental points arrange themselves ideally now using these constants. Only some results pertaining to very thin plates of composition E deviate noticeably. These points obviously correspond to pieces with a critical thickness which any chilling, however fast this may be, will transform into white cast iron down to the center. Composition E is rather low in silicon.

One notes also that the number of cells in the center is almost independent of the shape of the piece, at least within the limits of the authors' tests on plates, cylinders and spheres. Thus one may place the curve for one kind of piece on top of the other curve for the piece obtained earlier. The results appear for the irons A and B in Figs. 10 and 11.

Finally, the authors consolidate in one graph average curves obtained with the three compositions of iron studied. Figure 12 shows how close together the curves are.

C. Effect of "Abnormal" Grain Growth

The authors have previously noted that certain abnormal results which can be related to an increase of the grain size may be explained by nucleation, when dealing with rapidly cooled irons, or low-silicon irons.

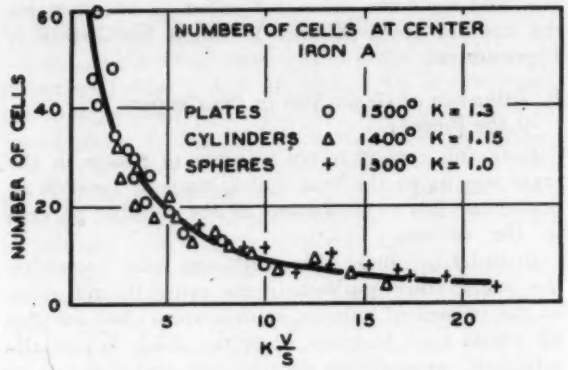


Fig. 10 — Relationship of number of cells vs KV/S.

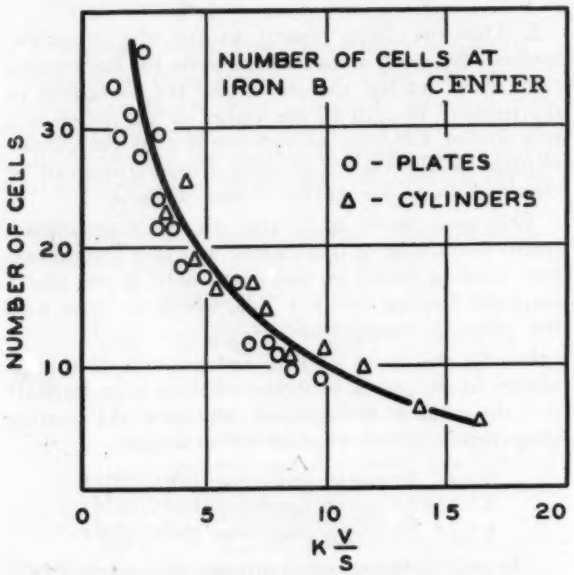


Fig. 11 — Relationship of number of cells vs KV/S.

As the test results were somewhat scattered, the authors ran some additional tests.

For this purpose a number of 20-mm (¾-in.) bars were cast in green sand and dried sand at pouring temperatures from 1500 C (2730 F) to 1300 C (2370 F). The analysis selected was a gray iron with a tendency to become tempered. The analysis was as follows:

Carbon	— 2.65 per cent
Silicon	— 2.10 per cent
Manganese	— 0.70 per cent
Sulphur	— 0.09 per cent
Phosphorus	— 0.14 per cent

Several pouring methods have been used as follows:

X: Upright pouring through an edge which forms a riser.

Y: Horizontal pouring with individual risers.

Z: Horizontal pouring without riser.

Each test value shown in the following table is average of eight tests on cylinders poured under identical conditions.

Pouring Method	Pouring Temperature C	Pouring Temperature F	Number of cells at center Green Sand	Number of cells at center Dried Sand
X	1500	2730	20.0	21.2
	1400	2550	20.1	23
	1300	2370	18.9	19.9
Y	1500	2730	20.2	21.2
	1400	2550	20.9	20
	1300	2370	19.2	20
Z	1500	2730	21.6	21.2
	1400	2550	23.1	22.1
	1300	2370	16.6	20.5

An analysis of results in the above table, notwithstanding that these results are not statistically different, shows that the number of cells drops when the pouring temperature drops, and that there is the tendency to increase the number of cells, if one uses dried sand instead of green sand.

In other words, the two factors, namely, type of the sand mixture and the pouring temperature, have an effect contrary to the assumptions on which our general observations were based; the special composition of the cast iron is the cause of this exception. This now leads the authors to conclude that for this cast iron poured under the conditions shown above, the representative point is found in the right-hand side of Fig. 7, i.e. in the curve above its maximum; thus a slightly slower cooling rate (dried sand instead of green sand, higher pouring temperature) will lead to a finer grain.

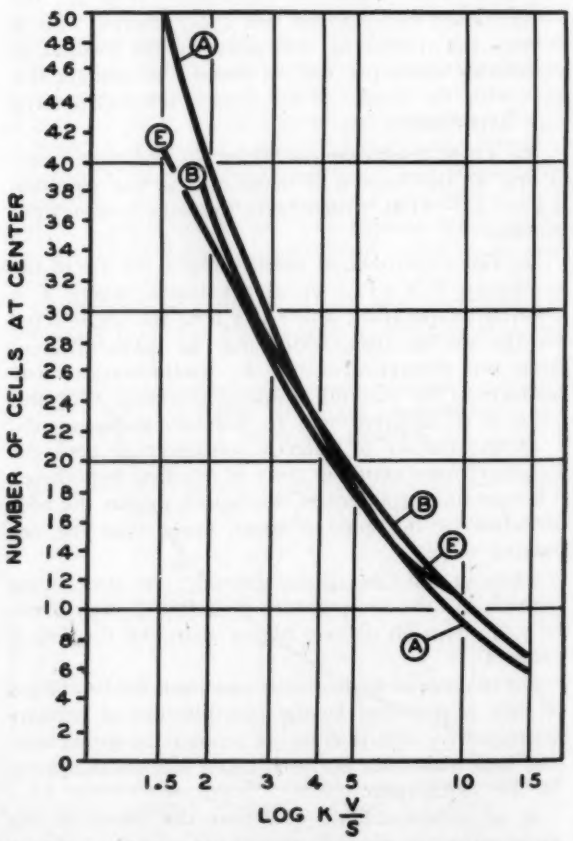


Fig. 12 — Relationship of number of cells at center vs log KV/S.

Conclusions to Section I

In summary, the authors' observations relative to conditions, as they affect grain size, may be stated as follows:

1. In gray cast irons the austenitic grain is generally finer the smaller the thickness of the piece, the closer the observed point is to the surface, and the lower the pouring temperature.

2. There are, however, exceptions to this rule. The most important ones are encountered in high-strength cast irons where the Si content is adjusted to the thickness of the casting, i.e. where it is kept as low as possible to still get a gray cast structure.

For these cast irons the eutectic grain varies little when the pouring temperature drops; this variation mostly favors an increase of the grain size.

3. Where the tendency to tempering does not manifest itself to any great extent, a decrease in pouring temperature brings about a stronger effect on the grain size at the surface as compared with the center.

In other words, when the pouring temperature drops, the differential in grain size throughout a piece or section becomes more pronounced.

4. The change in grain size of cast iron becomes more pronounced with change of pouring temperature, the thinner the piece or section.

For practical purposes, the effect of pouring temperature is very small at the center of pieces or sections larger than 20 mm (¾ in.) in thickness.

5. The size of the grains at the surface is the same for pieces similar in geometric form, poured at the same temperature of the same composition iron whose product $V/S\ \Delta T$ is constant, where ΔT is the supercooling of the liquid iron.

6. The grain size in the center of a piece is practically independent of the form of the piece. It varies mainly with the corrected cooling rate, $K.V/S$, where K is a function of pouring temperature when

$K = 1$ for a pouring temperature of 1300 C (2370 F)
$K = 1.15$ for a pouring temperature of 1400 C (2550 F)
$K = 1.3$ for a pouring temperature of 1500 C (2730 F)

In other words, the effect in the center of the piece is as if the cooling rate were higher, the higher the pouring temperature.

II. Influence of Composition on Grain Size

In preceding Section 1, when examining how the law of cooling of a piece influences the grain size, we have always tried to establish certain qualitative rules. These rules are generally valid, with the exceptions herein noted; the authors have never been disillusioned in this assumption in studying 40 various studies.

As would be expected, the grain size varied from one cast iron to the next, although the conditions of melting and solidification were carefully controlled. Examination of test results for cast irons A and B shows this clearly. Figure 3, for example, shows the curves for the number of cells per centimeter, n, as a function of the distance from the surfaces of the same cylinders, and one can see that the curves do not coincide with each other.

TABLE 5 — DATA ON ¼-IN. DIAM. CYLINDERS POURED AT 2550 F

Casting No.	Analysis, % C	Si	Mn	S	P	No. of Cells, at Surface	at Center
1 415	3.44	2.62	0.52	0.05	0.20	22	9
1 417	3.39	2.61	0.48	0.04	0.54	25	11
1 529	3.36	1.69	0.62	0.07	0.26	21	11
1 532	3.25	1.59	0.59	0.08	0.59	21	12
2 036	3.08	2.20	0.80	0.05	0.31	14	11
2 044	3.10	2.79	0.67	0.08	0.18	12	9
2 045	3.04	2.88	0.73	0.05	0.98	10	9
2 046	3.33	2.86	0.76	0.08	0.18	20	10
2 146	3.44	2.67	0.69	0.06	0.28	22	11
2 147	3.18	2.13	0.68	0.06	0.28	10	8
2 160	3.04	1.82	0.66	0.03	0.28	14	11
2 307	3.68	2.82	0.66	0.06	0.18	—	—
2 315	3.18	1.75	0.58	0.13	0.20	15	10
2 322	3.26	1.73	0.58	0.12	0.20	18	11
2 326	3.53	1.70	0.54	0.04	0.27	16	12
2 343	3.12	1.84	1.07	0.09	1.01	15	13
2 344	3.53	1.35	0.56	0.05	0.22	18	10
2 352	3.33	1.74	0.40	0.10	0.35	25	18
2 360	3.33	1.77	0.39	0.10	0.34	23	17
2 515	3.04	1.83	0.44	0.10	0.38	17	13
2 538	3.74	1.77	0.37	0.08	0.39	33	14
2 363	3.64	2.85	0.54	0.07	0.28	22	12
2 364	3.43	2.18	0.50	0.11	0.30	15	10
2 368	3.21	2.51	0.59	0.05	0.27	12	9
2 402	3.05	2.50	0.39	0.11	0.32	20	15
2 543	2.96	1.87	0.43	0.09	0.34	17	10

The most complete results we have thus far on 32-mm (1¼-in.) diameter cylinders poured at a temperature of about 1400 C (2550 F) after superheating the molten metal to 1500 C (2730 F) are shown in Table 5 with the number of cells at the surface and in the center for 21 irons of different compositions (five elements). These values are averages of tests on several cylinders. In order to check the validity of these results, by considering the results obtained on the 20-mm (¾-in.) cylinders from Heat No. 2036, the authors obtained a characteristic scattering of the results as follows:

Center: Average 11.15, scatter type 1.09
Surface: Average 13.93, scatter type 0.67.

Thus with a probable accuracy of 95 per cent one can count the number of cells at the surface and in the center, within at least two units if the scattering of the results follows the "normal" law. Thus one concludes that from one melt to the next, i.e. by changing the make-up of the charge and thereby the resulting chemical composition of the castings poured, one finds variations in the grain size which cannot be explained by the various experimental errors which may occur in the foundry or in the laboratory. These discrepancies are more relevant when measuring grain size near the surface than in the center of the cylinders.

1. Influence of the Total Carbon Contents

The authors made a chemical analysis on each pour and the relation between the various elements for all these melts is purely accidental. The authors made a statistical study of these analyses to determine whether there is a relationship between the carbon contents and the grain size. Calculation shows that the number of cells near the surface of the piece increases with

increasing carbon content according to the formula:

$$n = 18.2 \times \%C - 41.3 \pm 8.2 \text{ (Fig. 13).}$$

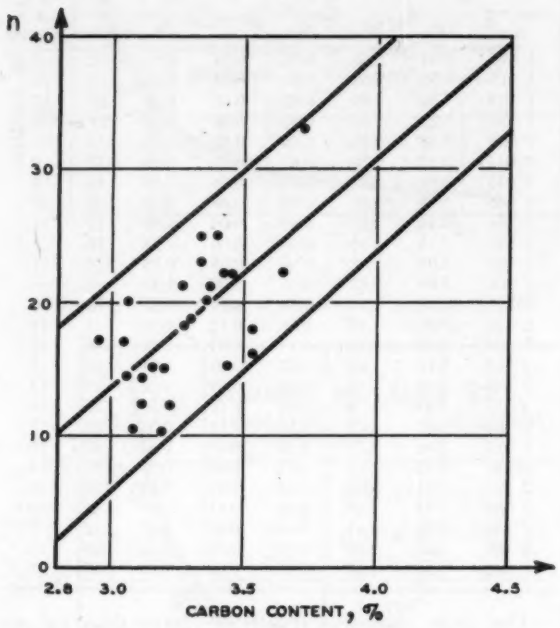


Fig. 13 — Relationship between grain size vs carbon content.

This formula was correct in at least 95 per cent of the cases. It is well to remember the reservations applying to statistical results. Thus it would be wrong to use this formula for cast irons which are not within the following range of analysis:

Total Carbon	— 3.0	- 3.7	per cent
Silicon	— 1.6	- 2.9	per cent
Manganese	— 0.4	- 1.1	per cent
Sulphur	— 0.03	- 0.12	per cent
Phosphorus	— 0.4	- 1.0	per cent

Even if the Total Carbon of a cast iron is known, the number of cells in the surface of a 32-mm (1¼-in.) diameter cylinder poured at 1400 C (2550 F), as specified, indicates uncertainty because the scattering is six times as high than in tests where one compares a number of cylinders being made in one casting only.

2. Influence of the Other Chemical Constituents

The authors can only give a brief interpretation of results shown in Table 5. These results tend to corroborate our earlier conclusions about the influence of phosphorus which is absent when the phosphorus content is in the range of 0.10 to 0.60 per cent. Thus if one compares test results on castings No. 1415 and 1417 with castings No. 1529 and 1532 one sees that doubling the phosphorus content produces insignificant results insofar as changes in grain size are concerned.

The authors did not notice much influence of the content of other elements, and should a complete statistical analysis show some actions or interactions, these presumably will be rather weak and can never explain the scattering of the results, which could also be caused by the variations in the non-analyzed-for elements.

Thus, the authors summarize the observations reported in Section 1 as follows:

By introducing into the crucible charge materials of various origin (hematite, semi-phosphorated charge, molten ingot, etc.) with additions, if necessary (ferrosilicon, ferrophosphorus, etc.) we made various kinds of test pieces with different chemical analyses, not only with respect to the five main elements, but undoubtedly also with respect to other elements which were not determined. But the influence of these "secondary" elements on grain size, if at all noticeable, cannot overshadow the influence of carbon which is the strongest.

III. Relation Between Grain Size and Mechanical Properties

The authors have thus far pointed out that the factors which influence the grain size of gray cast iron are many and it is not possible to examine each one individually. Some of these have a stronger influence on the other two essential characteristics of the structure, namely, number and size of the graphite flakes and the type and fineness of the matrix.

An analysis of these various influences can be made only with great difficulty.

But it is well established by now that mechanical properties of pearlitic gray cast iron depend, in part, on the nature of the matrix, which is similar to steel, if one considers that graphite has practically no tensile strength and perhaps very little compressive strength.

1. By applying knowledge of strength of materials one can analyze the distribution of stresses within the piece; but one must remember that even though it may be simple to evaluate the mechanical properties of the matrix and of the graphite present, it is more difficult to arrive at the distribution, because the form of the graphite flakes which play an important role in the concentration of strains, produces local overloads of the elastic limits and of the breaking loads.

2. One may also try to define the influence of each of the basic constituents individually, the matrix and the graphite, by reconciling and comparing the mechanical properties of one sample. One of the coauthors in a paper[4] discussed the relation between tensile strength, *R*, the modulus of elasticity, *E*, and the hardness, *H*. This relation was expressed by A. Collaud by the formula

$$R = K_2 \cdot E \cdot H$$

where K_2 is more or less constant for low-P gray irons with flake graphite. The interesting thing in regard to cast irons is that the modulus of elasticity does not depend upon the shape of the graphite flakes, that the hardness depends mainly on the matrix, but is of the graphite, especially if it is present in the form not completely independent of the amount and form of large flakes.

Most authors referred to do not take grain size into account. Yet this feature governs the distribu-

tion of the crystals of the phosphorus eutectic, especially if there are many of them; it also governs the distribution of the graphite. Thus should the grain size be overlooked? The authors feel that this is wrong.

In 1943 R. R. Adams[7] found that the strength of a cast iron cannot be expressed only in terms of graphite and matrix, but also in terms of grain size. He reported the case of two samples having the same composition (five elements), identical matrix and graphite, but one had a fine grain and the other a coarse grain, with a dimensional ratio of about 1:3. The tensile strengths were 37,000 to 47,500 psi (1.2-in. diameter bars) while the brinell hardness of these bars was practically the same. The author unfortunately did not specify how he measured such varying properties in the two samples. This also holds true for work done by H. Morrogh.[8]

This author suggested that the strength of a cast iron must be closely associated with the dimensions of the eutectic cells, he based his statements on results obtained by Bunin, Malinochka and Federova[9] according to which all the particles of graphite belonging to one cell, come from one and the same nucleus. Thus we must visualize the inside of the cell as a mass of graphite, spongelike, yet continuous. One can thus conceive that the strength of the aggregate matrix graphite does not solely depend on the shape of the graphitic nucleus, but also on its size, and these two characteristics could be related in some cases.

Examination of graphite under high magnification, comparing with a graphite type chart, would bring out only, if we assume Morrogh's ideas are correct, something of the extremely warped shape of this graphite nucleus; whereas a study of the grain size, as the authors did in this study, will lead almost directly to the apparent volume of the nucleus as it could by itself, as Adams found, act on the strength of the iron, all other factors remaining constant.

Based on our experimental data, we shall try now to get some indication of the influence of grain size itself, by expressing this factor as a function of variables which were defined in Section 1 of this paper as a function of the cooling rate and the pouring temperature.

These two factors have the same influence on the grain size that we simply have to multiply the value of V/S with an equilibrium coefficient, K_1, to take care of the differences in pouring temperatures. Such a process is unfortunately not applicable in general to determine the mechanical properties; they will depend on the effect of the product $K_1 \times V/S$ alone, as this would be the case if the grain size alone would be all that is needed to bring out the structure.

1. Cooling Rate As a Variable

Let us study some cylindrical castings ranging from 90 mm (3½ in.) to 100 mm (4 in.) in diameter. In Table 6 the authors show tensile strength, hardness, Young's modulus, and the number of cells per centimeter for the axial region of each of nine types of cast iron poured at 1400 C (2550 F). The authors have rejected from this table test values of mottled castings. The tensile strength, hardness, modulus of elasticity, and the number of cells, as a rule, increase, considerably as the diameter of the cylindrical casting decreases.

This qualitative result is in accordance with our earlier findings (Section 2) on five different cast irons where an increasing linear relation was found between shear strength and the number of cells per centimeter. However, the increase in tensile strength as a function of the number of cells seems to be better represented by a straight line in log-log coordinates. See Fig. 14.

This now brings us to the point that the function between R and n must be expressed as $R = Cn^{\gamma}$ where C is a constant and γ a coefficient with a positive value from 0.2 to 0.4 for irons with a phosphorus

TABLE 6 – DATA ON TEST CYLINDER CASTINGS POURED AT 2550 F

Casting No.*	Diameter, in.	R, psi	E, psi x 10^6	H (Brinell)	n (at center)
2 044	3½	21,200	14.4	177	3
	2	25,500	15.9	197	5
	1¼	33,000	16.1	212	9
	⅞	35,800	16.75	233	14
	19/32	42,200	17.2	246	20.5
	⅜	52,500	17.6	288	—
2 045	3½	12,850	14.7	193	—
	2	17,500	17.6	211	—
	1¼	22,600	16.75	229	9
	⅞	27,400	16.5	241	10
	19/32	31,300	18.45	257	16
	⅜	39,100	19.4	285	—
2 046	3½	20,500	10.5	145	3
	2	27,200	13.1	159	5
	1¼	33,000	14.25	175	10
	⅞	33,100	14.95	197	12
	19/32	33,000	15.4	218	22.5
	⅜	46,800	16.1	256	—
2 307	3½	19,300	12	156	7
	2	18,600	12.6	156	—
	1¼	19,400	12.85	149	—
	⅞	22,800	13.8	156	20
	19/32	26,400	13.3	202	40
	⅜	30,700	14.4	214	56
2 315	3½	32,700	18.6	206	3
	2	37,600	18.7	207	4
	1¼	40,500	18.9	212	10
	⅞	43,500	19.7	231	18
2 322	3½	28,200	16.5	186	3
	2	32,700	18.45	193	5
	1¼	39,100	19	204	11
	⅞	41,600	17.9	218	18
	19/32	43,200	19.2	246	25
2 326	3½	18,600	13	153	4
	2	21,200	14.95	159	6
	1¼	27,200	15.2	170	12
	⅞	34,200	15.5	197	13
	19/32	35,000	16.7	222	20
2 343	3½	16,200	17.6	184	4
	2	26,000	18.6	234	9
	1¼	35,000	19.1	262	13
	⅞	39,100	20.4	276	20
2 344	3½	13,300	15.4	157	3
	2	21,400	16.1	158	5
	1¼	24,700	16.6	189	10
	⅞	29,400	16.6	211	17
	19/32	33,000	17.5	222	22

*See Table 5 for composition.

content approximately 0.20 per cent; γ is about 0.6 for irons containing a higher phosphorus content up to 1 per cent. Comparison of the straight line graphs for each of the low-phosphorus irons will point out that the lines representing the higher-carbon irons belong to the "stronger" irons, i.e. in our case the ones with the low-carbon content.

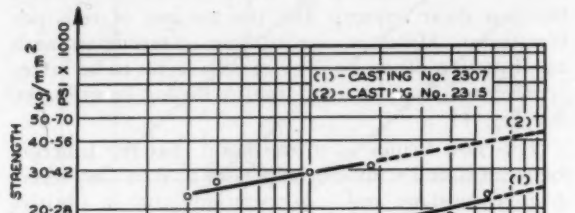


Fig. 14 — Relationship of tensile strength vs number of cells per centimeter.

If we overlook the high-phosphorus irons, then we may conclude from these observations that the influence of the grain size is additive only to the one of graphite and of the matrix so as to increase the tensile strength whenever the cooling rate decreases. We may state, however, that the grain size may well represent for each variable the whole of the structural appearance, the changes of which are interdependent.

Let us compare for instance the 3½-in. diameter cylinders from Heat No. 2307 and 2315. The number of cells per centimeter are 7 and 3 respectively, and the tensile strengths are 19,300 psi and 32,700 psi. The difference between these tensile strengths is considerable, but it would be still greater if the increase in graphite would not be accompanied by an improvement of the structure.

The extrapolation of the straight lines plotted in Fig. 14 shows that if the grain size would be comparable in the high-carbon cast iron as in the other, the tensile strength would presumably not be more than 15,400 psi instead of about 19,300 psi. And vice versa, one could hope to get 37,700 psi for the second type if one could make the grains as "fine" as in the 3½-in. cylinder of the first type.

2. Pouring Temperature As a Variable

In a paper entitled "Influence of Pouring Temperature on Mechanical Properties of Ordinary Gray Cast Irons"[10] one of the co-authors has proven that in medium-thickness castings the following test values are obtainable as summarized in Table 7 and expressed as follows:

a. For high-carbon cast iron (Casting No. 2146), one notes that cylinders poured at 1300 C (2370 F) have a much higher strength and Young's modulus than castings poured at 1500 C (2730 F); hardness varies little, showing no definite trend.

b. For low-carbon cast iron (Casting No. 2402) a drop in pouring temperature results in a decrease in tensile strength, the modulus changes in one direction and then another, and there is a slight decrease in hardness.

TABLE 7 — DATA ON 1.2-IN. DIAMETER TEST CYLINDERS

Casting No.	Pouring Temperature, F	Tensile Strength psi	Modulus of Elasticity, psi x 10^6	Brinell Hardness	No. of Cells at Center
2146*	2730	21,700	13.3	178	19.5
	2550	22,100	14	179	22
	2370	23,800	15	180	26.5
2147*	2730	38,000	17.3	225	10
	2550	39,000	18.3	222	10
	2370	37,000	17.5	226	11
2402*	2730	42,800	18.3	241	9
	2550	42,400	19.4	239	10
	2370	37,500	19.1	236	10

*See Table 5 for composition.

The grain size tends to become finer as the pouring temperature drops, and this change is more pronounced in low-strength irons. This now means that there could be a negative or a positive relation between tensile strength and grain size.

Thus by decreasing the pouring temperature, we get

a. for low-strength irons the strength increases, because the graphite and the grains are finer; the modulus of elasticity correspondingly increases.

If one examines thin sections, such as ⅜-in. diameter cylinders sometimes the strength decreases in spite of the fact that the graphite and the grain are much finer; this is due to large amounts of ferrite in the matrix and this decreases the hardness.

b. for high-strength irons the strength decreases. The graphite is more irregular and the grain size varies little. This effect is presumably due to the occurrence of ferrite in larger quantities in the matrix (slight drop in hardness).

TABLE 8

Variable	Range	Sign of correlation between n and the variable	n and one mechanical property and the variable	n and one mechanical property	Mechanical Property
Cooling Rate, $\frac{V}{S}$	1/16-⅝ in.	−	−	+	E
			−	+	H
			−	+	R
Carbon Total	3.0-3.6%	+	−	−	E
			−	−	H
			−	−	R
Cooling Temperature, T_c (Low-Strength Iron)	2370-2730 F	−	−	+	E
			?	?	H
			−	+	R
Cooling Temperature, T_c (High-Strength Iron)	2370-2730 F	−	?	?	E
			+	−	H
			+	−	R

+ indicates size of the two factors varies in direct proportion
− indicates size of the two factors varies in inverse proportion
? indicates no definite relationship
E is Modulus of Elasticity
H is Hardness
R is Tensile Strength

Thus we must differentiate the influence of the matrix per se on the tensile strength; but the influence of the grain size or of the microstructural appearance of the graphite are interrelated. More selective tests must be made to pinpoint these influences which always appear simultaneously. Let us say that the grain size effect on strength can be evaluated more easily than that of the graphite flakes which often are within a three number range in the ASTM classification chart without even determining the effect in per cent of each individual type.

The authors' observations are summarized in Table 8 which shows the correlation between the number of cells per centimeter (the reciprocal of the grain diameter), the mechanical properties, cooling rate, total carbon content, and pouring temperature.

Table 8 applies only to those irons used in this study; the number investigated is still too small to arrive at valid general conclusions.

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APPENDIX

Let us look at the mold as an infinite medium, limited by a plane $x = O$, the temperature T_1 which remains constant starting from the instant of pouring at $t = O$. At the very beginning the mold is at temperature T_o throughout, and its physical properties are independent of the temperature (i.e. k, the heat conductivity; c, the specific heat; and θ the specific gravity. We calculate now the surface of the plane of separation.

The differential equation for the temperature of the sand, the function T of the distance x in the plane and of the time t, is as follows, written as a sequence of the basic assumption by Fourier:

$$\frac{\partial T}{\partial t} = \beta^2 \frac{\partial^2 T}{\partial x^2} \quad \text{where} \quad \beta^2 = \frac{k}{c\varrho}$$

The solution which satisfies the starting conditions (boundary conditions) is as follows:

$$T = (T_1 - T_o)\,\theta(z) + T_o \quad \text{with} \quad \theta(z) = \frac{2}{\sqrt{\pi}} \int_z^{+\infty} e^{-u^2}\,du$$

$$\text{and} \quad z = \frac{x}{2\beta\sqrt{t}}$$

Thus one can now obtain the temperature gradient in the mold at the sand-metal interface. The absolute value G may be written as

$$G = \left|\frac{\partial T}{\partial x}\right|_{x=o} = \frac{T_1 - T_o}{\beta\sqrt{\pi t}} \qquad (1)$$

The value t_s which corresponds to the start of eutectic solidification is arrived at by writing the heat Q lost by the metal in the interval $Q - t_s$ is the sum of the amounts of heat lost at every moment. If T is the superheat of the liquid metal with respect to the eutectic solidification temperature at the moment of pouring, one gets

$$Q = -k\int_o^{t_s} \frac{T_o - T_1}{\beta\sqrt{\pi t}}\,dt = \frac{V}{S}\,\Delta T\; c^1 \varrho^1 \qquad (2)$$

where c' and ζ' are the specific heat and the specific gravity of the liquid iron, respectively, assumed to be constant.

Thus we get

$$t_s = \frac{\pi}{4k\,(T_1 - T_o)^2} \left(\frac{V}{S}\Delta T\right)^2 \frac{c'^2 \varrho'^2}{c\varrho}$$

This value could perhaps be useful for risering calculations; we substitute it in Equation (1) and thus get

$$G = +\frac{2}{\pi}\,\frac{c\varrho}{c'\varrho'}(T_1 - T_o)^2\,\frac{S}{V.\Delta T}$$

The absolute value G of the gradient allows to get the amount of heat q, which is lost by the piece through unit of surface and per unit of time at the start of eutectic solidification:

$$q = +k\,G = \frac{2}{\pi}\,\frac{k\,c\varrho}{c'\varrho'}\,(T_1 - T_o)^2\,\frac{S}{V.\Delta T}$$

This is the formula used in the text, where we substituted

$$A = \frac{2}{\pi}\,\frac{k\,c\varrho}{c'\,\varrho'}$$

MODIFICATIONS IN VERTICAL-GATING PRINCIPLES

By

K. R. Grube, R. M. Lang, and J. G. Kura*

ABSTRACT

A further investigation was conducted on the improved vertical gating system described the previous year. One of the objectives was to determine whether minor modifications could be made in the system without increasing the degree of turbulence within the mold cavity or without increasing the probability of getting dirt or dross into the castings.

This investigation showed that the use of step gates instead of a continuous web connection between the side riser and the mold cavity did not cause any increase in turbulence. However, step gates caused undesirable thermal gradients in a test casting.

Changing the gating ratio so that a smaller runner could be used caused only a slight increase in the amount of turbulence. Screens in the gating system were no more effective than a runner extension in preventing dirt from entering the mold cavity. A core placed opposite the web in the improved vertical-gating system did not increase the turbulence within the mold cavity.

As in the previous work, the research was conducted by the water-analogy technique whereby flow characteristics of molten metal were deduced from high-speed motion pictures of water flowing in transparent models of molds.

Introduction

In previous research[1,2] an improved vertical gating system was developed that would deliver clean metal to the mold cavity with a minimum amount of turbulence. The main features of the improved vertical gating system, which is shown in Fig. 1, were as follows:

(1) A rectangular sprue, tapered with the smaller end at the bottom to prevent aspiration of air into the flowing stream of metal.
(2) A cylindrical enlargement of the runner at the base of the sprue to aid in rapidly washing out entrained air. If a wide, shallow runner had been used instead of a narrow, deep runner, a well[3] would have been preferable to the enlargement.
(3) A runner with a cross-sectional area equal to twice the cross-sectional area at the base of the sprue. The enlarged cross section of the runner reduced the velocity of the stream and thus minimized turbulence.
(4) An extension of the runner past the gate to provide a trap for the first liquid to enter the gating system. This liquid usually contains some dirt or dross.
(5) A gate that was flared with its exit end being the larger to reduce the velocity of the stream.
(6) A side riser to prevent turbulent liquid from entering the mold directly from the gate. The side riser also serves as a reservoir for feeding the shrinkage occurring in the casting during solidification.
(7) A web connection between the side riser and mold cavity to introduce the liquid into the mold quietly and progressively from the bottom to the top.

After the previous report[2] was presented, a limited amount of research was done to determine whether minor modifications could be made in the improved vertical gating system without increasing the degree of turbulence within the mold cavity or without increasing the probability of getting dirt or dross into the castings. The different aspects of this problem that were investigated were:

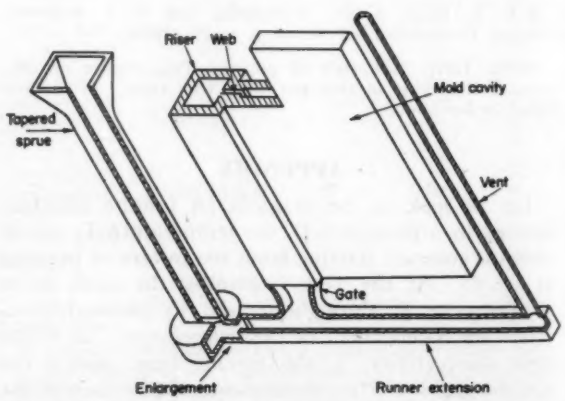


Fig. 1 — Schematic sketch of improved vertical gating system.

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(1) The use of step gates instead of a continuous web connection.
(2) Reduction of the cross-sectional area of the runner by changes in the gating ratio.
(3) The use of screens instead of a runner extension to trap dirt and dross.

Some work also was done to determine the effect on turbulence within the mold when the stream entering the mold cavity impinged directly onto a core.

The research on which this report is based was sponsored at Battelle Memorial Institute by the Frankford Arsenal under Contract No. DA-33-019-ORD-1383, and was guided by the Research Committee of the Light Metals Division of the American Foundrymen's Society.†

Method of Investigation

To investigate the effects of various changes in the improved vertical gating system, the flow of water through transparent plastic models of the gating systems was studied. The construction of these mold models has been described in a previous report.[1]

The effect of various changes in the gating system on turbulence within the mold cavity was evaluated by direct observation of the liquid flowing through the transparent models. When the action was too rapid to follow by direct observation, high-speed motion pictures were made of the liquid flowing through the model. With the aid of these high-speed motion pictures, the action in the molds was easily observed.

In the present tests, a rectangular plate mold 12 in. high, 6 in. wide, and ½ in. thick was used. The sprues were rectangular and tapered to a cross section of ⅛ by 1 in. at the bottom. Taper was applied to the thickness of the sprue only and was equal to 0.008 in. in total thickness per inch of length. The sprues were either 17 in. or 36 in. long.

Step Gates

Removal of gates from castings of irregular shape may be difficult if a continuous web connection is used between the side riser and mold cavity. To simplify removal of the gates, an intermittent web connection, commonly called a step gate, is sometimes used.

The effect of step gates on turbulence within the mold was investigated by blocking off sections of the continuous web, as shown in Fig. 2, and observing the flow of water through the system. Four step gates were used. The height of the two lower step gates was made smaller than the height of the two top ones to minimize the cascading that occurs when the liquid first builds up in the side riser. However, tests showed that cascading into the mold cavity during the early stages of the pour could be reduced more easily by increasing the thickness, T, (see Section A-A, Fig. 2) than by reducing the height of the lower gates. Gates of three different thicknesses, ¼, ⅜, and ½ in., were used.

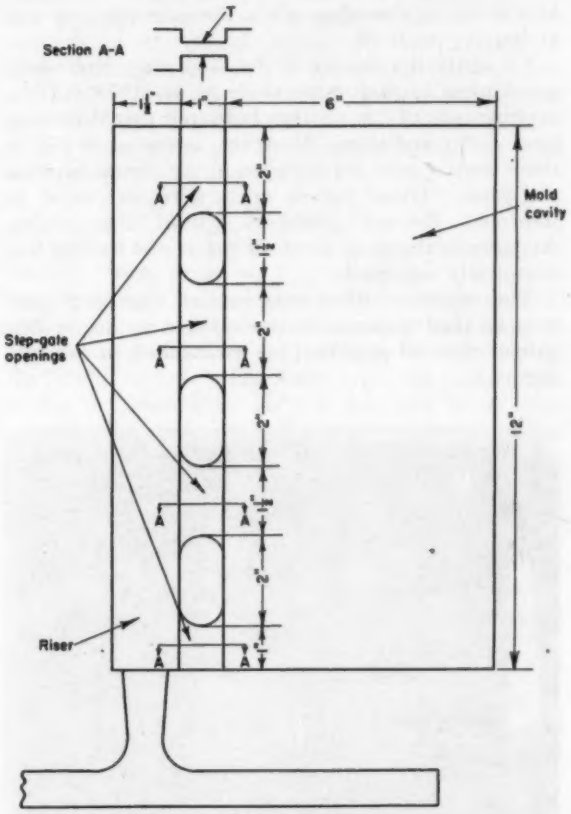


Fig. 2 — Schematic drawing showing location of step gates in the web connection.

Analysis of high-speed motion pictures taken of the flow pattern within the gating system indicated that no turbulence was created by the step gates. When the ¼-in.-thick gates were employed, the level of the liquid in the riser was considerably higher than the level of the liquid in the mold cavity until the mold was about one-third full. This difference in liquid level caused a slight cascading at the second gate but did not create severe turbulence. The degree of turbulence was comparable with the amount obtained when a ¼-in.-thick, continuous web connection was used. As the thickness of the step gates was increased, the level of the liquid in the mold during the early stages of the pour more closely approached the level of the liquid in the side riser. This, of course, reduced the amount of cascading and was the more desirable condition.

Dye introduced into the water near the end of the pouring operation revealed that the last liquid entered the mold cavity progressively from the second, third, and fourth gates. The last liquid poured was delivered near the top of the casting, thus promoting desirable temperature gradients that

†The Research Committee of the Light Metals Division of American Foundrymen's Society has consisted of the following members. W. E. Sicha, *Chairman;* W. Bonsack, R. F. Thomson, C. E. Nelson, T. D. Stay, W. J. Klayer, J. G. Mezoff, and H. J. Heine. H. Rosenthal represented Frankford Arsenal as an ex officio member of the committee.

would aid in the feeding of metal castings during solidification.

Results of these tests indicated that the continuous web connection can be replaced by a series of step gates without creating additional turbulence. The tests also showed that less turbulence was created as a result of cascading when the gate opening was at least 3/8 in. thick.

To verify the results of the laboratory tests, step-gated plate castings were made of an Al-7Mg alloy. Radiographs of the castings indicated that they were free of dirt and dross. However, as shown in Fig. 3, there were several porous areas in the region between the gates. These porous areas were attributed to improper thermal gradients within the casting. Apparently the gates froze off before the casting had completely solidified.

The results of these tests indicate that step gates may be used if precautions are taken to insure that proper thermal gradients are established in the casting.

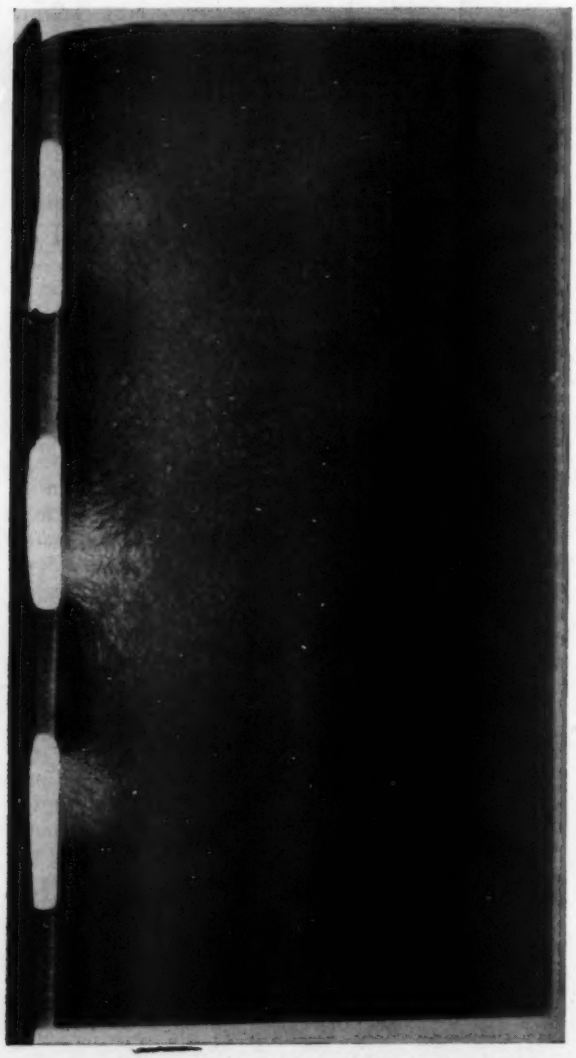


Fig. 3 — Radiograph of step-gated Al-7Mg alloy casting.

TABLE 1 — DIMENSIONS, IN INCHES, OF THE COMPONENTS OF THE VARIOUS GATING SYSTEMS

System	Gating Ratio	Base of Sprue	Runner	Runner Extension	Gate Entrance	Gate Exit
1	1:1:1:1:4	1/2x1	5/32x3/4	5/32x3/4	5/32x3/4	1/2x1
2	1:1:1:2:4	1/2x1	5/32x3/4	5/32x3/4	5/16x3/4	1/2x1
3	1:1:1:1:8	1/2x1	5/32x3/4	5/32x3/4	5/32x3/4	31/32x1
4	1:1:1:2:8	1/2x1	5/32x3/4	5/32x3/4	5/16x3/4	31/32x1
5	1:2:2:2:2	1/2x1	5/16x3/4	5/16x3/4	5/16x3/4	5/16x3/4
6	1:2:2:2:4	1/2x1	5/16x3/4	5/16x3/4	5/16x3/4	1/2x1
7	1:2:1:2:4	1/2x1	5/16x3/4	5/16x3/8	5/16x3/4	1/2x1

Note: Sprue length was 17 in.

Gating Ratios

Many foundrymen might prefer to use a gating ratio different from the one that was originally recommended for the vertical-gating system.[2] Therefore, the effect of different gating ratios on turbulence within the mold cavity was investigated.

A gating ratio is defined as the ratio of the cross-sectional area of the various components of the gating system. In the improved vertical gating system, the gating ratio consisted of the following area ratios:

sprue base:runner:runner extension:gate entrance:gate exit.

A ratio of 1:2:2:2:4 was previously recommended for the improved vertical-gating system shown in Fig. 1. In the present work, the following gating ratios were investigated:

1:1:1:1:4
1:1:1:1:8
1:1:1:2:4
1:1:1:2:8
1:2:2:2:2
1:2:2:2:4
1:2:1:2:4

The dimensions of the components of these various gating systems are presented in Table 1.

In the tests of the various gating systems, turbulence was measured by determining the height of the liquid jetting from the exit of the gate. The results of these tests are shown in Table 2. Examination of the data shows that the height of

TABLE 2 — EFFECT OF DIFFERENT GATING RATIOS ON THE JET HEIGHT, FLOW RATE, AND VELOCITY OF WATER IN THE RUNNER AND AT THE EXIT END OF THE GATE

System	Gating Ratio	Height of the Jet[a], in.	Flow Rate[a], lb/sec		Velocity[a], ft/sec	
			Runner	Gate	Runner	Gate
1	1:1:1:1:4	1 1/2	0.25	0.25	4.93	1.15
2	1:1:1:2:4	1 1/2	0.25	0.25	4.93	1.15
3	1:1:1:1:8	1/2	0.25	0.25	4.93	0.60
4	1:1:1:2:8	1/2	0.25	0.25	4.93	0.60
5	1:2:2:2:2	1 3/4	0.33	0.33	3.25	3.25
6	1:2:2:2:4	1/2	0.33	0.33	3.25	1.52
7	1:2:1:2:4	1/2	0.33	0.33	3.25	1.52

Note: Sprue length was 17 in.
[a] The average value of five tests.

the jet was not necessarily affected by the velocity of the water at the exit end of the gate. For instance, in System 1, the velocity of the water was 1.15 ft/sec and the jet height was 1½ in. but, in System 7, the velocity of the water was 1.52 ft/sec and the jet height was only ½ in. This may be explained on the basis of the difference in kinetic energy of the stream for the two gating systems, because the jet height is directly proportional to this kinetic energy. In a system with a 1:1 sprue: runner ratio, the liquid entering the gate has a higher velocity and, consequently, has higher kinetic energy than the liquid entering the gate of a system with a 1:2 sprue:runner ratio. One of the laws of hydraulics is that, if there is a sudden increase in the cross section of a flow channel, there is a loss in kinetic energy caused by eddies that are greater at the point of enlargement than in a straight channel. The greater the change in cross section, the greater are the energy losses. Thus, to reduce the kinetic energy to comparable values in both systems, the exit area of the gate had to be increased more in the system with a 1:1 sprue:runner ratio than in the system with a 1:2 area ratio.

To minimize jetting, the following gating ratios should be used:

1:1:1:1:8
1:1:1:2:8
1:2:1:2:4
1:2:2:2:4

With the first three gating ratios, less metal would be used in the gating system than was used in the 1:2:2:2:4 system previously recommended.[2]

Although the jet heights for Systems 1, 2, and 5 in Table 2 were about 1½ in. as compared with ½ in. for the other systems in the table, the greater jet heights caused only a slight increase in the amount of turbulence within the mold cavity. It is doubtful whether any of the gating systems listed in Table 2 created turbulence that would be sufficient to cause defects in metal castings. All the gating ratios presented here probably could be used with good results.

Screens

The first metal that enters a gating system usually becomes dirty with sand or dross as it displaces the air present in the system. This initial, undesirable liquid is trapped in the runner extension of the system shown in Fig. 1 so that it cannot enter the mold cavity. With some types of castings, it might be possible to prevent dirt and dross from entering the mold cavity by placing screens in the gating system and thus eliminating the runner extension.

The use of screens was investigated by entraining air bubbles in the water as it flowed down the sprue and observing the effectiveness of the screens in preventing the bubbles from entering the mold cavity. The screens used in this investigation were sheets of perforated metal of the type commonly employed by light-alloy foundries. Holes in the screens were 0.073 in. in diameter, and the area of the holes constituted 44 per cent of the total area of the screens.

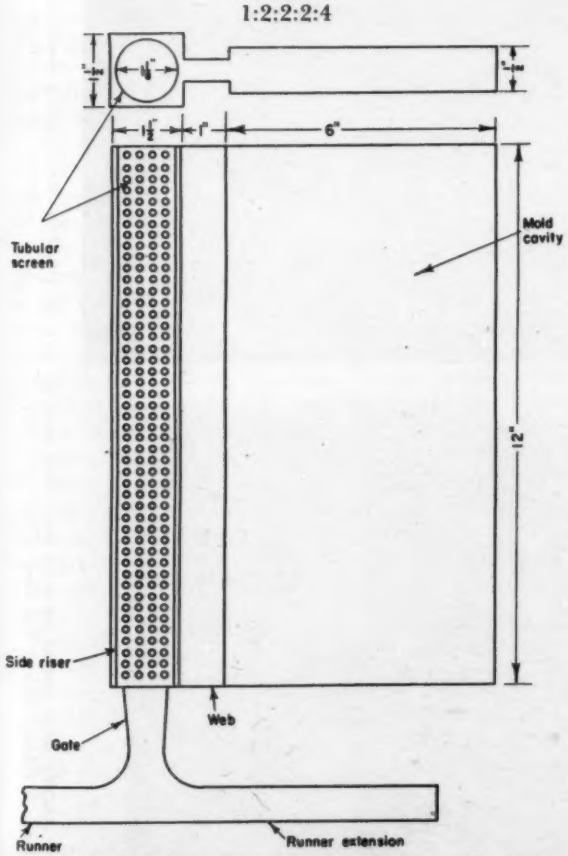


Fig. 4 — Schematic drawing showing location of tubular screen in side riser.

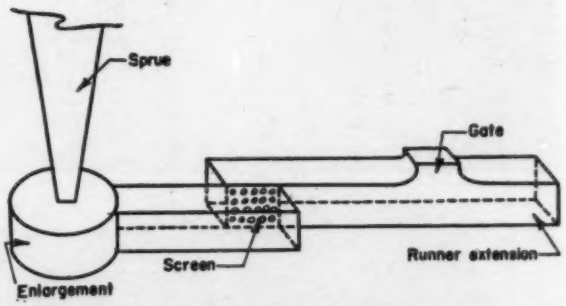


Fig. 5 — Schematic sketch showing location of screen in a horizontal offset in the runner.

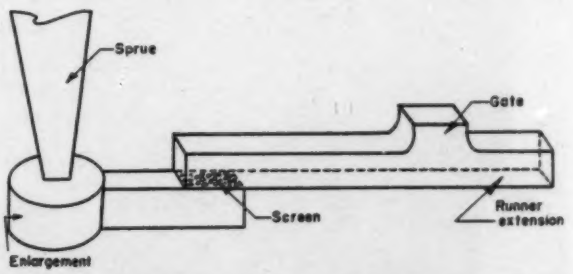


Fig. 6 — Schematic sketch showing location of screen in a vertical offset in the runner.

Fig. 7 — Al-7Mg alloy casting made by using bad pouring practice. A tubular screen was used in the side riser.

Fig. 8 — Al-7Mg alloy casting made by using bad pouring practice. No screen was used.

Fig. 9 — Al-7Mg alloy casting made by using good pouring practice. No screens were used.

The screens were tested in three different locations in the gating system:

(1) In the side riser.
(2) In a horizontal offset in the runner.
(3) In a vertical offset in the runner.

A tubular screen, 1⅛ in. in diameter, was used in the side riser (Fig. 4). This screen extended the full height of the riser so that all the liquid entering the riser had to pass through the screen before entering the mold cavity. Flat screens were used in the runner, and the total area of the holes in the flat screens was equal to the cross-sectional area of the runner. Thus, there was no restriction that would reduce the flow rate of the liquid.

The horizontal offset in the runner was made as shown in Fig. 5, and a screen was placed in a vertical position across the juncture of the two runner sections.

The vertical offset, shown in Fig. 6, was made by placing the section of the runner connected to the sprue below the section of the runner connected to the gate. The screen was placed in a horizontal position across the juncture of the two runner sections.

Tests showed that the screen in the vertical offset, Fig. 6, was not effective in retaining air bubbles. The force of the flowing liquid coupled with the buoyancy of the bubbles tended to push the bubbles through the screen and on into the mold cavity. The screen in the horizontal offset of the runner, Fig. 5, or the tubular screen in the side riser effectively retained the larger air bubbles, but small bubbles passed through the screens and entered the mold cavity.

No further tests were made with screens in the runner because they were considered impractical in the commercial foundry. However, a test was made in the foundry to determine whether a tubular screen in the side riser would effectively prevent dross from getting into an Al-7Mg alloy plate casting. To apply as severe a test as possible, the castings were poured from a ladle held with its lip about 1 ft above the pouring cup. Under these conditions, it was impossible to keep the pouring cup full and a considerable amount of dross entered the gating system.

Another mold containing no screen, but having a runner extension, was also poured in the same manner. These two castings were compared with a casting made from the same heat but poured from a ladle held close to the pouring cup so that the cup was kept full during the entire pouring operation. This third mold had a runner extension but no screen in the side riser.

Radiographs of the three castings shown in Figs. 7, 8, and 9 revealed that the two castings poured with the ladle held high contained considerable amounts of dross. The casting that was poured by conventional good practice had very little dross in it. These results indicate that, with good pouring practice, a screen is not necessary. With bad pouring practice, a considerable amount of dross will get into the casting even though a screen is placed in the side riser.

Effect of Cores on Turbulence Within Mold

In many designs of castings, the metal stream leaving the gate impinges upon a core located near the gate. To simulate these conditions in the transparent mold model and to determine the degree of turbulence that is created, the web connection between the side riser and the mold cavity was placed in the middle of the 6-in. side of the mold instead as shown in Fig. 10. The web extended the full height of the 12-in.-high mold cavity, and webs either ⅛ or ¼ in. thick were employed. The system was tested with a 17-in. and a 36-in. rectangular, tapered sprue.

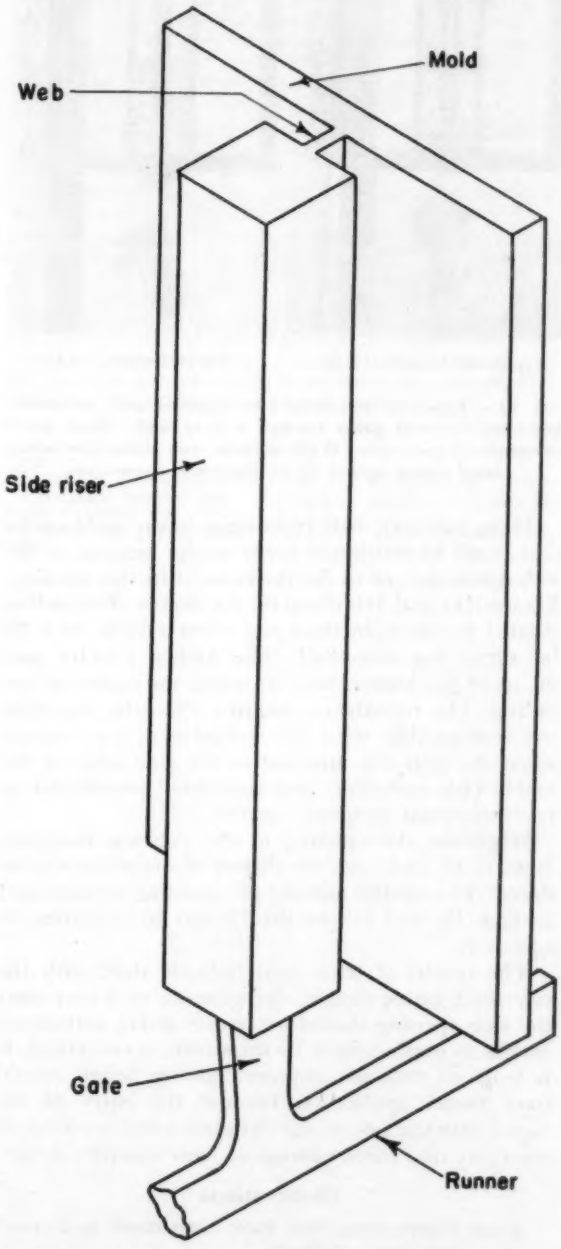
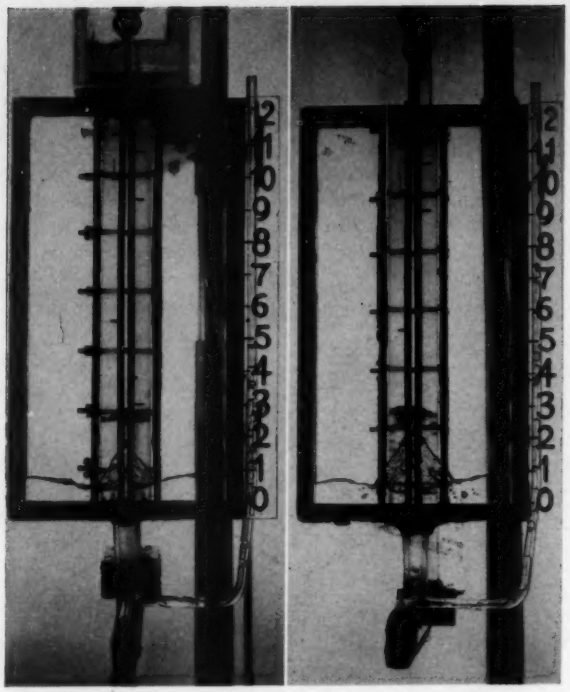
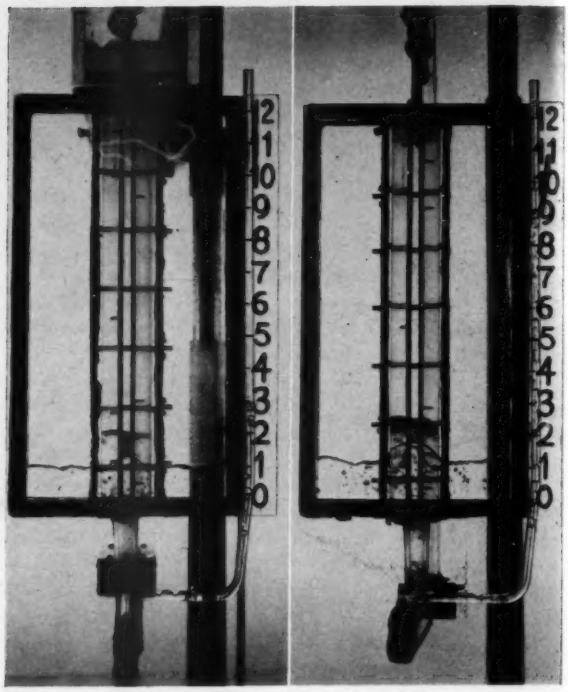


Fig. 10 — Schematic drawing showing how web was attached to side of mold to simulate effect of core opposite web.



a. Sprue Length, 17 In. b. Sprue Length, 36 In.

Fig. 11 — Effect of simulated core opposite gate on turbulence within mold gated through a ⅛-in. web. Mold gated in center of 6-in. side. Wall opposite web connection simulated a core placed ½ in. from web connection.



a. Sprue Length, 17 In. b. Sprue Length, 36 In.

Fig. 12 — Effect of simulated core opposite gate on turbulence within mold gated through a ¼-in. web. Mold gated in center of 6-in. side. Wall opposite web connection simulated a core placed ½ in. from web connection.

There was very little turbulence in the mold cavity that could be attributed either to the location of the web connection or to the thickness of its slot opening. Figures 11a and 11b illustrate the degree of cascading created by the ⅛-in.-thick slot when a 17-in. or a 36-in. sprue was employed. The higher velocity produced by the longer sprue increased the degree of cascading. The turbulence resulting from the cascading was comparable with the turbulence that existed when the web was attached to the side edge of the mold. This turbulence was considered insufficient to be detrimental to metal castings.

When the slot opening in the web was increased from ⅛ in. to ¼ in., the degree of cascading was reduced. The smaller amount of cascading is illustrated in Figs. 12a and 12b for the 17- and 36-in. sprues, respectively.

The results of these tests indicate that, with the improved gating design, the presence of a core near the web opening should not create severe turbulence insofar as impingement by the stream is concerned. It is believed that the improved gating design should have broad application because the entry of the liquid into the side of the flat-plate mold is similar to the entry into cored castings of more complex design.

Observations

Some observations that have been made as a result of this work are as follows:

1. The use of step gates does not cause an increase in turbulence. However, foundry tests with an Al-7Mg alloy showed that step gates may create undesirable thermal gradients which could result in the occurrence of shrinkage unsoundness in the vicinity of the step gates.

2. The water castings made in transparent molds showed that the gating ratio of the vertical-gating system can be modified considerably, without causing an adverse amount of turbulence within the mold cavity. However, as the ratio of the runner area to the gate exit area is reduced, the velocity of the liquid leaving the gate increases.

3. Foundry tests with an Al-7Mg alloy showed that a cylindrical screen in the side riser does not prevent dross from entering the mold cavity, if bad pouring practice is used. The water analogy showed that screens in the runner were not so effective in trapping dross as a tubular screen in a side riser.

4. A core placed opposite the web of an improved vertical-gating system does not increase turbulence within the mold cavity.

From these observations, it appears that the improved vertical-gating system should be adaptable to quite a variety of casting designs.

Acknowledgments

The authors wish to express their thanks to each member of the Research Committee of the Light

Metals Division of the American Foundrymen's Society who helped to plan the research program and to Mr. Hyman Rosenthal of the Pitman-Dunn Laboratories of Frankford Arsenal whose contribution to the planning was extremely valuable. The authors also wish to express their thanks to Dr. C. H. Lorig and J. H. Jackson of Battelle Memorial Institute for their guidance and consultation during the conduct of the investigation.

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DISCUSSION

Chairman: A. G. EALY, Littlestown Hardware & Foundry Co., Littlestown, Pa.

Co-Chairman: C. SZEGO, Mold Cast Products, Newark, N. J.

A. B. DEROSS[1]: Can the principles set forth in this progress report be applied to ring gating systems?

MR. KURA: We believe the principles set forth in this report are applicable to vertical castings employing a ring gating system. The principles undoubtedly will have to be adapted to the particular casting design. This does not detract from the principles, rather it validates them because the principles represent the starting point for the thinking that should go into the planning of the gating system for high-quality castings.

V. PASCHKIS[2]: What are the thermal gradients as referred to in the paper? What is the effect of thermal gradients on the quality of castings produced? Have attempts been made to correlate water experiments to use of actual metal considering difference in viscosity and density?

MR. KURA: We used the phrase "thermal gradients" to refer to temperature differentials that exist in a casting while the casting is solidifying. We specifically inferred that undesirable temperature differentials in the vicinity of the step gates contributed to the porous areas that are evident in the radiograph in Fig. 3. If thermal gradients are such that localized areas in a casting constitute a hot spot which freezes after the surrounding areas freeze, then shrinkage unsoundness may develop in the area of the hot spot.

With regard to Dr. Paschkis' other question, when the water-analogy technique was first considered back in 1947 for the study of principles of gating, its validity was based on the fact that the kinematic viscosity of water is similar to those of the common metals. Kinematic viscosity is obtained by dividing the density of the liquid by its dynamic viscosity. The value for the kinematic viscosity of water, aluminum, and magnesium is 1.00, 1.27, and 0.80 centistokes, respectively. Because these values fall within a narrow range, it is to be expected that water will have the steady-state flow conditions of these metals and, therefore, water may be used to study gating systems.

S. LIPSON[3]: There is a question in mind as to the analogy between air bubbles in water as in these experiments and dross particles in actual metal castings.

MR. KURA: In our experiments with screens in the gating system of water castings, air bubbles were introduced to represent dross particles in metal castings. In reply to Mr. Lipson's question, numerous types of foreign particles had been tried for this purpose, but none of them had the characteristics of dross in metal castings. The greatest deficiency was that either the smaller particles passed through the screen unhindered or the larger particles quickly and completely clogged the screen. Although air bubbles also were not the most accurate anology, they were the most convenient to use and served the purpose of indicating the relative effectiveness of screens.

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2. Technical Director, Heat and Mass Flow Analyzer Laboratory, Columbia University, New York.

3. Metallurgist, Pitman-Dunn Laboratory, Frankford Arsenal, Philadelphia.

AFS Research Progress Report on

GAS FLUSHING OF MOLTEN STEEL

By

C. H. Wyman*

A research project on gas flushing of molten steel was undertaken by the AFS Steel Division Research Committee to develop information on producing quality steel castings with fewer detrimental nonmetallic inclusions.

From preliminary studies it was found that mechanical or chemical agitation of a bath of steel either by induction steering or through chemical flushing by gases produced a cleaner steel from an inclusion and desulphurizing standpoint.

Objective of Project

The purpose of this project was to study (1) effects of introducing gases and solid deoxidation products under pressure extending same to the bottom of the ladle; (2) evolution of gases migrating and rising through the molten steel forming chemical compounds and creating levitation of undesirable inclusions to the surface.

Theoretically, from a physical chemistry standpoint the actions created by the flushing agent should produce a cleaner steel with a greater efficiency of deoxidation products resulting in lower gas contents and superior mechanical properties.

Research Project Procedure

In order to obtain authentic information from production heats, nine cooperating foundries in the Milwaukee and Chicago areas were scheduled to produce five or six heats which have been inoculated by an experienced and competent operator. A commercial gas producer who manufactures various inert gases and also has facilities for injecting the solid deoxidation products assisted in the project.

Argon, helium, polytetrafluorethylene ($C_2 F_4$), calcium carbide, sodium fluoride, and calcium fluoride were to be used experimentally as flushing agents in the project.

A commercial testing laboratory carried out chemical, physical, and metallographic tests on samples and test bars submitted by the Committee.

In June, 1955, the AFS Steel Division Research Committee made the initial test of an exploratory nature at the Burnside Steel Foundry Co. in order to develop various technical and testing procedures. The study included flushing of molten steel with gases only, and then also in combination with various special deoxidizing compounds. Two phases were covered as follows:

1. Gas injection was made in a 275-lb capacity monorail ladle;
2. Injections were made in 5-ton capacity teapot ladles;

The experimental work utilized three methods of addition:

1. A gas flush only.
2. Gas flush in combination with deoxidizer compound injection.
3. Manual deoxidation compound additions.

On all the tests made with the monorail ladles, a duplicate test block with a hand addition was made to establish a reference bar and comparative values between the hand addition and the steel which received the injection treatment.

A. For 275-lb monorail additions:

1. **Gas flush only** — Argon was injected through a 3/8-in. coated pipe for 21 to 65 seconds at a volumetric rate of 25-35 cubic feet per hour.

2. **Gas flush and deoxidizer compound** — Various fine-meshed deoxidizer alloys were gravity-fed through a 3/8-in. refractory-coated pipe in combination with argon gas. The gas rate was the same as in No. 1 above. The deoxidizer was added in quantities of 4 pounds per ton plus an addition of 0.085 per cent aluminum. These combination additions were made during a time interval of between 21-65 seconds.

3. **The manual additions** on the same basis as No. 2 were made by placing the compound on the ladle bottom and pouring the steel directly over it.

B. For full heat additions (tapping ladle) only the combination additions (gas plus deoxidizer) were made on the full heats of steel. In both monorail and

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full-heat additions aluminum at 0.085-0.125 weight per cent was added along with the deoxidizing alloy. At the outset several modifications in injection lance, gas pressure, and deoxidation compounds mesh-size, were made before the injection could be made satisfactorily. Three separate monorail heats were made in order to test the desired deoxidation. In all six such alloys were selected by the Committee (Table 1).

TABLE 1 — ADDITIONS AT 4 LB PER TON

	Per Lb Addition	Weight % Pickup
251 Grams CaMnSi		
Ca — 16.71%	0.092	0.033
Mn — 17.67%	0.098	0.036
Si — 53.00%	0.293	0.106
251 Grams CaSi		
Ca — 30.82%	0.170	0.062
Si — 62.67%	0.346	0.126
251 Grams CaSiCe		
Al — 0.86%	0.005	0.002
Ca — 22.20%	0.122	0.044
Si — 59.34%		0.119
Ce — 5.12%		0.010
Other RE — 5.88%		0.014
251 Grams CaSiAl		
Ca — 22.4%	0.124	0.045
Si — 46.25%	0.255	0.093
Al — 11.75%	0.065	0.024
C — 0.52%	0.003	0.001
251 Grams CaTiSi		
Si — 50.00%	0.276	0.099
Ca — 6.00%	0.033	0.012
Ti — 10.00%	0.055	0.020
MgFeSi		
Mg — 10.88%	0.060	0.029
Si — 14.18%	0.244	0.089
Ce — 0.53%	0.003	0.001
Al — 1.32%	0.007	0.003

For all manual additions shot aluminum was used instead of granular aluminum. Two full heats (5 tons) were made using two of the deoxidizers previously tested in the monorail set-up.

1. Calcium-aluminum-silicon plus granular aluminum plus argon.

2. Cerium-calcium-silicon plus granular aluminum plus argon.

Sufficient test bars were poured from each phase to insure adequate test coupons for all future work. All test bars were normalized at 1650 F, tempered at 1200 F and water quenched from temper to maintain uniform conditions. All test bars were then laboratory tested.

In order to maintain constant conditions all heats were made of 5-ton heats, acid process, Class B steel.

The control of cleanliness or inclusion content is probably less clearly understood than any other phase in the making of steel. Inclusions in steel resulting from deoxidation are primarily nonmetallic particles of oxides and sulphides that are precipitated from the liquidus to the solidus state. Both quantity and type of inclusions are established by many complex reactions which are controlled and influenced by the final deoxidation products. Chemical effects of deoxidation products are designed to lower the oxygen content of the liquid metal and through deoxidation to form the most desirable type of inclusions which have no detrimental effect on the ductility of the steel.

The author is a strong advocate of a high rate of bath activity in his melting practice, maintaining a high degree of oxidation, as so well exemplified by remarks taken from Walter Craft's and D. C. Hildy's unpublished paper on the significance of initial oxygen content:

> "The most disconcerting problem to the investigator of inclusion formations and a subject of greatest importance to the melter is the significance and control of initial oxygen content of the metal in the furnace prior to tap. The relative amounts of undesirable inclusion types, segregated oxide galaxies, and intergranular sulphides and related properties such as hot shortness and ductility have been observed to change with furnace practice and whereas the final combination — oxygen, sulphur, alloys and deoxidizers — may be the same in two steels, the properties may be quite different. Because the phenomena are difficult to observe and evaluate they have not received adequate attention and more extended consideration of initial oxygen content is desirable."

Another example is the ductility of steel castings in which high initial oxygen is desirable. It has long been recognized by foundrymen that a strong boil is essential to effective development of high ductility followed by strong deoxidation. Since ductility of castings is controlled primarily by intergranular sulphides, it is a reasonable assumption that the inclusion balance was shifted toward an effectively higher oxygen-sulphur ratio by over oxidation.

Source of Nonmetallic Inclusions

The origin of nonmetallic inclusions is as follows:

1. Those that occur as a result of reactions taking place in the molten or solidifying metal.

2. Those that result from mechanical erosion of refractories or other material which comes in contact with the metal, composed of oxides, sulphides and silicates; also, carbides and nitrides play an important part.

Inherent inclusions are caused by:

1. Oxidation which may occur during refining, tapping or teaming.

2. Deoxidation by alloys.

3. Change in equilibrium contents as the steel cools to freezing temperatures.

4. Decreased solubility with lower temperatures.

5. Change in concentration due to segregation during freezing with resulting saturation in the liquid portion.

Types of Inclusions

In this investigation we refer to different types of inclusions created by the various deoxidation products. These types have been classified by Sims and Dahle that result from deoxidation with various amounts of aluminum. They may be described as follows:

Type I inclusions consist of randomly distributed globular silicates and sulphides over a wide range of sizes. This type of inclusion is obtained in a silicon-

killed steel when no aluminum is used. It is characteristic of a cast steel with high ductility. The advantages associated with aluminum deoxidation such as degasification and absence of pinhole porosity, cannot be easily realized if type I inclusions are to be obtained.

Type II inclusions consist of sulphides distributed in a eutectic pattern in the primary grain boundaries with occasional clusters of alumina. The amount of aluminum required to produce this type of inclusion varies with the steelmaking practice, but it is of the order of 0.025 per cent. Type II inclusions are associated with low ductility steel.

Type III inclusions are produced by relatively high additions of aluminum (0.10-0.20 per cent). The inclusions consist of grain boundary sulphides and alumina, but the sulphides are larger and farther apart. They do not form a continuous zone of weakness, and steel containing this inclusion type is more ductile than steel containing inclusions of type II.

First Phase of the Project

The first step in the research project on flushing of steels was to make a series of three heats by introducing 1.7 oz/ton, 2.8 oz/ton, and 4.6 oz/ton of polytetrafluorethylene (C_2F_4) to a 5-ton teapot ladle of steel. The time consumed to complete the full reaction, judging by the evolution of the surface slag, varied from 2 to 3 minutes.

To obtain comparative values, two lots of test blocks were poured from each heat, one with metal direct from the receiving ladle with only the C_2F_4 treatment and the other following the regular secondary deoxidation treatment by adding 3 lb/ton of calcium-manganese-silicon plus 0.085 lb aluminum to a 275-lb monorail ladle. All of the experimental heats were made of Class B steel. Tensile bar, charpy V-notch impact specimens and gas analysis samples were obtained from the three heats shown in Table 2.

Porosity was present in all direct ladle-treated test coupons. These coupons were normalized at 1650 F and tempered at 1200 F.

Effect of Nitrogen on Charpy-V Impact Values

Plotting the values as shown in Fig. 1 by segregating the direct ladle treatment and the secondary deoxidation shows no definite correlation as to the effect of soluble and insoluble nitrogen on the impact

TABLE 2 — MECHANICAL TEST RESULTS

Heat No.	C_2F_4 Addition	Tensile Strength, psi	Yield Strength, psi	Elongation in 2 in., %	Reduction of Area, %
1631	1.7 oz/ton				
	(Ladle Treatment)	79,500	51,500	31.0	60.0
	(Secondary Deoxidation)	81,000	53,000	28.0	47.8
1634	2.8 oz/ton				
	(Ladle Treatment)	72,000	49,000	Flaw	
	(Secondary Deoxidation)	78,000	51,000	31.0	56.8
1640	4.6 oz/ton				
	(Ladle Treatment)	72,000	45,000	31.0	60.0
	(Secondary Deoxidation)	73,000	49,000	31.0	56.5

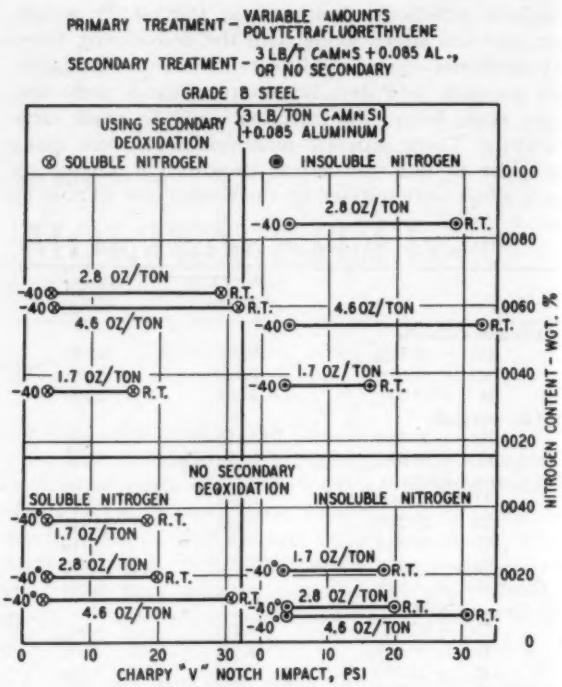


Fig. 1 — Graphs showing effect of nitrogen on charpy-V impact on full heats of Class B steel. Primary treatment consisted of addition of variable amounts of polytetrafluorethylene. Secondary treatment consisted of addition of 3 lb/ton Ca-Mn-Si plus 0.085 lb Al.

values at room temperature and -40 F. The 4.6-oz addition shows the highest impact value with the 2.8-oz addition rated second, and the 1.7-oz addition a poor third. In the case of the secondary deoxidation, the 1.7-oz addition had the lowest soluble and insoluble nitrogen content and on the ladle treatment deoxidation the 4.6-oz addition showed the lowest contents with the 1.7-oz addition being the highest. Table 3 shows the gas analysis determinations. In all cases the hydrogen is less than 0.0001 per cent.

One outstanding point on the gas analysis is that it showed a definite decrease in the insoluble nitrogen content, based upon the weight of the addition of the C_2F_4 from a figure of 0.002 to 0.0008 per cent.

TABLE 3 — RESULTS OF GAS ANALYSIS

Heat No.	Addition	Oxygen, %	Soluble Nitrogen, %	Insoluble Nitrogen, %	Total, %
1631	1.7 oz/ton (No Secondary Deoxidation)	0.016	0.0036	0.0021	0.0057
	(Secondary Deoxidation)	0.016	0.0034	0.0026	0.0060
1634	2.8 oz/ton (No Secondary Deoxidation)	0.017	0.0084	0.0010	0.0094
	(Secondary Deoxidation)	0.019	0.0064	0.0019	0.0083
1640	4.6 oz/ton (No Secondary Deoxidation)	0.017	0.0054	0.0008	0.0062
	(Secondary Deoxidation)	0.018	0.0060	0.0013	0.0073

The gas determinations were checked by two different laboratories with results consistently within a narrow range.

To prove the effect of the desulphizing powers of the $C_2 F_4$, samples were procured from the furnace just before the tapping operation and sulphur determinations were made, showing a sulphur level from 0.033 to 0.035 per cent. The finished sulphur in the heat, based upon the addition, showed a sulphur reduction from 0.003 to 0.007 per cent respectively with the various additions of the $C_2 F_4$ from 1.7 oz to 4.6 oz per ton.

Second Phase of Investigation

Following the straight gas flushing experimental heats, the authors proceeded to carry through the second phase of the investigation by flushing argon gas through the injection dispensing unit holding the solid deoxidation products. This unit shown in Fig. 2 was furnished by a commercial gas company.



Fig. 2 — Injection dispensing unit.
(Courtesy National Cylinder Gas Co.)

While injecting the 275-lb ladles, the unit was handled manually and the flushing time and flushing rate were recorded. Through this manual operation the author found it was difficult to maintain a consistent flow of deoxidation products through the unit and the refractory-coated lance. He also found it necessary, in order to maintain a definite degree of evolution in the ladle, that the material had to be of a consistent granular size. In the process of flushing the full-size heats, the experiment was perfectly timed and all discontinuity was practically eliminated from a flushing standpoint. After tapping the heat, the author set the ladle in the bottom of the pouring pit and had the crane lift the dispensing unit which was immediately immersed in the ladle with a 6-ft refractory-coated lance. From all observations the flushing reached the desired time with excellent surface evolution.

Figure 3 shows ultimate strength and per cent of reduction of area versus the type and method of deoxidation additions. On Class B steel the ultimate strength fell within the required range between 73,000 and 78,000 psi with a yield strength from

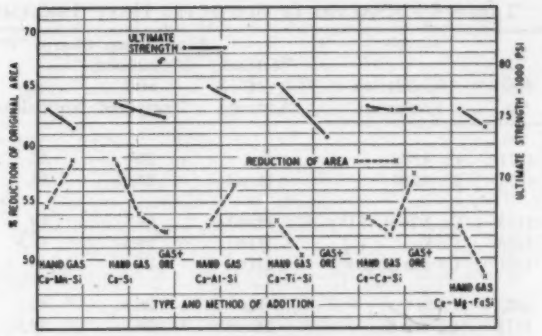


Fig. 3 — Comparison of tensile strength and reduction of area of Class B steel using various types of deoxidizers and methods of addition.

48,000 to 52,000 psi. On the reduction of area employing the hand addition, gas flushing with gas, ore and the solid deoxidation products, the reduction of area ranged between 42 to 59 per cent with the exception of one specimen where a flaw was evident in the fractured area.

Full chemistry analysis by the wet and spectrographic methods were made on the five heats and 20 coupons involved to determine the common elements, residuals and tramp elements. They were as follows:

Common Elements, %	
Carbon	0.21 — 0.27
Manganese	0.59 — 0.71
Silicon	0.46 — 0.60
Phosphorus	0.021 — 0.024
Sulphur	0.035 — 0.040
Residual Elements, %	
Chrome	0.12 — 0.22
Nickel	0.06 — 0.08
Molybdenum	0.01
Vanadium	0.01
Titanium	0.002 — 0.026
Aluminum	Nil — 0.064
Tramp Elements, %	
Zinc	0.004 — 0.02
Lead	0.001 — 0.002
Tin	0.01

Figure 4 is a comparison of charpy V-notch impact values at room temperature and –40 F with various deoxidizers and methods of additions. All of

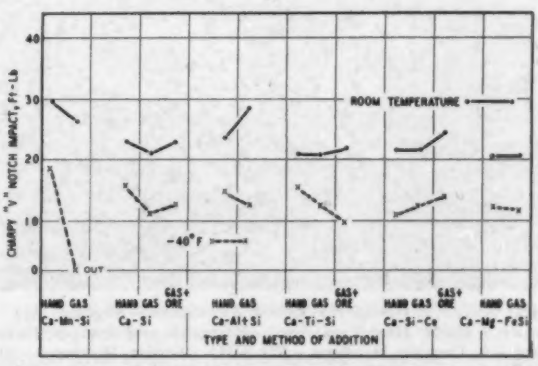


Fig. 4 — Comparison of charpy-V notch impact values of Class B steel at –40 F and room temperature using various type deoxidizers and methods of additions.

TABLE 4 — ANALYSIS OF SUBJECTED HEAT TREATMENT AND EFFECT ON IMPACT VALUES AND MICRO HARDNESS

Heat No.	Deoxidation Treatment	Method of Addition	Room Temp.-Charpy "V" Normalized and Tempered	Room Temp.-Charpy "V" Normalized	Tukon-Knoop Normalized and Tempered Carbide	Tukon-Knoop Normalized and Tempered Ferrite	Hardness, Normalized Carbide	Hardness, Normalized Ferrite	Impact % Increase	Grain Size
										N - 8
1071	Ca-Al-Si	Gas	29	55	323	227	373	184	47	NT-7-8
1071	Ca-Al-Si	Hand	24	53	327	223	347	176	55	N - 8
										NT-7-8
1113	Ca-Mn-Si (3 lb)	Hand	24.5	49					50	NT-7-8
1068	Ca-Si	Hand	23.5	48.5					52	NT - 8
1071	Ca-Mn-Si (3 lb)	Hand	25	46	324	193	358	193	46	N - 7
										NT - 8
1068	Ca-Si	Gas	26.5	45					41	NT - 8
1113	Ca-Mn-Si	Hand	26	43.5					40	NT-7-8
1113	Ca-Mn-Si	Gas	27	39.5					32	NT-5-7
1068	Ca-Mn-Si (3 lb)	Gas	29.5	37.5					22	NT-6-7

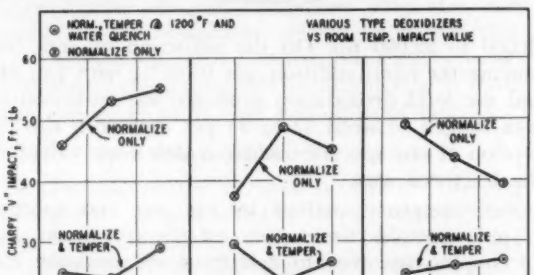
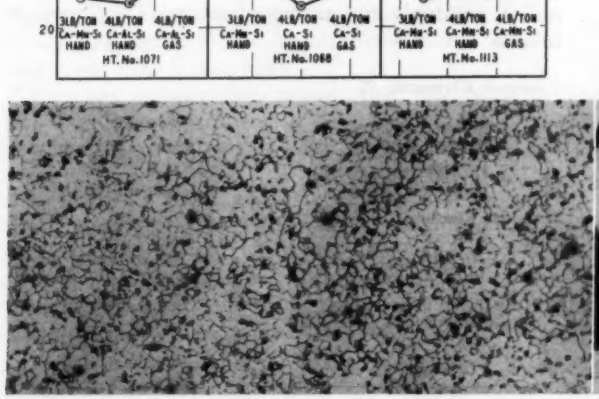


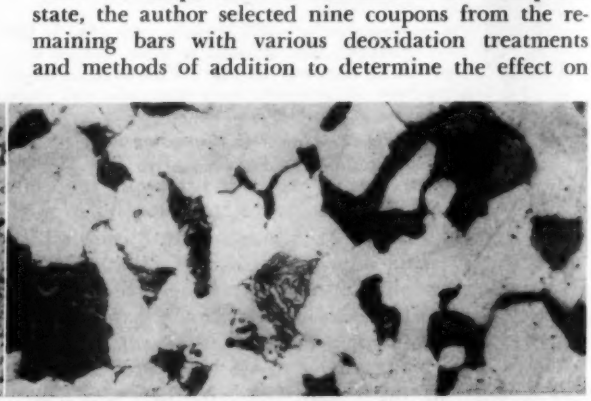
Fig. 5 (Left) — Relationship of various type deoxidizers and room temperature impact values.

these coupons were in the normalized and tempered state. Test values fell below the range which we consider normal for this type of steel. The foot-pounds reading at room temperature ranged from 21 to 30 and at –40 F they ranged from 10 to 18.

In view of the low impact values obtained from these test coupons in the normalized and tempered state, the author selected nine coupons from the remaining bars with various deoxidation treatments and methods of addition to determine the effect on

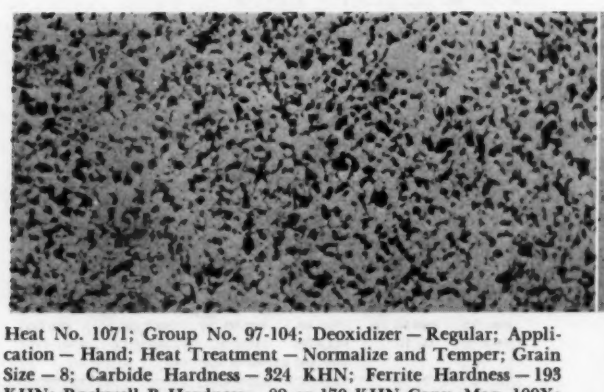


Heat No. 1071; Group No. 97-104; Deoxidizer — Regular; Application — Hand; Heat Treatment — Normalize; Grain Size — 7; Carbide Hardness — 358 KHN; Ferrite Hardness — 143 KHN; Rockwell B Hardness — 83 or 173 KHN Conv. Mag. — 100X; Etch — 2% Nital.

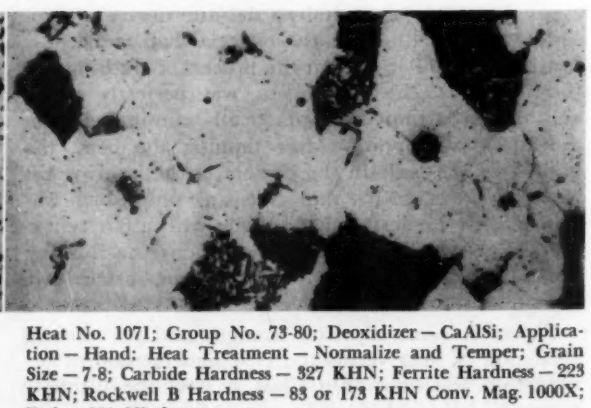


Heat No. 1071; Group No. 73-80; Deoxidizer — CaAlSi; Application — Hand; Heat Treatment — Normalize; Grain Size — 8; Carbide Hardness — 347 KHN; Ferrite Hardness — 176 KHN; Rockwell B Hardness — 81 or 167 KHN Conv. Mag. — 1000X; Etch — 2% Nital.

Fig. 6 (A-Left, B-Right) — Photomicrographs of samples from Heat No. 1071.



Heat No. 1071; Group No. 97-104; Deoxidizer — Regular; Application — Hand; Heat Treatment — Normalize and Temper; Grain Size — 8; Carbide Hardness — 324 KHN; Ferrite Hardness — 193 KHN; Rockwell B Hardness — 82 or 170 KHN Conv. Mag. 100X; Etch — 2% Nital.



Heat No. 1071; Group No. 73-80; Deoxidizer — CaAlSi; Application — Hand; Heat Treatment — Normalize and Temper; Grain Size — 7-8; Carbide Hardness — 327 KHN; Ferrite Hardness — 223 KHN; Rockwell B Hardness — 83 or 173 KHN Conv. Mag. 1000X; Etch — 2% Nital.

Fig. 7 (A-Left, B-Right) — Photomicrographs of samples from Heat No. 1071.

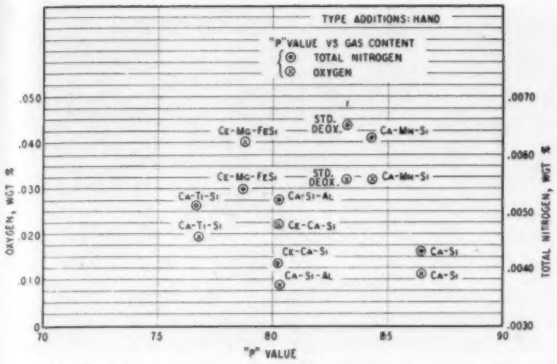


Fig. 8 — Effect of hand addition on oxygen and total nitrogen content against P value.

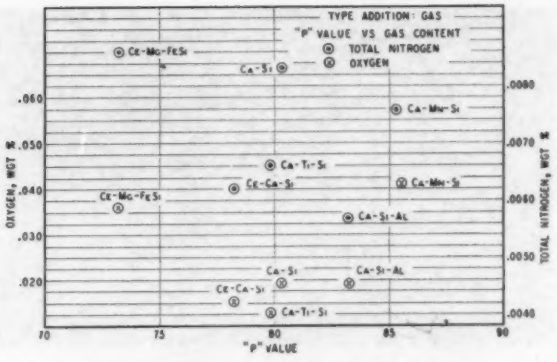


Fig. 9 — Relationship of P value versus gas content.

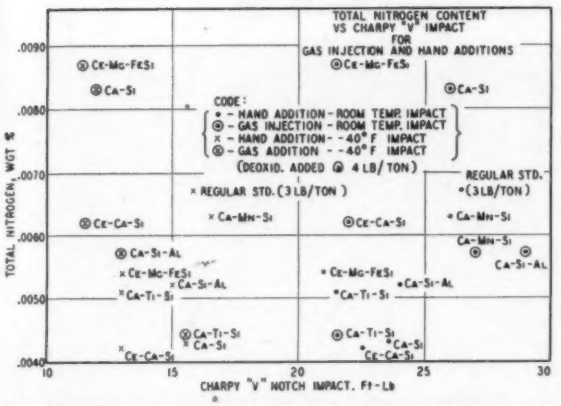


Fig. 10 — Relationship of total nitrogen content versus charpy-V impact values.

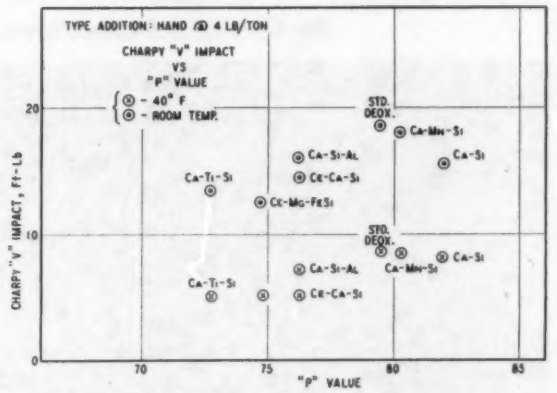


Fig. 11 — Relationship of charpy-V impact versus P values.

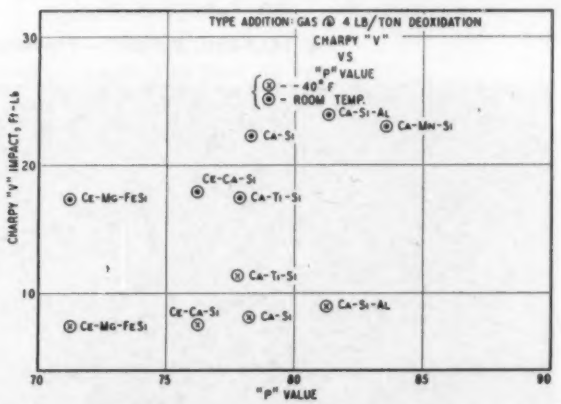


Fig. 12 — Relationship of charpy-V impact versus P values.

the impact values in the normalized state versus the normalized and tempered treatment. Following the same procedure, the author normalized these coupons in a laboratory furnace at 1650 F for 2 hours.

Table 4 shows the analysis of the heat treatment and effect on impact values and microhardness.

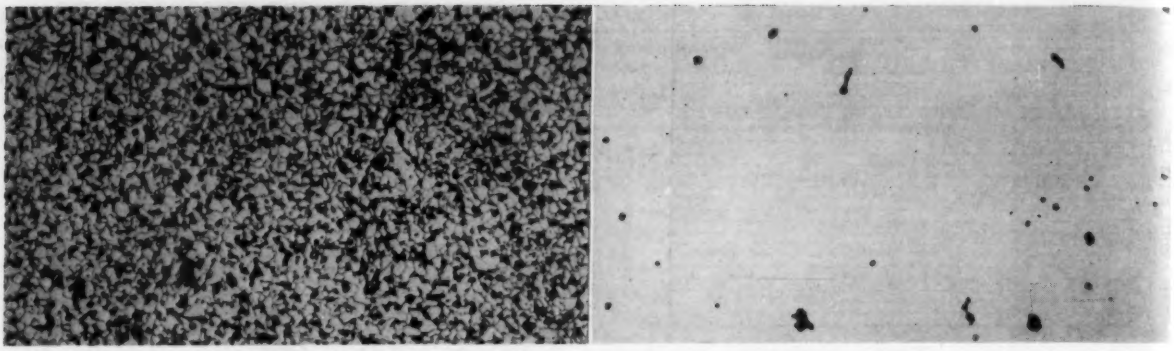
Figure 5 shows various type deoxidizers versus room temperature impact values in both the normalized and normalized and tempered state. The per cent increase in impact values in the normalized state ranged from 22 to 55 per cent, which from past experience is normal in the author's line of production.

In analyzing Table 4, it is the author's opinion, which cannot be substantiated without extensive research work, that the lower impact strengths of the normalized and tempered material are primarily due to an embrittled condition of the ferrite by a precipitation of nitrides and partial spheroidizing of the carbides during the tempering cycle. There is also indication that in the normalized structure the higher hardness pearlitic carbide areas are surrounded by a ductile lower hardness ferrite matrix which offers more resistance to impact resulting in higher notch values.

Metallographic Study of Specimens

To further substantiate our theory, a metallographic study was made of the author's regular hand method addition specimens in the normalized and normalized and tempered state. Although the grain size is 7 in the normalized state (Fig. 6), in the normalized and tempered state (Fig. 7) after studying the structure it clearly shows a finer dispersement of the pearlitic areas.

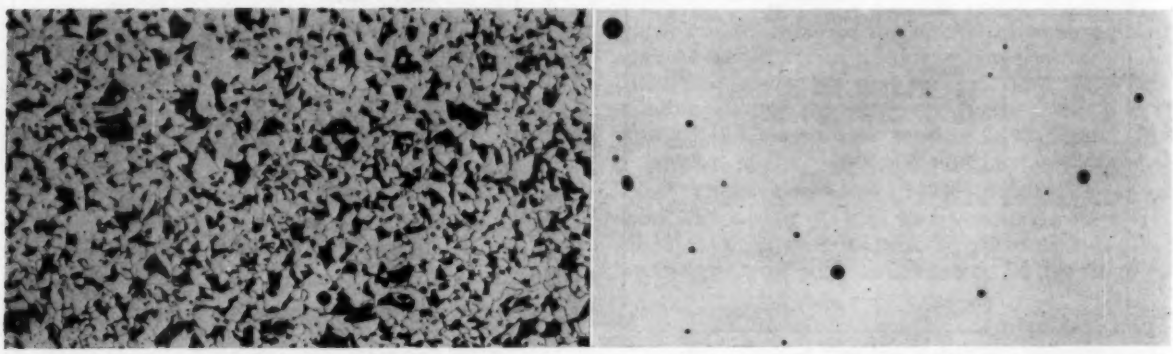
In order to establish a correlation between the influence or effect on the gaseous elements, Fig. 8 was plotted showing the effect of the hand addition on oxygen and total nitrogen against P value. P value is a strength-ductility merit index which has been used in previous investigations. The points are well



Heat No. 1068; Group No. 41-44; Deoxidizer — Al-CaSi; Method of Application — Hand; Grain Size — 8; Treated in 300-lb Monorail Ladle. Mag. — 100X; Etch — Nital, Picral.

Heat No. 1068; Group No. 41-44; Deoxidizer — Al-CaSi; Method of Application — Hand; Inclusion Type III; Count, 4; Treated in 300-lb Monorail Ladle. Mag. — 250X; Etch — None.

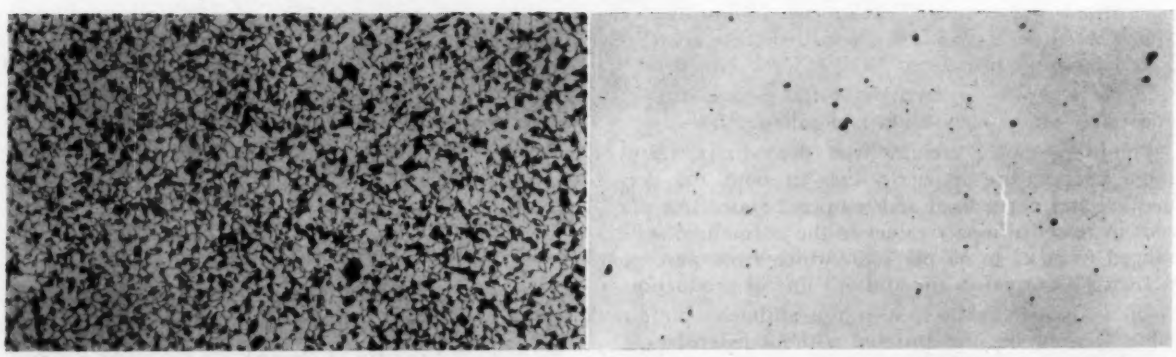
Fig. 13 (A-Left, B-Right) — Photomicrographs of samples from Heat No. 1068.



Heat No. 1068; Group No. 57-64; Deoxidizer — Al-Regular; Method of Application — Hand; Grain Size — 6-7; Treated in 300-lb Monorail Ladle. Mag. — 100X; Etch — Nital, Picral.

Heat No. 1068; Group No. 57-64; Deoxidizer — Al-Regular; Method of Application — Hand; Inclusion Type I; Count, 3; Treated in 300-lb Monorail Ladle. Mag. 250X; Etch — None.

Fig. 14 (A-Left, B-Right) — Photomicrographs of samples from Heat No. 1068.



Heat No. 1071; Group No. 81-88; Deoxidizer — Al-CaAlSi; Method of Application — Gas; Grain Size — 8; Treated in 300-lb Monorail Ladle. Mag. — 100X; Etch — Nital, Picral.

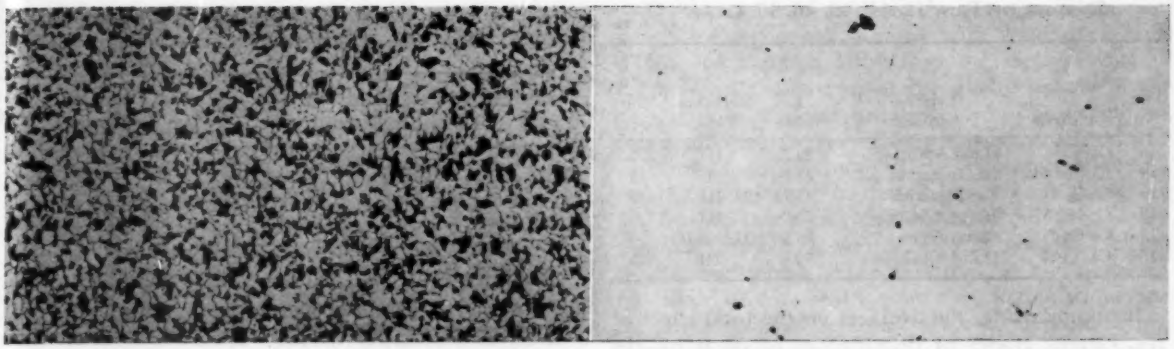
Heat No. 1071; Group No. 81-88; Deoxidizer — Al-CaAlSi; Method of Application — Gas; Inclusion Type III; Count, 5; Treated in 300-lb Monorail Ladle. Mag. — 250X; Etch — None.

Fig. 15 (A-Left, B-Right) — Photomicrographs of samples from Heat No. 1071.

scattered showing no correlation or effect of the gas contents on the P value.

The same procedure was followed in Fig. 9 on gas injection by plotting the oxygen and total nitrogen versus P value, again showing no definite trend or correlation with the points more widely scattered than in the hand method of addition.

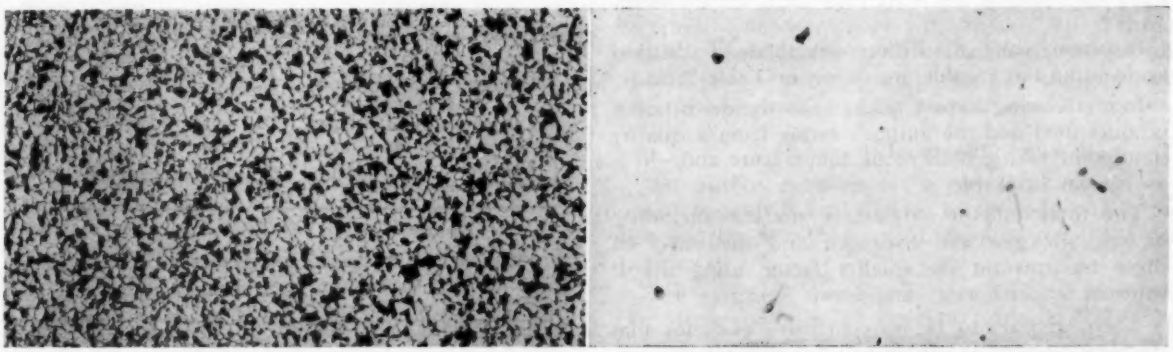
To further substantiate the effect of total nitrogen on the impact values of steel, Fig. 10 was drawn to include both gas injection and hand additions. It is apparent that again the points show no definite trend as to the effect of the total nitrogen on the charpy V-notch impact values basing them on the various types of deoxidizing agents. The gas analysis methods by vacuum fusion were checked by two laboratories. Analysis was made approximately 1½



Heat No. 1071; Group No. 89-96; Deoxidizer — Al-CeMgFeSi; Method of Application — Hand; Grain Size — 8; Treated in 300-lb Monorail Ladle. Mag. — 100X; Etch — Nital, Picral.

Heat No. 1071; Group No. 89-96; Deoxidizer — Al-CeMgFeSi; Method of Application — Hand; Inclusion Type III; Count, 3; Treated in 300-lb Monorail Ladle. Mag. — 250X; Etch — None.

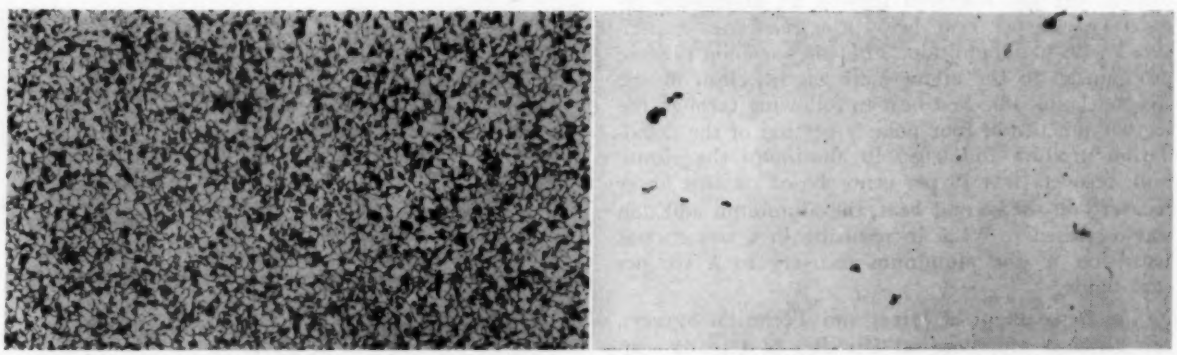
Fig. 16 (A-Left, B-Right) — Photomicrographs of samples from Heat No. 1071.



Heat No. 1071; Group No. 45-48; Deoxidizer — Al-CeCaSi; Method of Application — Gas-Ore; Grain Size — 8; Treated in 300-lb Monorail Ladle. Mag. — 100X; Etch — Nital, Picral.

Heat No. 1071; Group No. 45-48; Deoxidizer — Al-CeCaSi; Method of Application — Gas-Ore; Inclusion Type — II, III; Count, 4; Treated in 300-lb Monorail Ladle. Mag. — 250X; Etch — None.

Fig. 17 (A-Left, B-Right) — Photomicrographs of samples from Heat No. 1071.



Heat No. 1113; Group No. 1-8; Deoxidizer — Al-SiTiCa; Method of Application — Hand; Grain Size — 8; Treated in 300-lb Monorail Ladle. Mag. — 100X; Etch — Nital, Picral.

Heat No. 1113; Group No. 1-8; Deoxidizer — Al-SiTiCa; Method of Application — Hand; Inclusion Type III; Count, 3; Treated in 300-lb Monorail Ladle. Mag. — 250X; Etch — None.

Fig. 18 (A-Left, B-Right) — Photomicrographs of samples from Heat No. 1113.

months after casting of specimens.

Figure 11 was drawn to determine if there was a relationship between charpy impact strength and P value. These results were taken from the hand addition at room temperature and –40 F. There is a slight indication that the impact values show a trend toward the higher strength and ductility factor.

This was also repeated in Fig. 12 with the gas addition and the same deoxidation products. There is a wider distribution with a definite trend indicating lower values.

Typical photomicrographs showing structure and the type inclusion and count, based on the varying degrees of P value are shown in Figs. 13-18.

RESUME OF FIGS. 13, 14, 15, 16, 17 AND 18

Fig. No.	Additive	Type of Addition	P Value	Inclusion Type	ASTM Inclusion Count
13	Ca-Si	Hand Addition	86.7	III	4
14	Ca-Mn-Si	Gas Injection	85.7	I	3
15	Ca-Si-Al	Gas Injection	83.78	III	5
16	Ce-Mg-Si	Hand Addition	79.7	III	3
17	Ce-Ca-Si	Gas and Ore	85.5	II and III	4
18	Ca-Ti-Si	Hand Addition	79.8	III	3

In summarizing, the averages on the total effect of the various grades of deoxidation products on the quality basis using the P value as a merit index are shown in Table 5.

A further step was taken of breaking down the results to the method of addition from a quality standpoint again using the P value as the quality factor (Table 6).

In summarizing, the different methods of addition again using the P value are shown in Table 7.

In establishing impact values as to the deoxidation product used and the author's rating from a quality standpoint taking both room temperature and –40 F are shown in Table 8.

The total amount of gaseous products including oxygen, nitrogen and hydrogen and the effect of these amounts on the quality factor using the P value as a merit index are shown in Table 9.

There appears to be no conclusive evidence that the total amount of gas content has any definite correlation on the quality factor.

Practically in all cases the aluminum recovery was at a lower level with gas injection as compared with hand additions. In averaging the total aluminum content the gas injection showed an aluminum recovery of 25 per cent below the recoveries experienced with hand addition. This phenomenon is more pronounced in the argon inert gas injection in the five-ton ladle. The first heat in following through the regular practice of four pounds per ton of the deoxidation product and 0.085 lb aluminum the aluminum recovery was 19 per cent. Based on this lower recovery on the second heat, the aluminum addition was increased to 0.125 lb resulting in a tremendous reduction of the aluminum recovery to a six per cent figure.

The Department of Mines and Technical Surveys, the Physical Metallurgical Division of Canada, carried out a similar project on desulphurizing of steels. Their comments and constructive criticism on the AFS project were as follows:

"This type of research is somewhat complex due to the variables and technicalities involved. We consequently feel that the Committee has done as good work as far as they went. The conclusions based upon the actual limited number of tests carried out appear to be valid. More data are required, however, to confirm these conclusions. There are a number of items in connection with the work that need further explanation, for example:

TABLE 5 – SUMMARY OF DEOXIDANTS VS. QUALITY OF EXPERIMENTAL HEATS

Additive	P Value Average	Type Inclusion	ASTM Inclusion Count
1. Calcium-Manganese-Silicon	84.87	1 & 3	3 & 4
2. Calcium-Manganese-Silicon (Regular Addition)	83.26	1 & 3	3
3. Calcium-Silicon	81.76	2 & 3	4 & 5
4. Calcium-Silicon-Aluminum	81.51	2 & 3	2 to 5
5. Calcium-Titanium-Silicon	78.35	2 & 3	3 & 4
6. Cerium-Calcium-Silicon	77.47	3 & 4	3
7. Cerium-Magnesium-Ferrosilicon	76.03	2 & 3	3 & 4

TABLE 6 – SUMMARY OF DEOXIDANTS VS. P VALUE OF EXPERIMENTAL HEATS WITH VARIOUS ADDITION METHODS

Additive	Addition Method Hand	Gas	Gas Ore
1. Calcium Manganese-Silicon	84.3	85.5	
2. Calcium-Silicon	86.5	80.4	78.1
3. Calcium-Manganese-Silicon (Regular)	83.3		
4. Calcium-Silicon-Aluminum	80.3	83.2	
5. Calcium-Titanium-Silicon	76.8	79.9	Failure
6. Cerium-Calcium-Silicon	80.3	78.3	84.9
7. Cerium-Magnesium-Ferrosilicon	78.8	73.3	

TABLE 7 – SUMMARY OF METHODS OF ADDITION AND QUALITY FACTOR

Method of Addition	No. of Tests	Average P Value
1. Hand Addition	18	82.26
2. Gas Injection	12	79.59
3. Gas Ore	4	81.68
Gas Ore	2 (Failure)	64.68
4. Injection Full Heats	2	73.65

TABLE 8 – RATING OF DEOXIDATION PRODUCTS FROM QUALITY STANDPOINT

1. Calcium-Manganese-Silicon, Regular
2. Calcium-Silicon Aluminum
3. Calcium-Manganese-Silicon
4. Calcium-Silicon
5. Calcium-Silicon
5. Cerium-Calcium-Silicon
6. Calcium-Titanium-Silicon
7. Cerium-Magnesium-Ferrosilicon

TABLE 9 – QUALITY FACTOR VS. GAS CONTENT IN STEEL

	P Value	Total Gas Content
1.	86	0.0154
2.	85	0.0217
		0.0468
3.	83	0.0249
		0.0384
4.	81	0.0193
		0.0353
		0.051
5.	80	0.0143
		0.0165
		0.0276
		0.0294

1. Why the lower aluminum content when gas flush is used? Could this be due to oxygen or moisture in the argon gas?

2. Two of the tests show no aluminum to be present. One has very good low temperature impact 18 to 20 foot-pounds. The other has a low temperature impact of 2 to 3 foot-pounds. The difference between the two is that one is a hand deoxidation procedure whereas the other has been lanced. It is definitely felt that the low-temperature impact results should be checked since all of the other properties appear to be good.

3. Along with lower residual aluminum Type 2 grain boundary sulphides occur mainly in the injected samples. Has any thought been given to the length of time the ladle has been held before pouring in the molds? A variation of this time could have an influence on the state of deoxidation of the steel and on the inclusion count. Since it is felt that not enough tests were carried out for each deoxidant to enable more positive conclusions to be drawn, we would suggest the following minimum program to be carried out:

   a. That the best deoxidant be given more thorough tests. On the work already done, we think perhaps the calcium-manganese-silicon reagent would be the most desirable.

   b. Three or more tests should be carried out on this one reagent comparing it to the standard each time so that for the conditions used it could definitely be established whether or not the injection method is of any value at all. We also feel that a blank should be run wherein the aluminum standard is lanced with argon by itself to determine the effect of the argon."

In view of the above comments and suggestions, the members of the AFS Steel Division Research Committee have agreed to carry on further investigation to evaluate the validity of flushing full heats of a more sensitive steel

The program as outlined by the Committee will consist of argon gas flushing with solid deoxidants on six low-alloy steels by injecting the receiving or tapping ladle. Based upon the quality values of the deoxidation products on the previous investigation, it was decided to use calcium-manganese-silicon and calcium-silicon-aluminum at the rate of four pounds per ton by injection of three heats of each deoxidant. To further investigate the influence or effect of the flushing of steels, it was decided to pour test blocks of the base metal before flushing, an additional series after flushing with a primary deoxidation, and the third series of test blocks are to be cast with a secondary deoxidation.

Following each step, mechanical and chemical properties will be evaluated by spectrographic analysis which will include residuals and tramp elements. Duplicate tensile bars will be made in the heat treated state from each step and a transition curve on charpy impact values will be established from room temperature to –40 F. The investigation will also include a metallographic study of the structure and close observation of the type inclusion and count.

Acknowledgment

The author expresses his appreciation for the cooperation and efforts extended by the following members of the AFS Research Committee of the Steel Division:

H. H. Blosjo, Minneapolis Electric Steel Castings Co., Minneapolis; W. O. Igelman, National Malleable & Steel Castings Co., Melrose Park, Ill.; Dr. C. H. Lorig, Battelle Institute, Columbus, Ohio; W. R. Punko, Wehr Steel Castings Co., Milwaukee; J. A. Rassenfoss, American Steel Foundries, E. Chicago, Ind.; and V. E. Zang, *Chairman*, AFS Steel Division.

The author also acknowledges his appreciation for the interest and cooperation of the Steel City Testing and Engineering Laboratories, Chicago and the National Cylinder Gas Co., Chicago in this research project.

EFFECTS OF CHARGE PROPORTIONS, FURNACE ATMOSPHERE FLOW RATE, AND MELT-DOWN TIME ON PROPERTIES OF MALLEABLE IRON

By

A. H. Zrimsek,* E. H. Belter,** and R. W. Heine**

This annual report of progress on the AFS malleable iron research project deals with the following areas of investigation:

1. Effect of steel in the charge on chemistry of melting, casting properties, annealability, and tensile properties using the furnace atmosphere flow rates employed in previous experiments.
2. Effect of flow rate of melting furnace atmosphere on the above-mentioned properties when melting a charge containing steel.
3. Effect of melt-down time and melting rate on the properties mentioned in (1).

Procedure

One-hundred-pound heats of iron were produced following practices described in previous annual progress reports. When investigating the effects of steel, the metal charge consisted of 50 per cent duplex No. 2 sprue, 37.5 per cent SAE 1010 steel, and 12.5 per cent pig iron. Chemical compositions of charge materials are given in Table 1. In heat CU 50 per cent steel was used while in other heats the charge was 100 per cent sprue. Low-aluminum ferrosilicon and ferromanganese were used to adjust silicon and manganese.

In mixed charges, the steel portion was in the form of washers 0.125 in. thick x 2.25 in. OD x 1.50 in. ID, and disks 0.125 in. thick x 1.50 in. in diameter. The pig iron was a malleable grade obtained from a commercial foundry. Three different types of malleable sprue, as listed in Table 1 were used in separate experiments. The melting conditions for each heat are listed in Table 2. The normal melting schedule consisted of melting down to 2600 F in 70 to 90 minutes, heating to 2800 F in total of 1½ to 2 hr, holding at 2800 F for 1 hr for a total melting period of 2½ to 3 hr. The furnace atmosphere inlet tube was initially located close to the bottom of the furnace so that the gas would flow up through the cold metal charge. Gas flow rates are given in Table 2. As melting proceeded the tube was raised above the level of the rising metal bath. All atmospheres were applied both during the melting down and holding periods of melting.

TABLE 1 – CHEMICAL COMPOSITIONS OF MATERIALS IN THE CHARGE

Material	% C	% Si	% Mn	% S	% P	Average Surface Area per Pound, sq in./lb
Steel, SAE 1010	0.06	0.08	0.40	0.030	0.040	63.6
Sprue, Duplex No. 1	2.54	1.30	0.41	0.146	0.110	13.3
Duplex No. 2	2.60	1.15	0.47	0.150	0.120	11.9
Cold Melt	2.48	1.10	0.37	0.085	0.126	11.9
Pig Iron	3.70	1.80	0.75	0.030	0.100	9.2

Spectrographic analysis of the pig iron revealed the following residual elements present: 0.06% Ti; present but probably less than 0.005% B, 0.01% Al, and 0.02% Zr.

Information pertinent to the final chemical composition and casting properties of the heat is given in Table 3. Silicon oxidation losses and pertinent melting rates for heats where steel was employed in the charge are given in Table 4.

The different gas flow rates listed in Tables 2 and 4 were applied during the melting down stage. Upon melting down, the flow rate was set at the standard rate of 0.78 cu ft per min employed in previous heats. This was done to avoid excessive carbon losses during the 1 hr holding period at 2800 F.

*Formerly project assistant, currently with American Steel Foundries, Chicago, Ill.

**Malleable Project Assistant, and Associate Professor of Metallurgical Engineering, respectively, Dept. of Mining and Metallurgy, University of Wisconsin, Madison, Wisc.

This is the fifth progress report on a research project initiated and sponsored by the Research Committee of the Malleable Division of the American Foundrymen's Society. Members of this Committee are as follows: C. F. Joseph, *Chairman*, W. D. McMillan, *Vice-Chairman*, H. Bornstein, W. A. Kennedy, J. H. Lansing, R. V. Osborne, R. P. Schauss, R. Schneidewind, Milton Tilley, and P. F. Ulmer.

Table 2 – Melting Conditions

Heat	Atmosphere	Atmosphere Flow Rate, cu ft/min	Minutes Time of Meltdown	Wt. Ratio Pig/Sprue/ Steel	Area Ratio Sq In. Pig/Sprue/ Steel	Sprue Charged
CJ	21% CO_2 79% N_2	0.78	100	12.5/50/37.5	155/537/2385	Duplex No. 2
CK	21% CO_2 79% N_2	0.78	87	12.5/50/37.5	83/616/2387	Duplex No. 2
CL	21% CO_2 79% N_2	0.78	85	12.5/50/37.5	95/570/2387	Duplex No. 2
CM	21% CO_2 79% N_2	0.78	85	12.5/50/37.5	115/610/2387	Duplex No. 2
CN	21% CO_2 79% N_2	0.78	80	12.5/50/37.5	102/600/2387	Duplex No. 2
CO	21% CO_2 79% N_2	1.80	80	12.5/50/37.5	105/580/2387	Duplex No. 2
CP	21% CO_2 79% N_2	3.0	80	12.5/50/37.5	112/602/2387	Duplex No. 2
CR	21% CO_2 79% N_2	3.0	80	12.5/50/37.5	109/615/2387	Duplex No. 2
CS	21% CO_2 79% N_2	5.3	85	12.5/50/37.5	132/603/2387	Duplex No. 2
CT	100% N_2	5.3	80	12.5/50/37.5	140/612/2387	Duplex No. 2
CU	21% CO_2 79% N_2	3.0	85	0/50/50	0/640/3182	Duplex No. 2
CV	21% CO_2 79% N_2	0.78	180	12.5/50/37.5	146/590/2387	Duplex No. 2
CW	21% CO_2 79% N_2	0.78	180	12.5/50/37.5	102/600/2387	Duplex No. 2
CX	21% CO_2 79% N_2	0.78	130	12.5/50/37.5	118/580/2387	Duplex No. 2
CY	Air	0.78	80	12.5/50/37.5	92/580/2387	Duplex No. 2
CZ	Air	0.78	80	12.5/50/37.5	116/580/2387	Duplex No. 2
BA	(25% $CO+CO_2$) (75% N_2)	0.78	75	0/100/0	0/1140/0	Cold Melt
BB	(25% $CO+CO_2$) (75% N_2)	0.78	75	0/100/0	0/1120/0	Cold Melt
DA	21% CO_2 79% N_2	0.78	180	0/100/0	0/1344/0	Cold Melt
DB	21% CO_2 79% N_2	0.78	130	0/100/0	0/1137/0	Cold Melt
N	21% CO_2 79% N_2	0.78	75	0/100/0	———	Duplex No. 1
O	21% CO_2 79% N_2	0.78	75	0/100/0	— —	Duplex No. 1
DC	21% CO_2 79% N_2	0.78	130	0/100/0	0/1287/0	Duplex No. 1
DD	21% CO_2 79% N_2	0.78	180	0/100/0	0/1367/0	Duplex No. 1
BO	25% $CO+CO_2$ 75% N_2	0.78	70	0/100/0	0/1177/0	Duplex No. 2

Effect of Steel in Charge on Melting Chemistry

Previous experiments involving a 100 per cent sprue charge have shown no silicon oxidation loss from melting under a 21 per cent CO_2 and 79 per cent N_2 atmosphere (or from a 100 per cent CO_2 atmosphere.[1,2] Present results using the same atmosphere show up to 18.3 per cent silicon oxidation loss when 37.5 per cent steel is present in the charge (see heats CJ to CS, and CV to CX in Table 4. Steel in the charge reacts with CO_2 in the furnace gases to form iron oxide. In turn, the iron oxide causes oxidation of the silicon. Thus, steel can be regarded as contributing to more oxidizing melting conditions. If the amount of steel is increased to 50 per cent of the charge, a greater silicon oxidation loss occurs. This was done in heat CU to produce 25.5 per cent Si oxidation loss under melting conditions equivalent to those producing a maximum of 18.3 per cent loss with 37.5 per cent steel in the charge.

Since oxidizing melting conditions are produced by a combination of steel in the charge and CO_2 in the furnace atmosphere, any additional effects due to free oxygen in the atmosphere are important. Accordingly, heats CY and CZ, containing 37.5 per cent steel, were melted under an air atmosphere. Oxidation losses of 34.2 and 21.3 per cent Si were obtained respectively as compared with the high of 18.3 per cent under CO_2, Table 4. This additional oxidation is attributable to the fact that air has previously been proven to cause silicon oxidation of even a 100 per cent sprue charge with

TABLE 3 — FINAL CHEMICAL COMPOSITION AND CASTING PROPERTIES

Heat	Chemical Composition % C	% Si	% Mn	% S	% P	Mottling** Tendency	Hot Tear	Fluidity, In.	Pouring Temp., F
CJ	2.38	0.98	0.43	0.112	0.075	White	—	—	—
CK	2.28	1.23	0.40	0.135	0.075	White	—	—	—
CL	2.35	1.20	0.39	0.112	0.065	White	—	—	2800
CM	2.41	1.17	0.41	0.112	0.080	White	—	—	2810
CN	2.40	1.43	0.43	0.110	0.066	Very Heavy	—	25	2800
CO	2.42	1.52	0.43	0.100	0.065	Heavy	—	26	2800
CP	2.34	1.49	0.42	0.103	0.066	White	200	27	2820
CR	2.44	1.35	0.43	0.105	0.067	White	275	22	2790
CS	2.44	1.17	0.43	0.105	0.066	White	300	24	2810
CT	2.53	1.39	0.46	0.108	0.068	Slight	285	30	2800
CU	2.34	1.02	0.44	0.115	0.063	White		29	2820
CV	2.38	1.21	0.43	0.112	0.083	White	260	31	2810
CW	2.49	1.44	0.44	0.123	—	White	485	34	2800
CX	2.36	1.26	0.44	0.125	0.071	White	315	30	2830
CY	2.52	0.96	0.43	0.117	0.073	White	295	28	2810
CZ	2.50	1.44	0.43	0.123	—		335	26	2800
BA	2.37	1.10	0.39	0.085	0.130	White	340	24.5	2800
BB	2.31	1.04	0.40	—	—		505	21.25	2780
DA*	2.09	1.11	0.33	0.099	0.121	White	—	—	2760
DB*	2.10	1.12	0.33	0.096	0.121	White	—	22.5	2790
N	2.49	1.29	0.46	0.146	0.124	White	—	26.5	2785
O	2.42	1.27	0.47	—	—		310	28.75	2815
DC*	2.22	1.32	0.36	0.137	0.108	White	—	19.0	2820
DD*	2.24	1.31	0.37	0.136	0.108	White	—	11.0	—
BO	2.38	1.19	0.46	0.150	0.100	White	380	25	2800

*No ferromanganese added.

**2-in. diam x 9 in. long test bar.

no steel present. The increased oxidation accompanying the use of steel is due largely to the larger surface area of the steel as compared with the lesser surface area of the sprue and pig, see Table 1 and 2. The quantity and nature of the oxidation products differ also since the steel forms iron oxide while the sprue and pig tend to form an iron silicate. Heat CT, having no silicon loss, was made under a N_2 atmosphere to demonstrate the principle that the oxidation loss was due to gas-metal oxidation reaction rather than a direct effect of the steel itself.

Effect of Gas Flow Rate

Flow rate of the furnace atmosphere was varied from 0.78 to 5.3 cu ft per minute in different heats. This represents variation by a factor of 6.8, $(\frac{5.3}{0.78})$, from the standard rate of 0.78 cu ft per minute adopted for all previous experiments. No increase of silicon oxidation loss was caused by the highest flow rate employed, as indicated by the data in Table 4. A flow rate increase of 6.8 times should be adequate to detect any sensitivity of the melting operation to this variable. It is concluded that gas flow rate has no effect on silicon (and manganese) losses.

Effect of Melting-Down Time

Heats CV, CW, and CX, Table 4, were made

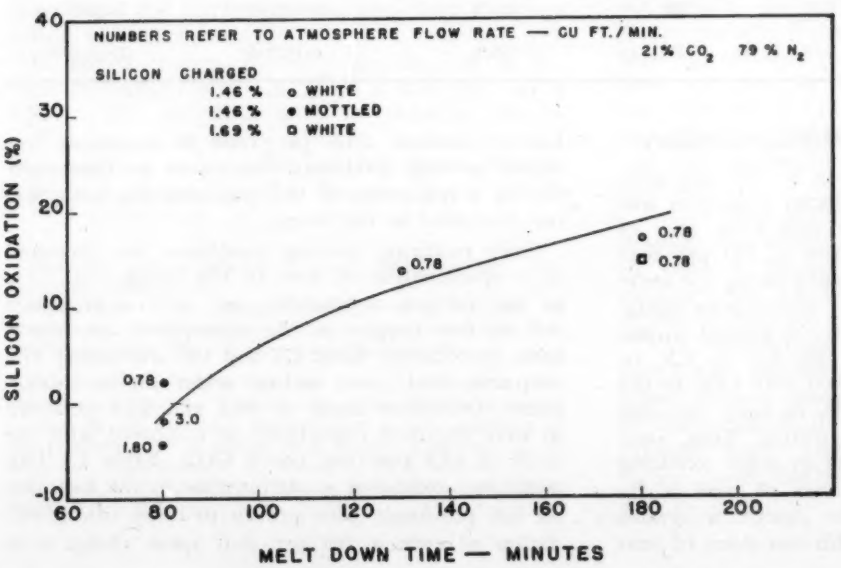


Fig. 1 — Effect of melt-down time on percentage of Si oxidized in a charge of 37.5% steel, 12.5% pig iron, and 50% sprue melted under a 21% CO_2-79% N_2 atmosphere.

Table 4 — Oxidation Losses, Melting Down Rate, and Gas Flow Rates

Heat	Gas Analysis	Flow Rate, cu ft/ min	Time to Melt-down	Silicon Charged	Final Silicon	% Si Oxidation loss	Pts. Si Loss per Pt. Charged	Final Carbon	Melting Rate of Steel, sq in./ min	Pts. Si Loss per sq in. of Steel per min
CJ	21% CO_2 79% N_2	0.78	100	1.20	0.98	18.3	0.183	2.38	23.8	0.925
CK	21% CO_2 79% N_2	0.78	87	1.37	1.23	10.2	0.102	2.28	27.4	0.51
CL	21% CO_2 79% N_2	0.78	85	1.37	1.20	12.4	0.124	2.35	28.2	0.60
CM	21% CO_2 79% N_2	0.78	85	1.37	1.17	14.6	0.146	2.41	28.2	0.71
CN	21% CO_2 79% N_2	0.78	80	1.46	1.43	2.1	0.021	2.40	29.8	0.10
CO	21% CO_2 79% N_2	1.80	80	1.46	1.52	− 4.1	−0.041	2.42	29.8	−0.20
CP	21% CO_2 79% N_2	3.0	80	1.46	1.49	− 2.1	−0.021	2.34	29.8	−0.15
CR	21% CO_2 79% N_2	3.0	80	1.37	1.35	1.5	0.015	2.44	29.8	0.067
CS	21% CO_2 79% N_2	5.3	85	1.37	1.17	14.6	0.146	2.44	27.4	0.73
CT	N_2	5.3	80	1.37	1.39	− 1.5	−0.015	2.53	29.8	−0.067
CU	21% CO_2 79% N_2	3.0	85	1.37	1.02	25.5	0.255	2.34	37.5	0.935
CV	21% CO_2 79% N_2	0.78	180	1.46	1.21	17.1	0.171	2.38	13.23	1.89
CW	21% CO_2 79% N_2	0.78	180	1.69	1.44	14.8	0.148	2.49	13.23	1.89
CX	21% CO_2 79% N_2	0.78	130	1.46	1.26	13.7	0.137	2.36	18.35	1.09
CY	Air	0.78	80	1.46	0.96	34.2	0.342	2.52	29.8	1.34
CZ	Air	0.78	80	1.83	1.44	21.3	0.213	2.50	29.8	1.31
BA	25% $CO+CO_2$ 75% N_2	0.78	75	1.10	1.10	0.0	0.000	2.37	—	—
BB	25% $CO+CO_2$ 75% N_2	0.78	75	1.10	1.04	5.5	0.055	2.31	—	—
DA	21% CO_2 79% N_2	0.78	180	1.10	1.11	− 0.9	−0.009	2.09	—	—
DB	21% CO_2 79% N_2	0.78	130	1.10	1.12	− 1.8	−0.018	2.10	—	—
DC	21% CO_2 79% N_2	0.78	130	1.30	1.32	− 1.5	−0.015	2.22	—	—
DD	21% CO_2 79% N_2	0.78	180	1.30	1.31	− 0.8	−0.008	2.24	—	—
N	21% CO_2 79% N_2	0.78	75	1.30	1.29	0.8	0.008	2.49	—	—
O	21% CO_2 79% N_2	0.78	75	1.30	1.27	2.3	0.023	2.42	—	—
BO	25% $CO+CO_2$ 75% N_2	0.78	70	1.19	1.19	0.0	0.000	2.38	—	—

for comparison with CO, CN, and CP to determine whether melting-down time was a factor influencing silicon oxidation. Time for melting down and heating to 2600 F was controlled at 80 minutes, 130 minutes, and 180 minutes. Thus, this variable was altered by a factor of 2.25 $(\frac{180}{80})$. Increased time for melting down causes increased silicon oxidation as shown in Fig. 1. This variable is far more potent than atmosphere flow rate. The increased time evidently permits a greater amount of oxide skin to be formed on the steel portion of the charge and this results in the higher silicon loss. Therefore, it appears that surface oxidation and oxygen diffusion rate in the solid metal are controlling factors rather than gas flow rate past the surface.

Since melt-down time is so important, a question may be raised relative to oxidation of 100 per cent sprue charges. If the melting down time were prolonged would a silicon oxidation loss occur with 100 percent sprue charge and a 21 per cent CO_2-79 per cent N_2 atmosphere? As mentioned earlier, no loss has been heretofore encountered with normal melt-down time of about 80 minutes. To study this point, heats DA, DB, DC and DD were made with 100 per cent sprue and with prolonged melt-down times of 130 and 180 minutes. Comparison of these heats in Table 4 with BA, BB, N, and O having normal

melt-down time shows that no silicon oxidation was caused by the prolonged melting-down time. Sprue and pig iron then differ fundamentally from steel in the charges in their reaction with CO_2. The gas-metal reaction with sprue, even with prolonged melt-down time, does not produce a high enough iron oxide concentration to cause a silicon oxidation loss. Substantial decarburization, however, does occur and this results in a low-carbon content in the iron, Table 4. By comparison, steel has a severe oxidizing effect on silicon through formation of iron oxide. Accordingly, per cent of steel in the charge, surface area of the steel, and time for melting down (melting rate) all influence the total per cent silicon oxidation loss. On the other hand, only free oxygen in the furnace atmosphere will cause a silicon loss from the sprue portion of the charge.

Effect of Melting Rate

Because of the importance of melt-down time, melting rate (of the surface especially) should have a regulating effect on silicon losses. To study this relationship, the melting rate of the steel surface in square inches per minute was calculated for each heat. Only the steel was considered since this is the factor causing the melting loss in these heats. The points of silicon lost per square inch of steel surface was also calculated. The calculated values are given in Table 4. A plot of the values for all heats containing 37.5 per cent steel and melted under 21 per cent CO_2, 79 per cent N_2 atmosphere is shown in Fig. 2. The graph shows that decreasing the surface melting rate (long melt-down time) causes increasing silicon oxidation loss. The graph also indicates that if a certain minimum melting rate is exceeded there will be virtually no silicon loss during melting down. This amounts to melting down in about 80 minutes or loss for the 100-lb charge, and approximately 2400 sq in. of steel charged in these experiments. For a given charge,

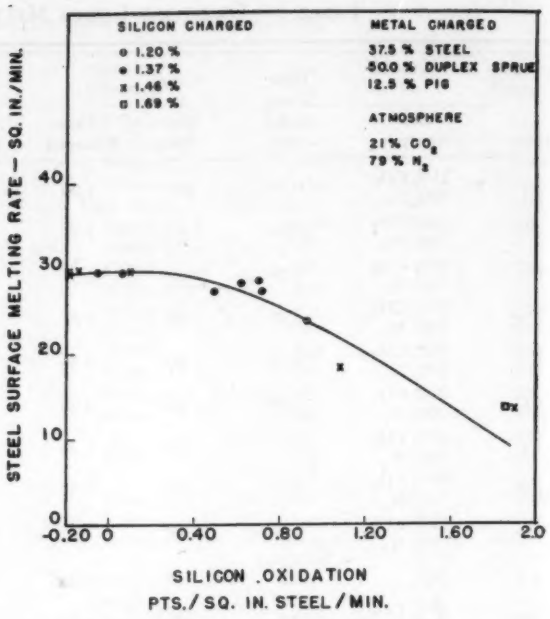


Fig. 2 — Effect of melting rate of the steel surface on the points of Si oxidized per sq in. of steel per minute.

then it appears that there is a critical melting down time. This is illustrated in Fig. 3, which shows that melting this 37.5 per cent steel charge in less than 80 minutes results in no net silicon oxidation loss. The critical melting down time also appears to be relatively insensitive to the percentage of silicon in the charge. The critical melting down time also is determined only by the steel in the charge since no such value exists for 100 per cent sprue charges which show no silicon oxidation from prolonged melting down time.

Heat CU is not plotted in Figs. 2 and 3 since 50 per cent steel and no pig was present in the

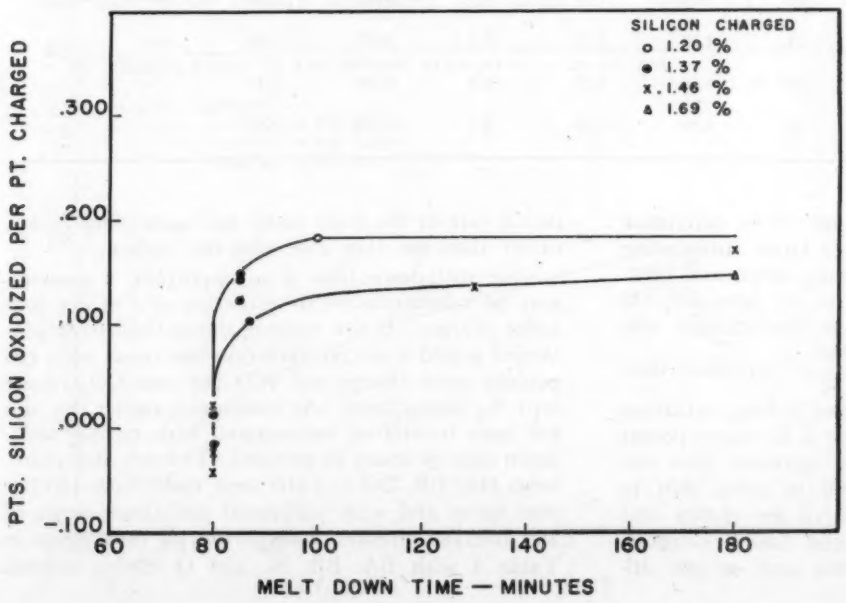


Fig. 3 — Effect of melt-down time on points of Si oxidized per point Si charged in heats containing 37.5% steel, 12.5% pig iron, and 50% sprue. Furnace atmosphere contained 21% CO_2 and 79% N_2.

Table 5 — Mechanical Properties*

Heat	Yield Strength	Tensile Strength	% Elong.	% Si Oxidation	I.Q. Uncorrected	I.Q. Corrected to 2.43% C	I.Q. Corrected to 2.43% C, 0.085% S
CJ	34,500	51,300	20	18.3	91.0	91.0	93.5
CK	36,300	53,800	20	10.2	95.5	95.5	100.5
CL	37,600	53,100	19	12.4	94.5	94.5	97.2
CM	36,800	52,900	19	14.6	93.5	93.5	96.2
CN	37,400	54,500	21	2.1	98.0	98.0	100.5
CO	37,500	54,800	16	−4.1	90.5	90.5	92.0
CP	38,400	55,000	21	−2.1	100.0	100.0	101.8
CR	41,000	56,200	18	1.5	99.5	100.5	102.5
CS	36,100	51,800	20	14.6	92.5	93.5	95.5
CT	37,000	52,200	16	−1.5	88.5	98.5	100.8
CU	31,600	47,400	18.6	25.5	80.5	80.5	83.5
CV	35,000	49,900	18.6	17.1	88.0	88.0	90.7
CW	36,800	52,400	17.9	14.8	91.5	97.5	101.3
CX	35,900	51,400	19.4	13.7	91.5	91.5	95.5
CY	33,100	46,500	18.0	34.2	83.0	92.0	95.2
CZ	36,900	52,500	17.2	21.3	90.5	97.5	101.3
BA	37,700	53,500	23.3	0.0	100.0	100.0	100.0
BB	38,700	53,400	20.7	5.5	98.2	98.2	98.2
DA	41,140	55,150	23.0	−0.9	106.4	106.4	106.4
DB	39,290	55,200	24.4	−1.8	105.5	105.5	105.5
DC	41,125	55,100	16.0	−1.5	96.1	96.1	101.1
DD	41,150	54,240	14.4	−0.8	93.3	93.3	98.3
N	40,000	53,900	16.8	0.8	95.8	101.8	106.8
O	41,500	55,900	18.9	2.3	101.2	101.2	106.2
BO	36,600	51,700	18.0	0.0	90.5	90.5	98.0

*Average of six test bars.

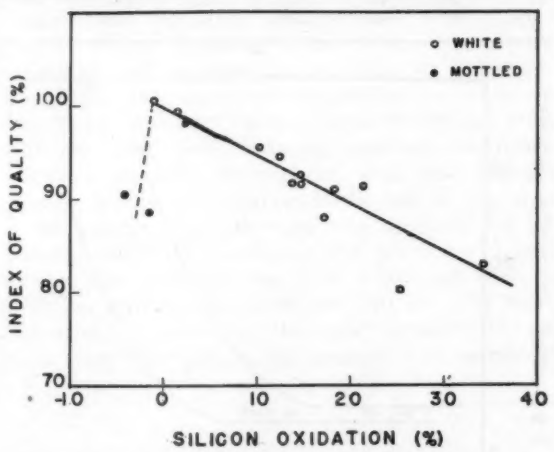


Fig. 4 — Effect of per cent of Si oxidation loss on Index of Quality of malleable iron. The I.Q. value shown is not corrected to 2.43% C but shows the overall effect of oxidizing melting conditions.

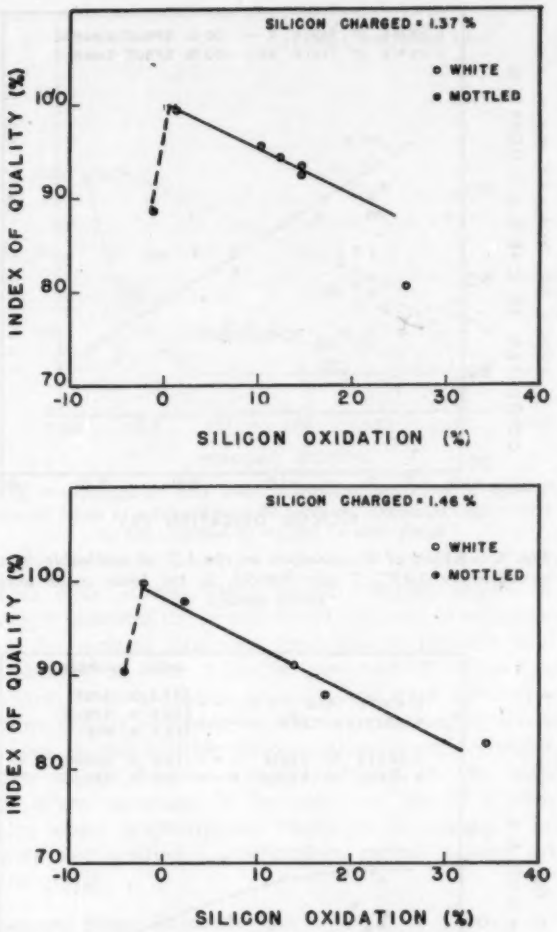


Fig. 5 — Effect of per cent of Si oxidation loss on Index of Quality of malleable iron made from two different series of charges containing 1.3% Si and 1.46% Si.

charge. The slag products and quantity are different and the heat is, therefore, not comparable. Likewise, heats CY and CZ were not plotted because they were melted under air atmosphere.

Tensile Properties

The tensile test bars were annealed according to procedures used previously. Tensile properties of the heats are given in Table 5. No direct effect of

Table 6 — I.Q. and Silicon Losses in Heats of Reference 1

Heat No.	Melting Conditions	Charge 100% Sprue	% Si Charged	% Si Final	% Si Oxidation	I.Q. Uncorrected	Corrected I.Q. 2.43% C	Corr. I.Q. 2.43% C + .085% S
V	Air Atm.	Cold Melt iron	1.10	0.94	14.55	92	92	92
P	Air Atm.	Duplex 1	1.27	1.02	19.65	91	91	98
R	Air Atm.	Duplex 1	1.27	1.03	18.8	87.1	87.1	94.1
BW	Soda Ash — 12-14% CO_2 12-14% CO Bal. N_2	Duplex 2	1.20	0.98	18.3	89.2	89.2	89.2
BV	Soda Ash — 12-14% CO_2 12-14% CO Bal. N_2	Duplex 2	1.20	1.00	16.7	86.4	92.4	92.4

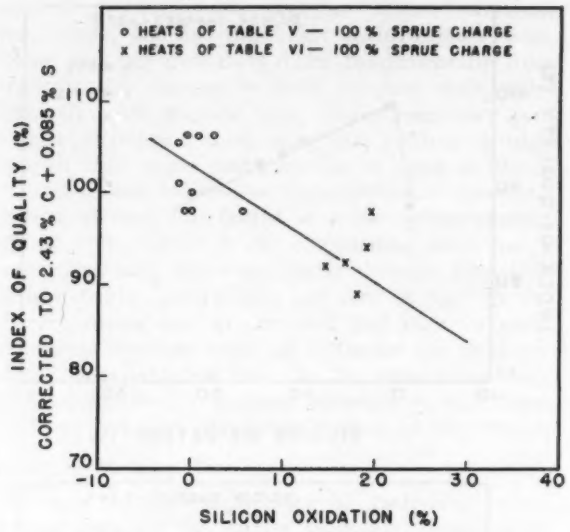


Fig. 6 — Effect of Si oxidation on the I.Q. of malleable iron, corrected to 2.43% C and 0.085% S, for heats made with 100% sprue.

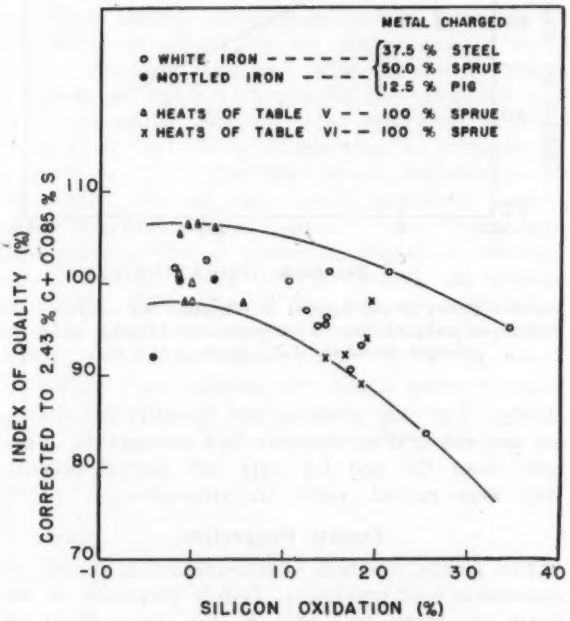


Fig. 7 — Effect of Si oxidation on I.Q. of malleable iron, corrected to 2.43% C and 0.085% S for all heats listed in Tables 5 and 6.

steel in the charge, melting furnace gas flow rate, melt-down time, or melting rate was observed. However, silicon oxidation loss in charges containing steel has an effect on the index of quality (I.Q.)* of the iron as shown in Fig. 4. Steel in the charge thus has an indirect effect on tensile properties through its ability to cause oxidation losses. Figure 4 is a plot of all I.Q. values for these heats con-

*Index of quality is calculated from the combination of tensile strength, yield strength, and per cent elongation for each heat by a method described in Reference 1.

TABLE 7 — ANNEALABILITY

Heat	Time at 1700 F FSG, hr	Cooling Rate of SSG, F, /hr	Nodule Number	% Si Oxidation	% Si	% C
CJ	15	<10	2650	18.3	0.98	2.38
CK	5	20	2250	10.2	1.23	2.28
CL	5	20	2950	12.4	1.20	2.35
CM	5	20	2300	14.6	1.17	2.41
CN	5	20	3200	2.1	1.43	2.40
CO	5	20	2700	— 4.1	1.52	2.42
CP	5	20	2150	— 2.1	1.49	2.34
CR	5	20	2650	1.5	1.35	2.44
CS	10	20	2100	14.6	1.17	2.44
CT	5	20	2100	— 1.5	1.39	2.53
CU	10	<10	1900	25.5	1.02	2.34
CV	10	20	2600	17.1	1.21	2.38
CW	5	30	6075	14.8	1.44	2.49
CX	5	20	3900	13.7	1.26	2.36
CY	30	<10	1200	34.2	0.96	2.52
CZ	10	30	4150	21.3	1.44	2.50
V	30	<10	730	14.55	0.94	2.29
P	20	<10	1800	19.65	1.02	2.41
R	20	<10	1900	18.8	1.08	2.35
BW	15	<10	4350	18.3	0.98	2.43
BV	15	<10	4450	16.7	1.00	2.49
BA	10	10	4450	0.0	1.10	2.37
BB	10	10	2450	5.5	1.04	2.31
DA	10	<10		— 0.9	1.11	2.09
DB	10	<10		— 1.8	1.12	2.10
DC	5	30		— 1.5	1.32	2.22
DD	5	20		— 0.8	1.31	2.24
N	16	10	3900	0.8	1.29	2.49
O	16	10	3400	2.3	1.27	2.47
BO	10	10	4450	0.0	1.19	2.38

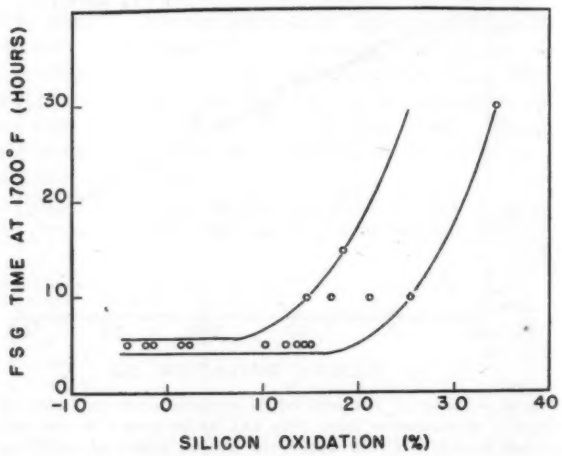


Fig. 8 — Effect of Si oxidation on time required for first stage graphitization (FSG) at 1700 F for heats containing 37.5% steel, 12.5% pig iron, and 50.0% sprue.

taining steel uncorrected as to carbon content of the iron. The graph shows the over-all effect of oxidizing melting conditions. Figure 5 shows the same effect at two different silicon levels in the charge.

In previous work, Ref. 1, silicon oxidation was encountered in 100 per cent sprue charges using air as the atmosphere, or using soda ash during melting down. The I.Q. of the iron and silicon loss of these heats is listed in Table 6. The I.Q. values corrected to 2.43 per cent C and 0.085 per cent S for these heats are compared to the un-

corrected values of the present work in Fig. 6. Heats made with 100 per cent sprue and a prolonged melting down time are also plotted on Fig. 6 with good agreement.

For reasons stated in Reference 1 the I.Q. values should be compared at a common level of 2.43 per cent C and 0.085 per cent S to determine the specific metallurgical effect of silicon oxidation on I.Q. This is done in Fig. 7 for all heats considered thus far. From Fig. 7 it appears that the first severe loss of I.Q. occurs as a silicon oxidation loss of about 10 to 15 per cent is reached. Mottling, a variable factor when oxidation losses are minimum or silicon pick-up occurs, is a factor that produces erratic I.Q. values depending on its severity.

Annealibility

Annealability test bars were treated according to previous practice with results as presented in Table 7.

A plot of FSG time at 1700 F in hours versus per cent silicon oxidation is presented in Fig. 8 for the heats listed in Table 7 that contain 37.5 per cent steel and 12.5 per cent pig iron. One may ask if the increase in FSG time above 15 per cent silicon oxidation is due to lower silicon content in the iron, the effect of oxidation, or some other metallurgical variable. A plot of FSG time versus per cent silicon in the iron for all 100-lb heats throws some light on this question as shown in Fig. 9. Figure 9 includes the values for a wide variety of melting condition reported in References 1 and 2. The minimum graphitization times in Fig. 9 are associated with silicon percentages over 1.10 per cent, with reducing melting conditions, with low humidity atmosphere, with low nitrogen content in the iron and with the use of pig iron in the charge. The pig iron was analyzed for residuals and found to contain 0.06 per cent Ti, and present but probably less than 0.005 per cent B, 0.010 per cent Al, and 0.002 per cent Zr. The residuals could account for the good annealability of irons made with pig in the charge. The spread of

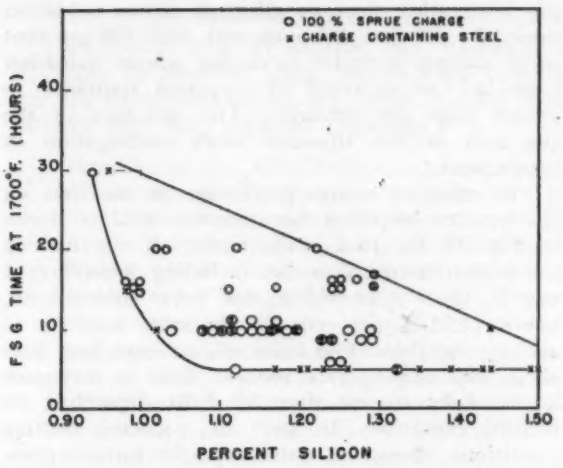


Fig. 9 — Range of first stage graphitization times at 1700 F for all heats as a function of Si content. Encircled digits refer to the number of values at that point.

FSG time at any given silicon content shown in Fig. 9 indicates the sensitivity of the iron to variables in the melting practice. Sensitivity is reduced substantially at over 1.25-1.30 per cent Si. The low silicon irons are more sensitive and have FSG time more readily increased by silicon oxidation, by water vapor in the melting furnace atmosphere, by nitrogen in the furnace atmosphere and in the iron, or other variables. A limitation of Fig. 9 is that first stage graphitization times of less than 5 hr were not studied and therefore cannot appear on the graph.

Second Stage Graphitization

Cooling rate for second stage graphitization does not appear to be affected by per cent silicon oxidation loss in the heats containing steel and pig iron except in the case of heats CU and CY having 25.5 and 34.2 per cent silicon loss respectively. Again this may be due to the residual elements in the

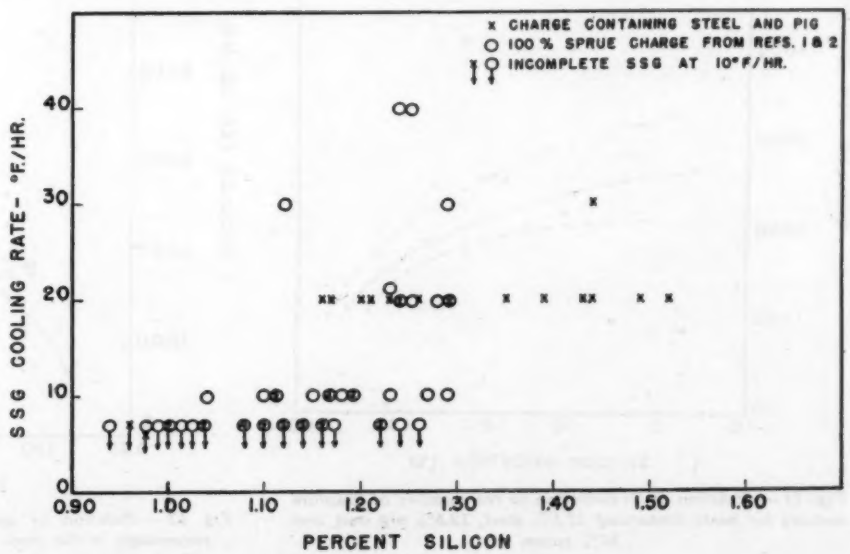


Fig. 10 — Range of second stage cooling rates producing complete graphitization as a function of Si percentage. Encircled digits refer to number of values at that point.

pig iron rather than an effect of silicon oxidation caused by the steel. Previous work with 100 per cent sprue charges indicates increasing silicon oxidation losses are accompanied by increased resistance to second stage graphitization. The influence of the pig iron in this situation bears investigation in future work.

The effect of silicon percentage in the iron on cooling rate required for complete SSG is shown in Fig. 10 for tests made under all the melting conditions studied thus far, including References 1 and 2. Again it is evident that lower silicon irons, below 1.25-1.30 per cent Si, are more sensitive to melting variables. The lower silicon irons may have their SSG cooling rate reduced from a maximum of 30 F/hr to less than 10 F/hr depending on melting conditions. In any case, reducing melting conditions, absence of water vapor in furnace gases, minimum nitrogen in the iron and furnace atmosphere are the melting conditions favoring rapid SSG. Heats containing pig iron have a uniformly rapid SSG rate and point more positively to an important effect of the pig iron.

Nodule Number

Nodule numbers for heats containing 37.5 per cent steel, 12.5 per cent pig iron and 50 per cent sprue are plotted in Fig. 11 as a function of per cent silicon oxidation loss. There seems to be some indication that very high oxidation losses decrease nodule number. However, this may be due to a low final silicon percentage in the iron. Nodule numbers of all 100-lb heats made thus far are plotted in relation to silicon content of the iron in Fig. 12. A nodule count of less than 2000 per cu mm of iron does not occur until the percentage of silicon in the iron drops below 1.20 per cent. Above 1.15 per cent Si, charges containing steel display consistently lower nodule number than do 100 per cent sprue charges. Steel in the charge thus appears to lower the nodule number especially above 1.20 per cent silicon. High nodule numbers are promoted by reducing melting conditions.

Mottling Tendency

Heats CN, CO and CT of this series were found to be mottled. These heats had the low silicon oxidation loss of 2.1 per cent and gains of 4.1 per cent and 1.5 per cent respectively. Total of carbon + silicon was 3.84 per cent, 3.94 per cent, and, 3.92 per cent in the same order. Other heats of the same carbon + silicon level, CW and CZ, were not mottled if they had a high silicon oxidation loss. Oxidation losses for the latter heats were 14.3 per cent and 21.3 per cent respectively. See Fig. 4 for connection of mottling and silicon oxidation losses in this series of heats. Thus, these data agree with past observations on mottling.

Casting Properties

Fluidity and hot tearing data are presented in Table 3. No unusual effects were noted. However, the amount of data is inadequate for consideration at this point.

Discussion

Correlation of specific values of critical melting-down time, points silicon loss per sq in. per minute,

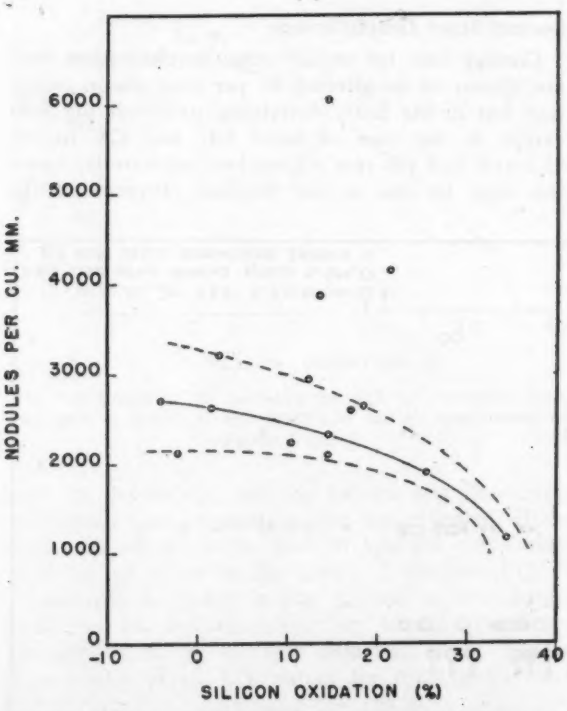


Fig. 11 — Relation of Si oxidation to the number of graphite nodules for heats containing 37.5% steel, 12.5% pig iron, and 50% sprue.

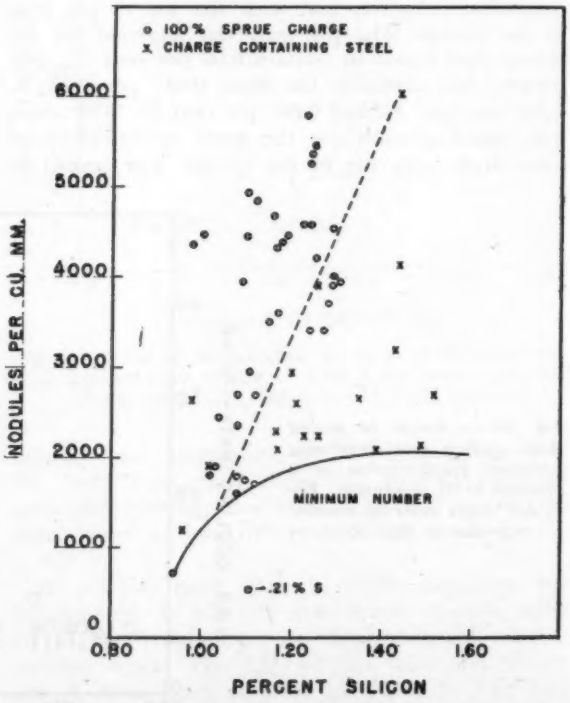


Fig. 12 — Relation of graphite nodule number to the Si percentage in the iron. Heats of Ref. 1 and 2 included.

or I.Q. and oxidation loss obtained in these tests with the same phenomena in melting in other foundries should not be anticipated. However, the general principles described should hold for a wide variety of melting conditions. Due to differences in charge materials, furnace geometry, melting cycle, refractories, etc., the specific values would be expected to vary from one operation to another. However, the existence of the values should hold throughout. For example, in this work a 15 per cent silicon loss or higher was found to lower the tensile properties of the iron. With another combination of melting materials and practice, lowering of properties might not occur until 25 per cent silicon loss was exceeded. Nevertheless, in both cases a critical level exists, which when exceeded produces poor results. The point to be emphasized is that each melting operation must be evaluated on its own merits to discover the specific values of melting down time, charge proportioning, surface area, etc. which are important to that operation.

The data of this report emphasize the importance of the chemical changes occurring during the melting cycle. Metal properties are seen to be related to oxidation which in turn depends on the materials in the charge, proportioning of charge, physical structures of the charge, melting time, and furnace atmosphere. These chemical effects of the melting practice are further complicated by the influence of oxidation on residual elements introduced especially by one component of the charge, pig iron. It is notable that charges containing pig iron (as well as steel) display superior annealability except where excessive oxidation has occurred. If this effect of pig iron is related to its residual elements, then the influence of oxidation during melting on residual elements carrying through the cycle may be a governing factor. Whatever the reason for the effect of the pig iron, it should be studied in future work.

Acknowledgments

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1. Members of the AFS Malleable Iron Research Committee: C. F. Joseph, *Chairman;* W. D. McMillan, *Vice-Chairman;* H. Bornstein; W. A. Kennedy; R. Schneidewind; Milton Tilley; R. P. Schauss; J. H. Lansing; P. F. Ulmer; H. J. Heine, AFS Technical Director; R. J. Osborne, and E. Welander.
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INFLUENCE OF SAND DISTRIBUTION AND SURFACE COATINGS ON METAL PENETRATION

By

AFS Mold Surface Committee (8-H)*

In the 1954 report[1] of the Mold Surface Committee, results of the gray iron penetration tests on dry sand cores indicated that a core wash compounded with silica flour gave superior protection against this defect as compared to a core wash compounded with crystalline graphite. In view of the fact that the 1954 report was limited in scope and included only one core sand and two types of wash, the Committee decided to continue the gray iron penetration study and expand it to include three basic types of non-proprietary core washes and several blends of sand for the core specimens.

Molding and Pouring

The present studies consist of two types of molds designed to show the resistance to metal penetration into cores. In one case Standard Gertsman's Molding Method was used. For the metal penetration tests the test mold follows closely the one used in earlier investigations, insofar as using the ram-up core (Fig. 2) in the bottom of each mold.

In the earlier investigations, the mold was constructed as shown in Fig. 1, with a 20-in. high metal head with a 2-in. diameter riser cast over four test specimens located around the center pouring gate. Specimens were equally spaced on a 3-in. diameter circle and spaced 90 degrees apart. In the so called Revised Gertsman Test, the core stack was extended to determine the influence of the static head on metal penetration. Thus, by extending the 6-in. diameter cavity over the test specimens, the cores could be stacked the full height of the mold as shown in Fig. 3.

In the penetration test molds the four locations around the ingate enabled the placing of equally-spaced stacks of cores using the $1\frac{1}{8}$-in. diameter by 2 in. long. In the total stack height of 20 in., ten cores were pasted end to end to form the stack. The several molds made contained cores of each sand series, with each mold containing one series but coated.

I. No coating
II. Silica-base core wash
III. Crystalline graphite-base wash
IV. Mexican graphite-base wash

For details of these coatings, reference is made to the section of the report dealing with them and their application. One may see the construction of the mold enables determination of several factors, namely:

A. Effect of sand blends and distribution on metal penetration
B. Effect of sand fineness on metal penetration
C. Ferro-static pressure influence on metal penetration
D. Metal penetration with and without sand protection
   1. No coating
   2. Silica flour-base coating
   3. Crystalline graphite base coating
   4. Mexican graphite base coating

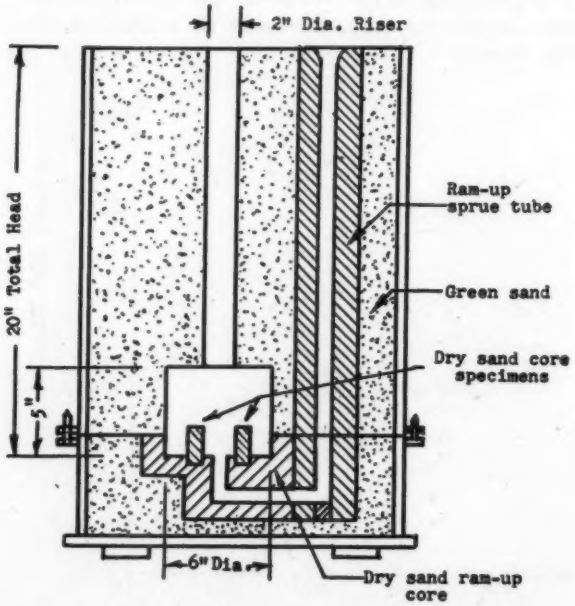


Fig. 1 — Standard Gertsman test mold.

*Mold Surface Committee (8-H) consisted of the following personnel: C. C. Sigerfoos, *Chairman;* C. E. McQuiston, *Secretary;* J. B. Caine, A. Dorfmueller, Jr., S. L. Gertsman, J. E. Haller, C. W. Hockman, H. A. LaForet, J. A. Ridderhof, R. R. Schaaf, C. J. Schwetz, G. J. Vingas, D. C. Williams, and E. E. Woodliff.

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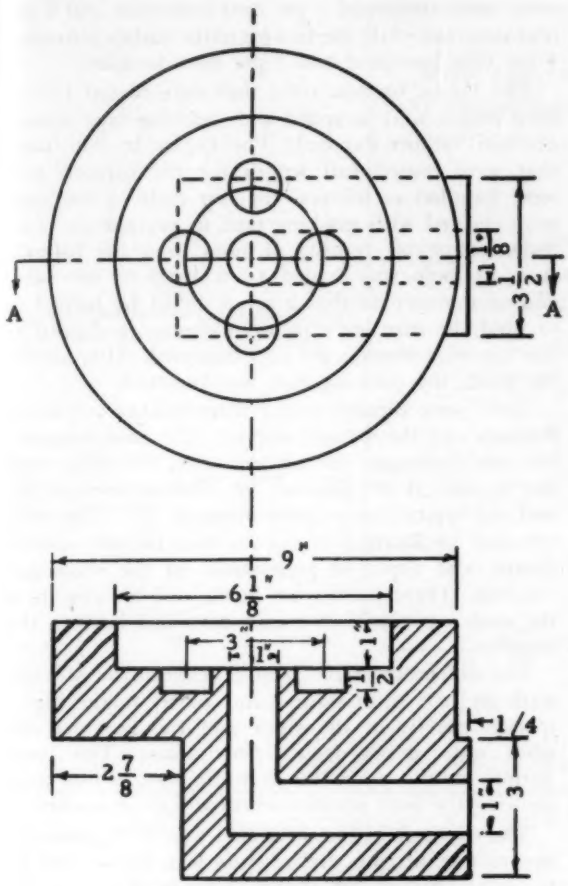


Fig. 2 — Dry sand ram-up core.

The test cores were made according to the sand blends arrived at to give the proper distribution of normal, skewed right and skewed left from normal, as explained elsewhere in this report. The following standard mixture was employed:

95.50% Sand, as blended in Table 4
1.25% Cereal (gelatinized type)
1.75% Core oil (linseed oil original)
1.50% Iron oxide

Moisture was varied to provide a constant "temper" or "feel" to the sands of increasing fineness.

Mulling of the sand was in a laboratory size muller with the following cycle: 2 minutes mixing of dry ingredients and 5 minutes with core oil and water.

Test cores were made of 1⅛-in. diam by 2 in. long size and the 2-in. diam by 2-in. long size in sufficient quantity for all tests. Each size of specimen was rammed the standard three rammer blows of the standard specimen rammer. All cores were baked in standard AFS test core baking ovens 1½ hr at 400 F.

In addition to the making of the test cores for the castings, green and baked physical properties were determined on the sand mixtures and these data are reported under data section, Table 1.

The test cores for the Revised Gertsman Test were coated with the respective washes and wrapped

Table 1 — Mechanical and Physical Properties of Blended Core Sand Mixtures

	x=300 Normal	x=300 Skew Left	x=300 Skew Right	x=200 Normal	x=200 Skew Left	x=200 Skew Right
Moisture, %	3.6	2.3	3.5	4.9	4.3	5.8
Green Permeability	85.0	182.0	99.0	52.0	61.0	37.0
Baked Permeability	118.0	225.0	127.0	69.0	86.0	51.0
Green Compressive Strength	1.3	0.5	1.0	1.6	1.2	1.6
Wt. 2-in. x 2-in. Specimen	170.0	168.0	165.0	173.0	176.0	172.0
Tensile Strength, psi	283.0	260.0	152.0	266.0	368.0	279.0

in paper. In the foundry the test cores were first drilled with a ⅜-in. drill prior to assembly in stacks. The drilled hole extending lengthwise through the specimens of the 1⅛-in. diam size. The cores were assembled on a ⅜-in. diam steel pipe or arbor which had been drilled with small holes about every 2 in. of its length for vents. Cores were assembled on this arbor after the ends were coated with paste and by pressing them together over the arbor formed a firm tight joint and a core stack. These were dried at a low heat to dry the paste and remove any absorbed moisture prior to using. The core arbor pipes were so made as to extend out of the mold top of each core stack to enable gripping them for alignment of the four stacks in each mold and to weight the stacks down to prevent floating. (See Fig. 3).

The molding sand used for making the molds was the stock molding sand of the Foundry Laboratory of the Michigan State University. It was a mulled sand riddled and contained clay bond and ample sea coal to stabilize it. No failures were experienced

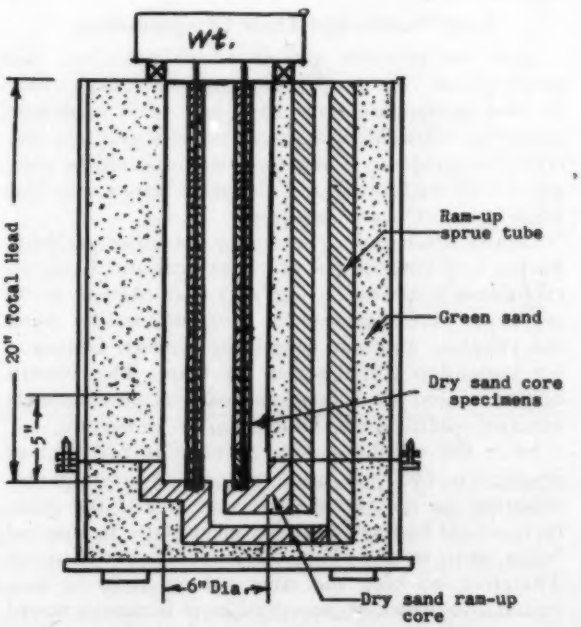


Fig. 3 — Revised Gertsman test mold.

in any form by the molding materials. The molds were successfully made and assembled by members of the Committee. All molds were of green sands with ram-up cores.

Test molds were made and poured immediately with gray iron melted in an electric arc furnace. Pouring temperature was held to the range of 2600 to 2625 F as observed with an optical pyrometer. This practice allowed close control of the pouring temperature and analysis. The charge analysis of the metal was as follows:

Carbon, %	— 3.10
Silicon, %	— 1.75
Manganese, %	— 0.78
Phosphorus, %	— 0.12
Sulphur, %	— 0.06

An alloy of 75% ferrosilicon was used to inoculate the melt during tapping.

Pouring was done by a large ladle which was preheated and kept under heat between casts. The rate of pouring was medium fast as the 1-in. sprue and gate took metal rapidly. The stack core or penetration molds were poured full through the gate with no "touch-up" allowed for this would have subjected the top cores to uneven heating. The shrinkage of the high casting resulted in the loss of the top two cores for comparison purposes, for the drop in casting length was in the range of ½ to ¾ in.; thus, they were lost so far as the test was concerned as metal did not surround them completely.

All castings were allowed to cool down to black heat in the molds before shakeout. They were then sprued and the penetration molds were sandblasted outside and sectioned through the test core areas; the sections were removed by milling along the length of the castings. The studies of these core surfaces with photographs form the basis for the remarks and conclusion.

Core Washes and Their Compounding

Based on previous tests in which silica-base and graphite-base core washes were used, the Committee decided to compare three basic core wash materials; crystalline silica, Mexican or amorphous graphite and crystalline graphite. These three washes could be compared with each other and also with results obtained when no core wash was used.

Ottawa silica flour, (140 mesh), Mexican graphite, having a carbon content of 80 per cent and Bavarian crystalline graphite, having a carbon content of 90 per cent, were compounded into core washes using the simplest formulas containing western bentonite for suspension and dextrine for bond. The western bentonite used was a premium material purchased on viscosity specifications (minimum 30 centipoise).

Since the specific gravity of silica is 2.55 and of graphite is 1.93, and since the particle size and distribution for the two materials are nearly the same, there would be more particles of graphite to suspend for a unit weight than there would be of silica. Therefore, to keep the same volume ratio of base material to bentonite, one third more bentonite would be required for the graphite than for the silica. The silica wash contained 3 per cent bentonite and 2 per cent dextrine while the two graphite washes contained 4 per cent bentonite and 2 per cent dextrine.

The 1⅛-in. by 2-in. cores that were coated 1-in. of their length were wrapped with masking tape leaving the 1-in. of core exposed. The 1⅛-in. by 2-in. cores that were coated full length for the column tests were handled as follows: the top ends of the cores were covered with masking tape to prevent the wash from sealing off venting of gases from the column. A ⅛-in. hole was drilled ½-in. deep in the other end of the core so that a screw could be turned in to hold the core for dipping. Cores were dipped to the top edge leaving the end uncoated. After drying the wash, the masking tape was removed.

Cores were dipped in the three washes at various Baumes and the Baume selected that gave comparable core coverage. For all tests run, the silica wash was applied at 53° Baume, the Mexican wash at 43° and the crystalline graphite wash at 32°. The wide variance in Baumes is due to two factors: specific gravity and depth of penetration of the wash into the core. These factors are borne out by weights of the washes per unit core area described later in this report.

The difference of 10° Baume between the Mexican wash and the silica wash is due partly to the higher specific gravity of the silica and the fact that the silica wash shows deeper penetration. The lower Baume for the crystalline graphite wash is due to its showing the least penetration of the three washes.

The coverage of the three core washes in grams per square foot of core surface area was determined for the sands. Silica wash showed a range of 15.3 to 20.3 grams per square foot, the Mexican wash 9.4 to 17.9 grams and the crystalline graphite, a range of 5.3 to 10.7 grams. The coarser the sand, the deeper the wash penetration and the greater the weight of wash per unit area of core surface.

The initial work employing the Revised Gertsman Test was conducted on cores of 1⅛-in. diam by 2-in. made using a series of washed and dried silica-base sand of the following fineness ranges: 39.1, 47.9, 60.8 and 76.3. The core sand mixture was bonded as in following tests, i.e., with 1.75 per cent core oil and 1.25 per cent cereal binder. Baked cores were coated with the same basic washes used later. Casting results of the stacked cores showed metal penetration results comparable to those of the blended sand mixtures with designed grain fineness and distributions.

Sand Distribution

The previous work on metal penetration performed by this Committee had been done using various commercially available sands. It was felt that the distribution characteristics of sands exerted an influence on the behavior of metal penetration and the coating of cores with various types of washes. Hoar and associates[2,3] have reported in their works that distribution of sands is important and their work influenced the Committee in deciding to approach this problem. An interpretation of Hoar and Atterton's work leads to the possible conclusion that certain sta-

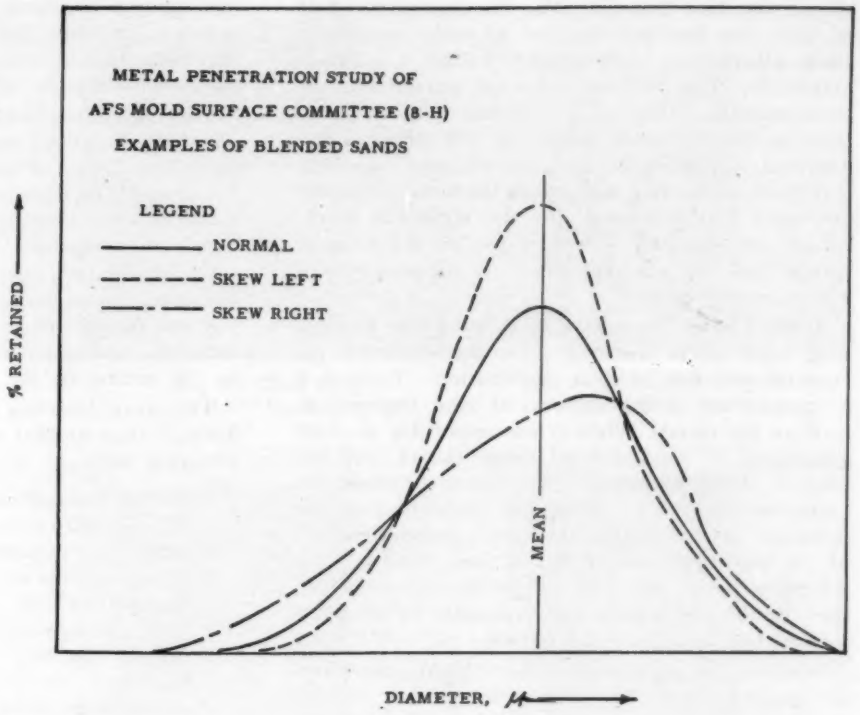


Fig. 4 — Curves showing statistical parameters.

Table 2 – Screen Analysis of Blended Sands

Screen Size	x̄=200 Normal			x̄=200 Skew Left			x̄=200 Skew Right		
	Design	Calculated	Actual	Design	Calculated	Actual	Design	Calculated	Actual
20		0.1	0.1						0.1
30		0.1	0.2					0.2	0.1
40		2.4	2.0		0.4	0.5	1.9	1.5	1.2
50	12.2	12.6	11.7	10.8	11.5	11.2	13.5	8.9	8.6
70	35.3	33.9	33.9	43.4	41.1	40.8	23.4	23.3	24.2
100	25.8	26.6	26.9	27.1	26.5	31.0	24.5	23.8	21.8
140	12.8	10.9	11.4	10.8	10.7	10.9	17.5	18.9	20.0
200	6.0	7.6	7.5	4.5	4.9	4.1	8.9	14.3	14.2
270	3.3	2.7	3.4	2.0	1.1	1.9	4.6	4.7	6.2
Pan	4.6	1.9	2.9	1.2	1.0	0.6	5.7	2.8	3.6
AFS Clay	—	1.2		—	0.6	—	—	1.4	
Total	100.0	100.0	100.0	99.8	97.8	100.0	100.0	99.8	100.0
	x̄=300 Normal			x̄=300 Skew Left			x̄=300 Skew Right		
	Design	Calculated	Actual	Design	Calculated	Actual	Design	Calculated	Actual
20			0.1		0.2	0.3		0.6	0.5
30	0.5	2.1	1.6		3.8	3.3	3.2	5.0	3.9
40	18.3	18.1	13.4	20.5	21.0	21.4	16.2	13.6	12.4
50	34.2	34.0	29.5	44.1	43.4	45.5	24.0	23.4	23.9
70	21.7	22.4	25.3	19.0	23.2	21.8	24.3	24.9	26.7
100	11.9	11.7	16.3	9.7	6.3	6.7	15.5	17.0	18.8
140	5.6	6.7	8.0	4.3	1.1	.9	7.7	6.7	6.6
200	2.8	2.4	3.1	1.9	.2	.1	4.2	4.1	3.6
270	1.7	.6	.9	.5	—	—	2.3	1.4	1.7
Pan	3.3	.9	1.8	—	—	—	2.1	1.3	1.9
AFS Clay	—	.9	—	—	—	—	—	1.4	—
Total	100.0	99.8	100.0	100.0	99.2	100.0	99.5	99.4	100.0

tistically designed sands will show a decided improvement in the influence to prevent metal penetration. The work indicates that one of the more efficient methods of retarding metal penetration is to have the pore diameter of the surface voids as small as possible.

Commercially available sands do not lend themselves, in general, to definitive statistical parameters. Thus, their use for research type work may be limited. In any event, the Committee decided to design sand distributions to certain statistical parameters. It is realized, of course, that the distribution of all foundry sands is a statistic.

The statistical parameters chosen were those of mean or "average" and a measure of the distribution

spread, standard deviation. Also the Committee chose to skew the distributions and to make some flat, while others were more peaked. Figure 4 indicates graphically these various statistical parameters just mentioned. Six different distributions were used, i.e., three sands each with means of 200 microns (approximately 70 mesh) and 300 microns (approximately 50 mesh). For each mean diameter (3 sands) one sand had a normal (in the statistical sense) spread; one was skewed to the right and flat or large spread; and one was skewed to the left and peaked with a narrow spread.

Table 2 gives the screen analysis of the blended sand used, while Table 3 gives the statistical parameters obtained of these distributions. Table 4 is a compilation of the commercial sand required to produce the blends. While it was impossible to cover adequately all conditions of mean, spread, and unbalance of the distribution with these six sands, the Committee hoped to gain some indication of the influence of distribution on metal penetration. All of the statistical parameters of these designed distributions were based on the arithmetic scale, i.e., they would not follow the geometric or standard screens, but rather linear intervals.

Statictical techniques have been highly developed for analytical use. It was not the purpose of this investigation to pursue any of these techniques, but rather to indicate that they are applicable in such problems, and to show that the statistical methods are rather useful in solving problems of this nature.

The Committee would like to point out that these statistically designed sand distributions were blended from several types of basic sands; namely, lakes, banks, silicas, etc. The distributions are to be considered as research and development sands, and their use is purely experimental.

Earlier in this paper it was noted that iron oxide was added to the basic core mixtures. This iron oxide was not considered in these distributions. The distribution data given are that of the sand only prior to the adding of the binders.

The actual blending operation was somewhat more difficult than at first anticipated. Steps necessary to blending were:

1. Obtain commercial sands on the basis of analysis supplied by the producer.
2. Screen each commercial sand.
3. Design blends on the basis of screen analysis of commercial sand.
4. Screen blends to check design.
5. Repeat the steps to modify the blends where necessary.

Problems were encountered in this blending operation because sand producers' sands varied considerably; sampling techniques caused some fluctuations in screen analysis; and the screening is not 100 per cent efficient. The actual design of the blends is difficult because in using from the 30-mesh screen to AFS clay, ten screens or their equivalent are encountered. Thus, mathematically, in order to design the blends, it would be necessary to have ten commercial sands. This would yield a set of tenth order simultaneous equations which is impractical to solve without a high speed computer. Consequently, it was necessary to design with a set of fourth order equations which introduced some slight errors.

TABLE 3 — STATISTICAL PARAMETERS OF BLENDED SAND

	Design Mean	Actual Mean	Design Standard Deviation	Actual Standard Deviation
		$\bar{X} = 300$		
Normal	300	290	100	101
Skew Left	300	333	75	105
Skew Right	300	308	125	138
		$\bar{X} = 200$		
Normal	200	208	66.7	82.5
Skew Left	200	205	50	62.5
Skew Right	200	189	83.5	86

TABLE 4 — SANDS USED TO MAKE BLENDS

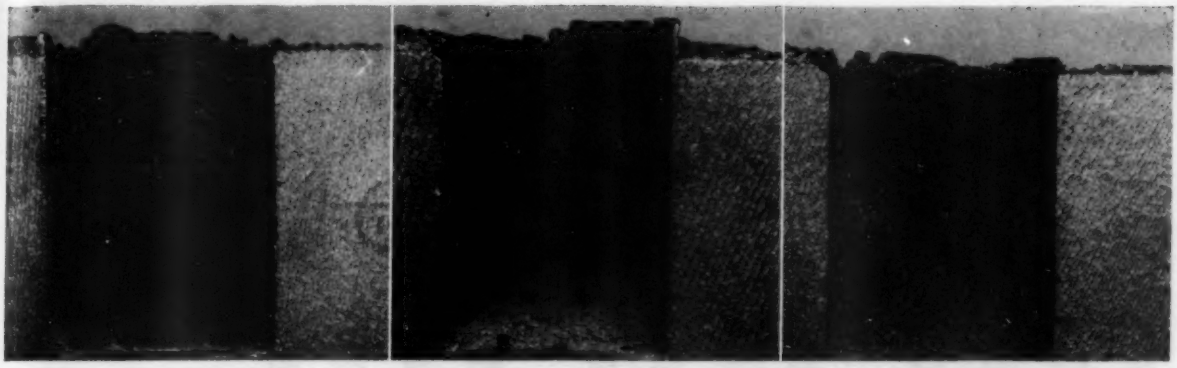
$\bar{X}$=300 normal	19.4%	"Blended Coarse" Sand Co. A
	9.7%	Sand Co. B
GFN — 60.0	67.9%	Silica Sand Co. C
	3.0%	90 Mesh N. J. Silica Flour
$\bar{X}$=300 skew left	14.0%	Sand Co. B
	6.0%	#3 Dry Silica Sand Co. D
GFN — 41.9	80.0%	"Blended Coarse" Sand Co. A
$\bar{X}$=300 skew right	48.5%	#4 Tenn. Silica Sand
	38.8%	#3 Dry Silica Sand Co. D
GFN — 61.4	9.7%	85 Mesh Dry Silica Sand Co. D
	3.0%	90 Mesh N. J. Silica Flour
$\bar{X}$=200 normal	24.5%	85 Mesh Dry Silica Sand Co. D
	24.5%	#63 Silica Sand Co. E
GFN — 75.5	49.0%	#60 McConnelsville
	2.0%	90 Mesh N. J. Silica Flour
$\bar{X}$=200 skew left	25.9%	Bond Sand Ottawa
	46.8%	#50 Sand Ottawa
GFN — 67.0	25.3%	"Blended Fine" Sand Co. A
	2.0%	90 Mesh Silica Flour, N. J.
$\bar{X}$=200 skew right	29.4%	85 Mesh Dry, Silica Sand Co. D
	29.4%	#103 McConnelsville
GFN — 93.3	39.2%	#60 McConnelsville
	2.0%	90 Mesh Silica Flour, N. J.

Discussion of Results

The results indicate, as may be seen from Figs. 5 and 6 and Tables 5 and 6, that the Revised Gertsman Test is less severe than the Standard Gertsman Test. Partial explanation for this seems to be that the end (90° corners) effects are not present. However, this does not seem to explain adequately the entire difference. The fact that these stacked cores had to be supported on an arbor (vent) may have had an influence on the severity. On the top end of the Revised Gertsman Test, some apparent roughness is in evidence. It is felt that this condition is not indicative of the test in general, but that this condition occurs for the following reasons:

1. Considerable shrinkage occurs in this area.
2. Metal is colder because of bottom gating.
3. Obviously, ferrostatic pressure is lower.

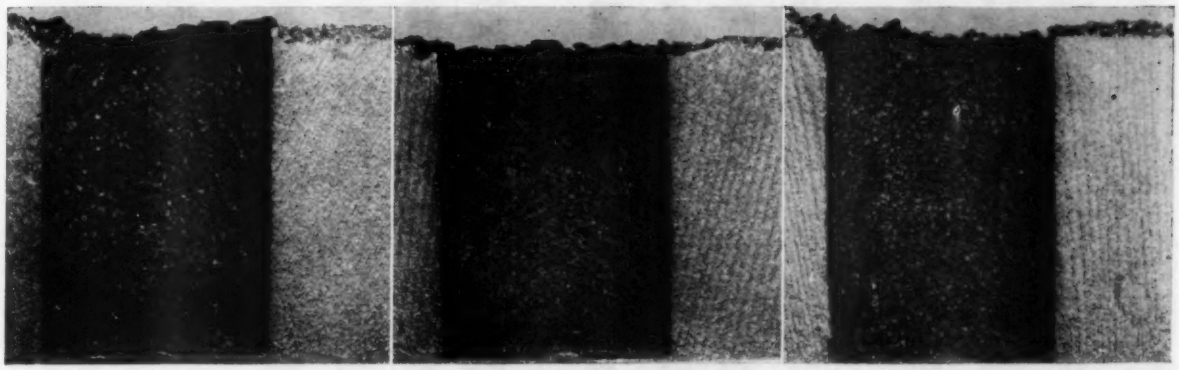
Figures 5 and 6 are indicative of the type of results that are obtained in the Standard Gertsman Test. It is extremely difficult to evaluate the results of this test without some yardstick or quantitative measure



No. 1 — Silica Flour
X 300 Normal

No. 2 — Silica Flour
X 200 Normal

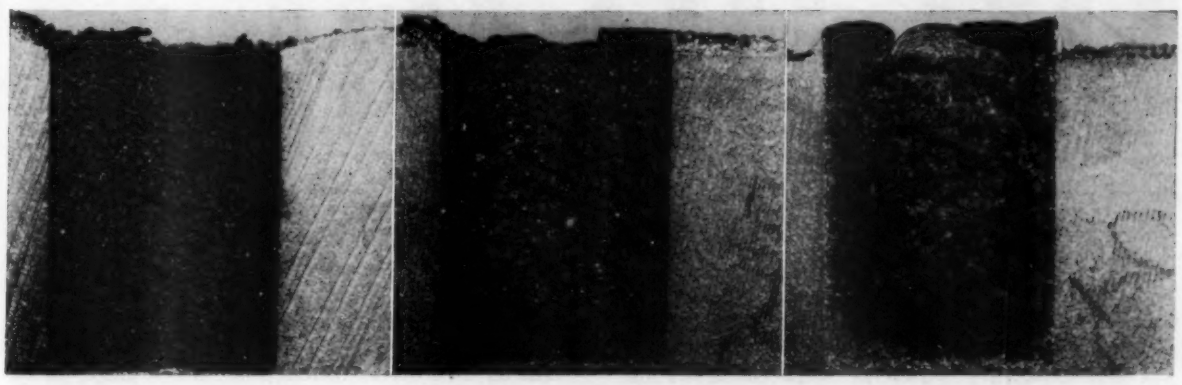
No. 3 — Crystalline Graphite
X 200 Skew Right



No. 4 — Mexican Graphite
X 300 Normal

No. 5 — Mexican Graphite
X 200 Normal

No. 6 — Crystalline Graphite
X 200 Skew Left



No. 7 — Crystalline Graphite
X 300 Skew Right

No. 8 — Crystalline Graphite
X 300 Normal

No. 9 — Uncoated
X 300 Skew Left

Fig. 5 — Typical results of standard Gertsman Tests.

of the penetration. Thus, the Committee has attempted to set up such a yardstick. It recognizes as empirical any test or measure of the severity of penetration. These figures are the means by which the Committee chose to evaluate the results of the penetration studies in this work.

In some cases it was difficult to find an exacting demarcation between grades of penetration. Thus, in the case of ratings 3, 4, and 5 it is recognized that the measure may at first seem inconsistent. However, it will be noted that Class 4 is considerably rougher on the top than Class 5, and smoother on the bottom, and that it is smoother on the bottom than on some parts of Class 3. Consequently, it was decided to make this an intermediate grade in view of the fact that it was not entirely consistent from top to bottom.

A summary of the results will be found in Tables 5 and 6. These tables contain the same data classified in different forms to facilitate the analysis of test results. Primarily on the basis of these tables and Fig.

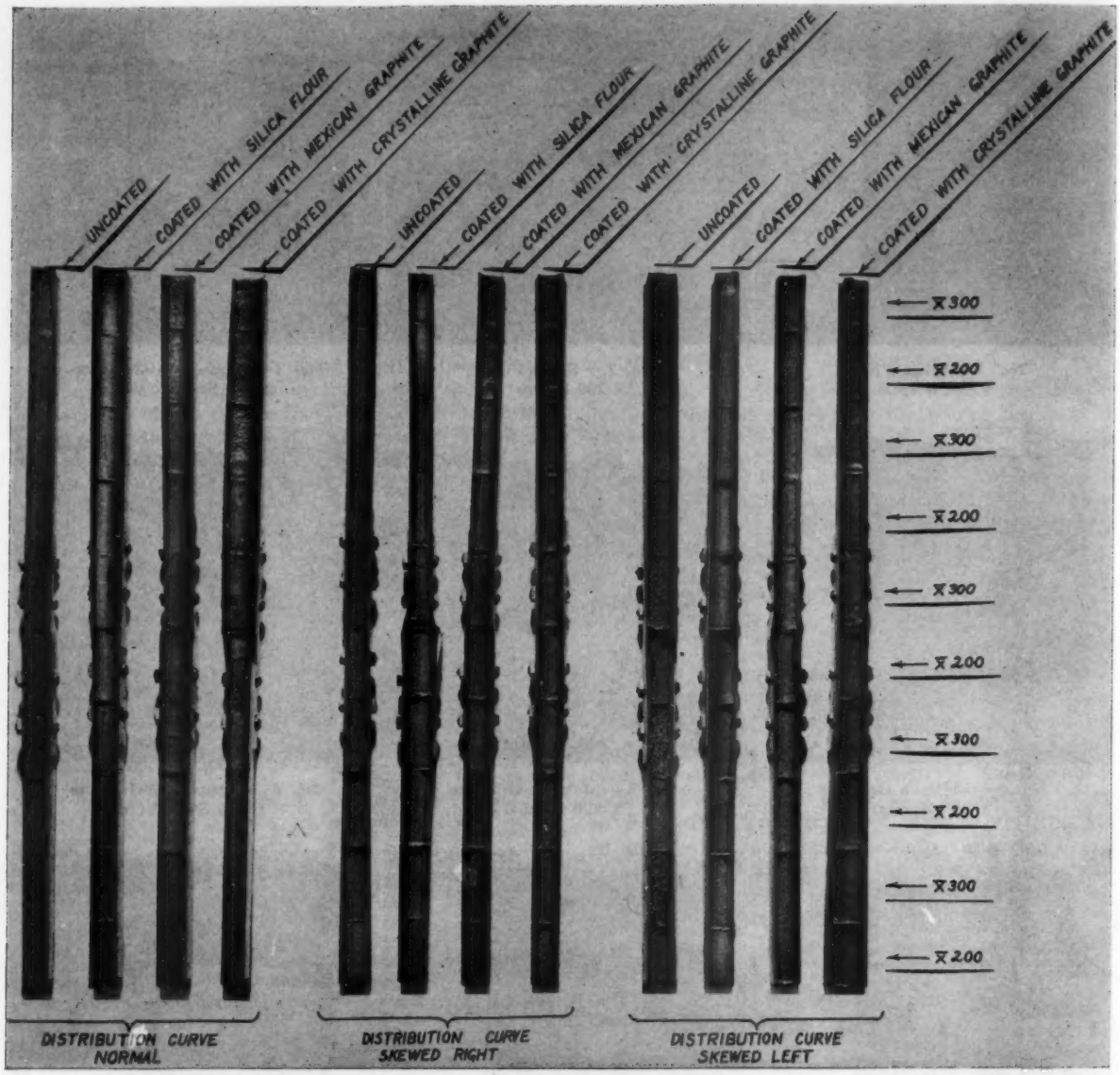


Fig. 6 — Sections cut from 20-in. high specimens; Revised Gertsman Test.

6 the Committee has drawn its conclusions which are presented later in the paper.

The comparative difference between crystalline graphite and Mexican graphite requires explanation, insofar as with the normal and skew right sands, the sand voids are smaller making for absorption of finer Mexican graphite as against the larger voids of the skew left sand which is more conducive to absorption of the crystalline graphite. Generally, the coarser sands show less penetration with the flat crystalline graphite wash as compared to the showing of less penetration of fine sands when coated with Mexican graphite. In the comparison of the skew right (finest) sands there was no noticeable difference in any of the three washes.

Some question may arise as to the correlation between the AFS Grain Fineness Number and the statistical parameter used to design our blends. The AFS Grain Fineness Number is a weighed average and is based on a geometrical scale. The statistical parameter used herein included not only a mean (arithmetical) but also a measure of the spread. There is no conflict between these two methods of describing the distribution. In fact the same methodology is used in calculating both. The only difference is that the statistical parameter used herein defines not only the mean but also measures of the distribution such as spread, peakness, and skew. Those which have more spread or are flatter will have a higher grain fineness number than those of the same mean which are more peaked.

The mechanical properties of the core mixtures used in this test are given in Table 1 as reported earlier. The water in these mixtures was varied for workability considerations on the basis of the grain fineness numbers.

Table 5 — Penetration Test Results
Based on Visual Inspection and Classified 0 to 9
(0 = no penetration; 9 = deep penetration)

Standard Gertsman Test			Revised Gertsman Test*		
Degree of Penetration	Type of Wash or Coating	Sand Blend	Degree of Penetration	Type of Wash or Coating	Sand Blend
1	Silica Flour	x 300 Normal	0	Silica Flour	x 200 Skewed Left
1	Silica Flour	x 300 Skewed Right	0	Silica Flour	x 200 Skewed Right
1	Silica Flour	x 200 Skewed Left	0	Silica Flour	x 200 Normal
2	Silica Flour	x 200 Normal	0	Silica Flour	x 300 Normal
2	Silica Flour	x 200 Skewed Right	0	Crystalline Graphite	x 200 Skewed Left
3	Uncoated	x 200 Skewed Right	0	Crystalline Graphite	x 200 Skewed Right
3	Crystalline Graphite	x 200 Skewed Right	1	Silica Flour	x 300 Skewed Right
3	Mexican Graphite	x 200 Skewed Left	1	Mexican Graphite	x 200 Skewed Right
3	Mexican Graphite	x 200 Skewed Right	1	Mexican Graphite	x 300 Skewed Right
4	Mexican Graphite	x 300 Normal	1	Crystalline Graphite	x 300 Skewed Left
4	Mexican Graphite	x 300 Skewed Right	1	Crystalline Graphite	x 300 Skewed Right
5	Mexican Graphite	x 200 Normal	2	Silica Flour	x 300 Skewed Left
5	Uncoated	x 200 Normal	2	Mexican Graphite	x 200 Normal
5	Crystalline Graphite	x 200 Normal	2	Mexican Graphite	x 300 Normal
6	Crystalline Graphite	x 200 Skewed Left	2	Mexican Graphite	x 200 Skewed Left
6	Silica Flour	x 300 Skewed Left	2	Uncoated	x 200 Skewed Right
6	Uncoated	x 300 Normal	3	Uncoated	x 200 Skewed Left
6	Uncoated	x 200 Skewed Left	3	Uncoated	x 300 Skewed Right
6	Uncoated	x 300 Skewed Right	3	Uncoated	x 200 Normal
7	Crystalline Graphite	x 300 Skewed Right	3	Crystalline Graphite	x 200 Normal
8	Crystalline Graphite	x 300 Normal	3	Mexican Graphite	x 300 Skewed Left
9	Uncoated	x 300 Skewed Left	5	Crystalline Graphite	x 300 Normal
9	Mexican Graphite	x 300 Skewed Left	7	Uncoated	x 300 Normal
9	Crystalline Graphite	x 300 Skewed Left	8	Uncoated	x 300 Skewed Left

*Evaluation was from cores located at bottom of core stack, the location of cores compared in Standard Gertsman's Test.

Table 6 — Penetration Test Results. Supplementary Classification

Based on Visual Inspection and Classified 0 to 9
(0 = no penetration; 9 = deep penetration)

	Gertsman Test	Revised Gertsman Test
x 200: (Silica Flour)		
Skew Right	2	0
Normal	2	0
Skew Left	1	0
x 300: (Silica Flour)		
Skew Right	1	1
Normal	1	0
Skew Left	6	2
x 200: (Mexican Graphite)		
Skew Right	3	1
Normal	5	2
Skew Left	3	2
x 300: (Mexican Graphite)		
Skew Right	4	1
Normal	4	2
Skew Left	9	3
x 200: (Crystalline Graphite)		
Skew Right	3	0
Normal	5	3
Skew Left	6	0
x 300: (Crystalline Graphite)		
Skew Right	7	1
Normal	8	5
Skew Left	9	1
x 200: (Uncoated)		
Skew Right	3	2
Normal	5	3
Skew Left	6	3
x 300: (Uncoated)		
Skew Right	6	3
Normal	6	7
Skew Left	9	8

Conclusions

On the basis of the work performed by the Committee, the following conclusions have been made:

1. The results of this work substantiate previous findings of Gertsman and this Committee; namely, that increasing ferrostatic head increases metal penetration all other things being equal.

2. The three washes gave protection against metal penetration. In the case of the Standard Gertsman Test, the silica flour was best, Mexican graphite good, and crystalline graphite was poor. The silica flour wash also gave the best protection in the Revised Gertsman Test, while neither Mexican or crystalline graphite were consistently better than the other.

3. In both tests, the silica flour wash gave added protection against metal penetration regardless of the base core sand.

4. The sand distribution which gave the best protection against metal penetration, in both tests, was one which was flat and skewed to the right. This sand also has a greater AFS Grain Fineness Number (for a given mean). The statistically normal distribution was next in order of protection while one which was skewed left and peaked gave poor results.

5. Metal penetration, in general, is less severe in the Revised Gertsman Test than in the Standard Gertsman Test.

6. Sand distribution is a factor which has an influence on metal penetration. An unwashed core made from a sand with a flat skewed right distribution was less penetrated than some of the coarser sands with core washes.

Acknowledgment

The Committee wishes to acknowledge thanks to the following companies for their cooperation in supplying materials and technical services: F. B. Stevens, Inc., Detroit; Foundry Sand Service Engineering Co., Detroit; Archer-Daniels-Midland Co., Minneapolis; Nugent Sand Co., Muskegon, Mich.; Central Silica Co., Zanesville, Ohio; Great Lakes Foundry Sand Co., Detroit; New Jersey Silica Sand Co., Millville, N.J.; Ottawa Silica Co., Ottawa, Ill.; Industrial Silica Corp., Youngstown, Ohio; Virginia Glass Sand Co., Berkley, West Virginia; Clayton Silica Sand Co., Clayton, Iowa; Dawes Silica Mining Co., Thomasville, Ga.; and Hardy Sand Co., Evansville, Ind.

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DISCUSSION

Chairman: D. F. SAWTELLE, Malleable Iron Fittings Co., Branford, Conn.

Co-Chairman: B. H. BOOTH, Carpenter Bros., Inc., Milwaukee, Wis.

Secretary: L. E. WILE, Lynchburg Foundry Co., Lynchburg, Va.

D. C. EKEY[1] *(Written Discussion):* The basic approach by the Committee seems excellent and the design of the test appears to have much forethought.

The work on "Sand Distribution" as an influence on penetration is of particular interest, and more specifically the use of statistical techniques to describe sand distribution is of interest. We suggest that the paragraph starting at the bottom of the first column of page 85 be rewritten to read: "Commercially available sands have not been described in the past by definitive statistical parameters. However, they do lend themselves to statistical parameters and such a system would be advantageous for both research and practical control applications. For this reason the Committee decided to design sand distributions to certain statistical parameters."

Apparently the Committee used the micron size for the cell divisions in the statistical calculations. Although the Committee may have a valid reason for using microns we feel that their use will in effect produce a weighted calculation introducing an inherent skewing effect. It would be of interest to see their calculations in the paper.

Considering this skewness effect of micron usage, we coded the cell intervals as a straight line function and ran skewness tests, with the following results:

Distribution	$\bar{x} = 300$ Normal	$\bar{x} = 300$ Skew Left	$\bar{x} = 200$ Skew Right
Skewness Factor	Plus 0.50	Plus 0.25	Plus 0.29

To discuss this slightly, the "$\bar{x}$ = Normal" distribution is more skewed then the "skewed" distributions and the "$\bar{x} = 200$ skew right" distribution is more skewed to the left than the "$\bar{x} = 300$ skew left" distribution.

It appears that this straight line cell interval method may be more valid in statistical calculations. However, the Committee may have their reasons for using microns (if that is what they used). Certainly the remainder of the work is of excellent quality and a definite contribution to better technology in the industry.

C. E. MCQUISTON *(Committee Reply to Dr. Ekey):* The Committee wishes to thank Dr. Ekey for his very provocative discussion. We agree that Dr. Ekey's suggested rewrite of the paragraph on statistical parameters of the sand grain distributions more clearly presents the Committee views than that given originally in the paper.

On page 86 in the first paragraph, it was stated: "All of the statistical parameters of these designed distributions were based on the arithmetic scale, i.e., they would not follow the geometric or standard screens, but rather linear intervals". Dr. Ekey has calculated the skewness of the distributions based on geometrical cell intervals whereas, as stated above, the distributions were designed on arithmetic cell intervals. Obviously, any distribution which is normal on an arithmetic cell interval will be positively skewed when calculated by geometric cell intervals.

The Committee chose to use the arithmetic cell interval for two reasons: (a) it gives equal weight to all diameters in the entire range used in the foundry, and (b) it allows consideration to be given to pan and clay-size material. On the other hand, the use of arithmetic cell interval is cumbersome for calculating purposes since screen analysis data are not obtained in this manner.

DR. D. V. ATTERTON[2] *(Written Discussion):* The tentative results of the Committee are similar to those being obtained by Sub-Committee T.S. 48 of the Institute of British Foundrymen who were engaged on a similar study with non-ferrous alloys. The main technique being employed by this Sub-Committee was similar in many ways to Gertsman's apparatus and consisted of having five standard AFS cores arranged symmetrically around a central down-gate and connected by ingates so that the metal flows evenly over the end face of each core. The height of metal above the test face can, of course, be varied as desired over a wide range.

On shakeout, the surface finish of the metal in contact with the standard cores is assessed in a semi-quantitative manner by comparing the roughness tactually and not visually as is the case with the present work using seven standard specimens varying in progressive stages of finish from very rough (designated 0) to very smooth (designated 6). The results obtained have found to be quite reproducible and the method has the added advantage that it does not involve the use of complicated or expensive equipment.

One of the main conclusions of the work to date is that there is a remarkably good co-relation between surface roughness and dry permeability of the molding sand being tested; quite a smooth curve had been obtained with tests on over 40 different molding and core sands. It was felt that in the case of the present work reported by the AFS Mold Surface Committee there was perhaps a more direct co-relation in their work between surface roughness and permeability rather than between surface roughness and the shape or skew of the distribution curve of the sand.

I was particularly interested to know that of the three washes tried on the cores, silica flour had apparently given the best surface finish. I was anxious to know whether this also gave the quickest peel. In some work which I had carried out myself I found that a distinction should be made between smoothness of finish and the ease of stripping, as I had found that on gray iron castings silica flour washes did tend to frit to the surface at times quite tenaciously. It would be interesting to know if the Committee had carried out any work on a combination of silica flour and graphite.

C. E. MCQUISTON *(Committee Reply to Dr. Atterton):* The Committee thanks Dr. Atterton for his comments and is gratified to learn of the general agreement between independently performed research. In regard to the correlation between surface roughness (penetration) and permeability, the Committee feels that permeability is a function of the distribution and thus the correlation exists between parameters of distribution and penetration. Permeability may serve as a good and practical measure of the distribution, however.

The Committee has done no further work on distributions or on the combination of silica flour and graphite for washes. Both should prove fruitful "avenues" for further research.

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1. Lebanon Steel Foundry, Lebanon, Pa.
2. Foundry Services Ltd., Birmingham, England.

COMPARISON OF LIQUID- AND AIR-QUENCHED PEARLITIC MALLEABLE IRONS

PART I — DUPLEX IRON

Report by

AFS Pearlitic Malleable Committee (6-E)*

Demand of design engineers for materials with higher yield strength and a generally better level of mechanical and processing properties has caused a marked increase in the use of pearlitic malleable iron. Current ASTM specifications for pearlitic malleable are given in Table 1. Increased use of pearlitic malleable has heightened interest in its methods of production and range of properties. Historically, three main methods of producing pearlitic malleable have evolved.[1] These methods include the following:

TABLE 1 — ASTM SPECIFICATIONS FOR PEARLITIC MALLEABLE IRON CASTINGS
(ASTM A220:55T)

ASTM No.	Typical Hardness Range, Bhn	Yield Point, Min, psi	Ultimate Strength, Min, psi	Elongation, % Min
45010	163-207	45,000	65,000	10
45007	163-217	45,000	68,000	7
48004	163-228	48,000	70,000	4
50007	179-228	50,000	75,000	7
53004	197-241	53,000	80,000	4
60003	197-255	60,000	80,000	3
80002	241-269	80,000	100,000	2

1. Alloying of standard malleable iron so that a pearlitic iron is produced with the normal malleabilizing anneal.
2. Reheat-treatment of normal ferritic malleable iron to produce a pearlitic iron.
3. Arrested annealing of malleable iron to produce pearlitic grades.

In the latter process, the iron is air cooled after first-stage graphitization and subsequently drawn at some subcritical temperature. These methods all produce engineering quality pearlitic malleable irons. However, the range of properties and casting section sizes in which they are available are presently limited in each of the three methods named. To obtain a higher level of properties and closer control, the industry currently is turning to heat treatments involving reheating or austenitizing first-stage graphitized iron, followed by liquid quenching, and then drawing. This is heat treating in the sense of hardening and tempering as commonly applied to other ferrous materials. In such heat treating, air quenching and liquid quenching from the austenitizing temperature produce different results.

By comparison with air quenching, liquid quenching followed by drawing appears to have the following advantages:

1. A higher yield strength is obtainable.
2. A narrower brinell hardness range is obtained from a given heat treating cycle. Less variation in mechanical properties results from the narrower brinell range.
3. Certain processing properties are improved. Response to selective hardening is better. Machinability of some liquid-quenched and tempered irons is better. These are all reasons for an increasing interest in reheating, liquid quenching and drawing as a process for producing pearlitic malleable. Air quenching is a simpler, more direct operation and is accomplished at the end of first stage graphitization in the "arrested anneal" process. Drawing then is performed after the air quench.

The purpose of this report is to set forth the properties obtained by the two methods of heat treatments and to clarify the differences existing between pearlitic malleable irons made by these two treatments.

Plant Data

The AFS Pearlitic Malleable Committee has obtained data for this comparison from its membership. Committee members from three different plants currently producing duplexed iron for pearlitic mal-

*Members of AFS Pearlitic Malleable Committee (6-E) are as follows: R. W. Heine, University of Wisconsin, Madison, *Chairman;* J. E. Foster, American Foundrymen's Society, Des Plaines, Ill., *Secretary;* W. M. Albrecht, Chain Belt Co., Milwaukee; A. H. Karpicke, Saginaw Malleable Iron Plant, Central Foundry Div., General Motors Corp., Saginaw, Mich.; J. E. Kruse, Albion Malleable Iron Co., Albion, Mich.; J. H. Lansing, Malleable Founders' Society, Cleveland; H. R. Saurer, Dayton Malleable Iron Co., Dayton, Ohio; Milton Tilley, National Malleable & Steel Castings Co., Cleveland; P. F. Ulmer, Link-Belt Co., Indianapolis, Ind.; and E. N. Wheeler, Belle City Malleable Iron Co., Racine, Wis.

leable have described these practices and their results in the following sections.

Plant A

Melting Practice

Molten malleable iron within the following compositional range, C, 2.30 to 2.40 per cent; Si, 1.58 to 1.68; Mn, 0.44; S, 0.140; Cr, 0.030; and P, 0.040 is obtained from a cupola-air furnace duplex operation. The iron is melted in one cupola at the rate of 27 tons per hour and run into a 60-ton air furnace for superheating and refining. The temperature of the iron at the air furnace spout is 2860 to 2900 F.

Both ferritic and pearlitic malleable iron are poured within the above compositional range. No addition of ferro-alloys is made after the iron leaves the furnace.

Annealing Practice

All castings to be heat treated to the various pearlitic malleable iron specifications are given a first-stage heat treatment in a controlled atmosphere continuous oven. They are air quenched when discharged from the furnace to within a brinell hardness range of 241 to 302 (3.9 to 3.5 mm). The first-stage cycle consists of heating to 1700 F in 3.5 hours, holding at 1700 F for 22 hours, dropping to 1550 F in 7 hours, and finally air quenching.

ASTM grades 45007, 48004, and 53004 are drawn in a belt-type continuous furnace at oven temperatures of 1345, 1320, and 1300 F for 4.25, 3.15 and 2.60 hours respectively.

Grades 60003 and 80002 are reheated to 1600 F for 25 minutes and quenched in circulating oil to within a brinell hardness range of 477 to 600 (2.8 to 2.5 mm). The oil-quenched castings are drawn at temperatures of 1200 and 1300 F respectively for 2.60 hours. The castings are drawn in the same type furnaces as for the air quenched material. Table 2 presents the properties obtained from these practices at Plant A.

Plant B

Melting method consists of cupola – air furnace. The air furnace is fired with pulverized coal and/or oil. The average analysis of the product is 2.50 per cent C, 1.25 per cent Si, 0.50 per cent Mn, 0.170 per cent S, 0.08 per cent P, and 0.0015 per cent B. The cupola charge consists of 0 to 10 per cent malleable pig, 25 to 40 per cent steel scrap and the balance sprue and malleable scrap and alloys of manganese and silicon.

Intermittent furnaces of 5-ton capacity are used in the interest of flexibility of heat treatment as well as types of castings. With five different specifications on brinell range and castings varying from 0.40 to 100.00 lb, 15 different heat treatments, including the heats rejected for being too hard and too soft, as well as air quench and liquid quench are employed. These furnaces also handle pearlitic malleable made by alloying and arrested anneal as well as malleabilizing. They also handle strain-relieving of hard iron and drawing of welded castings. The castings are packed on grids and handled into and out of furnaces and quench and storage by gantry crane.

TABLE 2 – AVERAGE MECHANICAL PROPERTIES*

ASTM Designation	Brinell Hardness Diameter	Tensile Strength, psi	Yield Point, psi	Elongation, %	Quench Medium
80002	3.76	112,280	96,160	4.06	Oil
60003	4.15	96,640	76,360	7.59	Oil
53004	4.18	92,690	61,260	6.96	Air
48004	4.24	91,510	60,770	7.43	Air
45007	4.48	80,720	55,495	9.16	Air

*Gage length of test bar is machined.

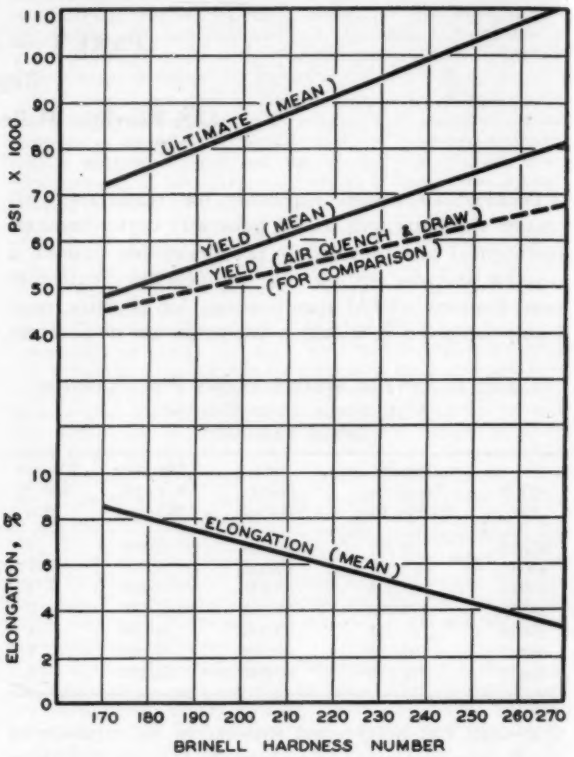


Fig. 1 — Plant B pearlitic malleable iron mechanical properties; iron was air cooled after first stage anneal, reheated, liquid quenched and drawn.

The grids are moved from trim bench to heat treat and then to finishing by cars. The mechanical properties, as depicted by the standard malleable test bar, appear in Fig. 1. The yield strength varies with the method of heat treating and quenching as shown by the dashed lines on Fig. 1.

Standard deviation values applying to Fig. 1 are given below.

STANDARD DEVIATION – ONE SIGMA (Fig. 1)

Yield point	– 1550 psi
Tensile strength	– 3300 psi
Elongation	– 0.60%

Plant C

All iron is melted in three continuous tap cupolas with metal flowing into 15-ton forehearths for mixing and desulphurizing. The cupola charge consists of high grade steel scrap, sprue, pig iron, ferrosilicon, ferromanganese, limestone and coke. The molten metal is tapped from the forehearth and carried by

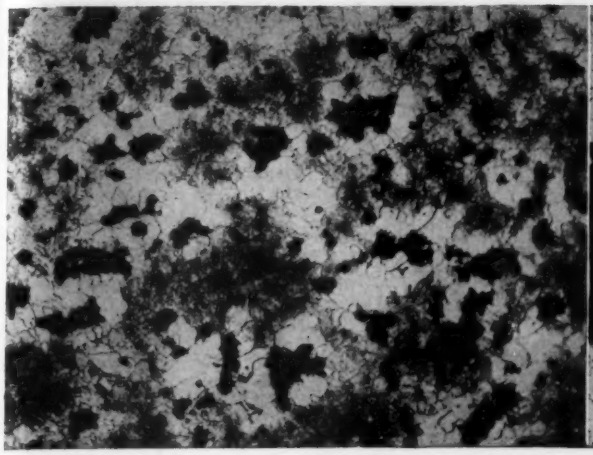


Fig. 2A — Air quenched and drawn to 4.70 mm brinell diameter; etched, 100X; ferrite and pearlite matrix.

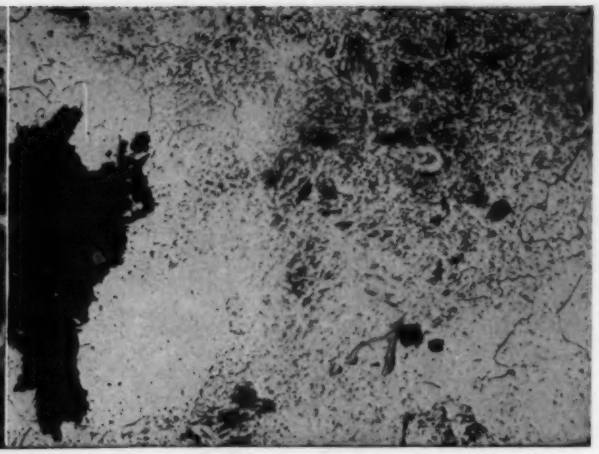


Fig. 2B — Air quenched and drawn to 4.70 mm brinell diameter; etched, 500X; ferrite and pearlite matrix.

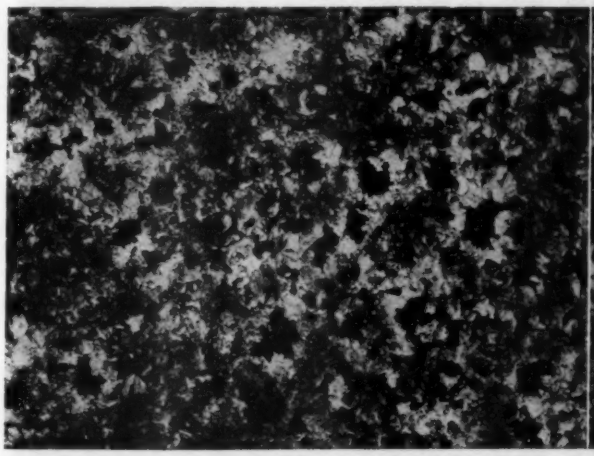


Fig. 2C — Air quenched and drawn to 3.90 mm brinell diameter; etched, 100X; pearlitic matrix.

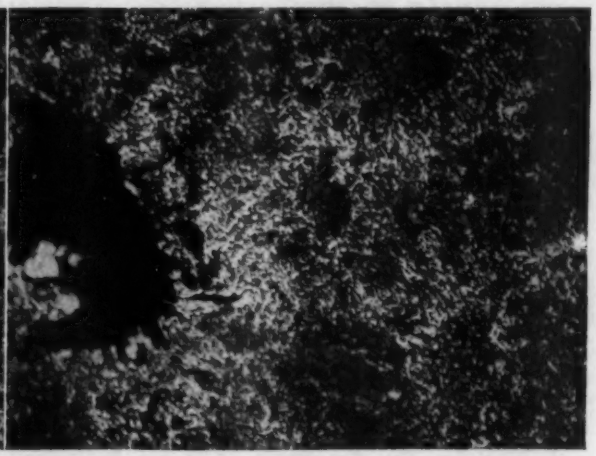


Fig. 2D — Air quenched and drawn to 3.90 mm brinell diameter; etched, 500X; pearlitic matrix.

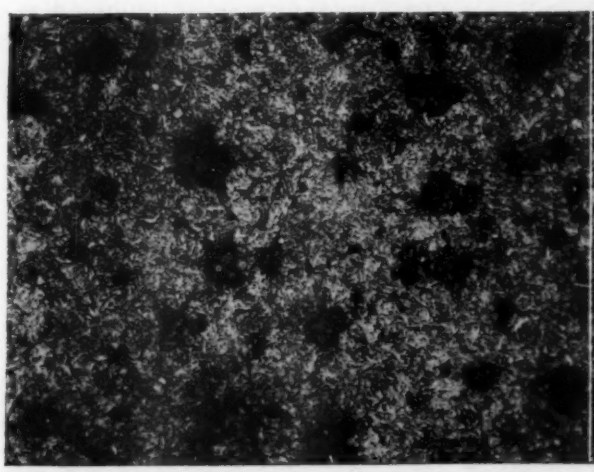


Fig. 2E — Reheated to 1600 F and oil quenched, 2.70 mm diam; etched, 100X; martensitic matrix, ½ in. section.



Fig. 2F — Reheated to 1600 F and oil quenched, 2.70 mm diam; etched, 500X; martensitic matrix, ½-in. section

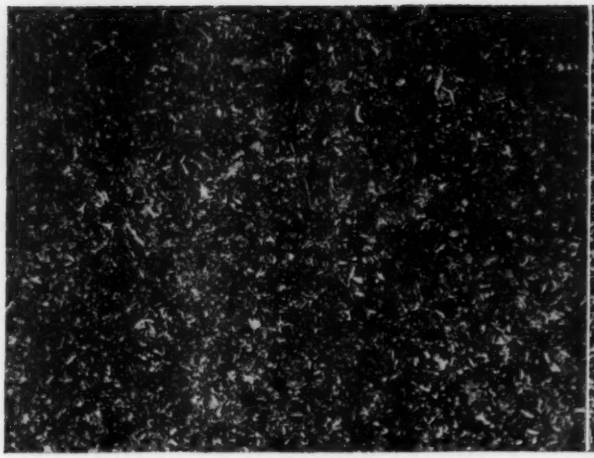


Fig. 2G — Oil quenched and drawn to 3.7 mm diam; etched, 100X; tempered martensite, ½-in. section.

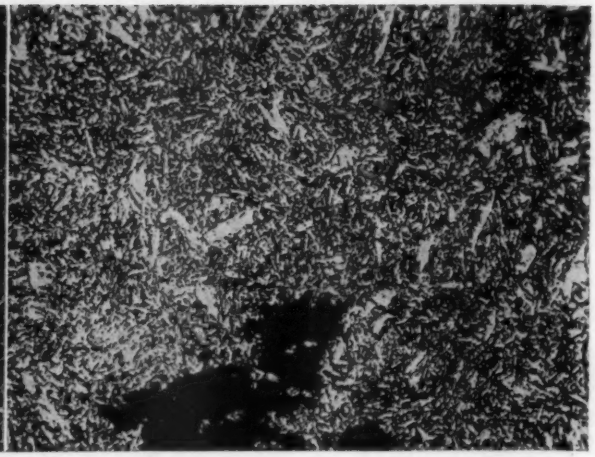


Fig. 2H — Oil quenched and drawn to 3.7 mm diam; etched, 500X; tempered martensite, ½-in. section.

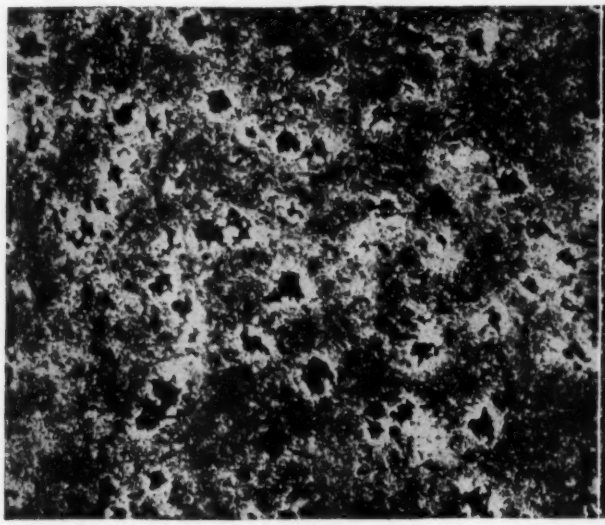


Fig. 2I — Microstructure of light section pearlitic malleable iron air quenched; etched, 100X.

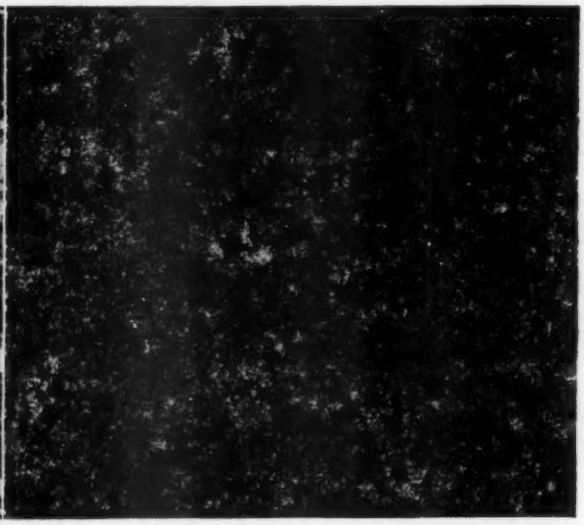


Fig. 2J — Microstructure of light section pearlitic malleable iron liquid quenched; etched, 100X.

4-ton transfer ladles to three electric furnaces, which handle 110 to 120 tons an hour. These furnaces superheat the metal to a controlled temperature of 2750 F to 2775 F to produce a highly refined melt. After the molten metal has been heated to proper pouring temperature in the furnace the metal is tapped into 1-ton transfer ladles that carry metal to small covered ladles at the mold pouring stations.

After the molds are poured they are shaken out on an automatic shakeout. The entire gate of castings is transferred by a conveyor from the shakeout to the hard iron room. In the hard iron department the sprue and runners are removed from the castings for remelting, and the castings are sorted for packing into the annealing furnaces.

The basic composition of the white iron is as follows: C, 2.55-2.65; Si, 1.35-1.45; Mn, 0.40-0.46; S. 0.11-0.12; P, 0.05-0.06; Cr, 0.03 max; and B, 0.0015-0.002. The cupola operation is controlled to obtain iron within the above specifications. Very little change in this base analysis takes place in the electric furnaces. The humidity of the cupola air blast is automatically controlled by passing the air through a humidity control unit. The moisture is held at about 57 grains per pound of air or about 4.6 grains per cubic foot of blast. The blast is heated by external gas-fired heat exchangers. The blast temperature will vary from 300 F to 425 F depending on conditions encountered.

The heat treatment of the white iron to produce pearlitic malleable iron is done in two stages, and for a few jobs in three stages. In the first stage the castings are packed into heat resistant alloy trays and are pushed through a continuous type radiant tube gas-fired atmospheric controlled tunnel furnace. The total time in this first stage is 19.3 hours with

the following times at various temperatures:

Room Temp to 1400 F	2 hr, 12 min
1400 F to 1525 F	1 hr, 12 min
1525 F to 1630 F	1 hr, 46 min
1630 F to 1680 F	1 hr, 25 min
1680 F to 1750 F	2 hr, 37 min
1750 F — Hold	5 hr, 30 min
1750 F to 1720 F	2 hr, 20 min
1720 F to 1650 F	2 hr, 15 min
	16 hr, 197 min or 19 hr, 17 min

All pearlitic malleable iron is air quenched from 1650 F. The gate connections are removed by grinding and then the castings are tempered to one of two brinell hardness ranges. These two ranges are 197 to 241 bhn (3.9 mm to 4.3 mm) and 163 to 207 bhn (4.2 mm to 4.7 mm). Tempering of the castings is done in a continuous tray type recirculating heat furnace. To produce the harder material it is in the furnace for a total time of 6 hr and 40 min at a temperature ranging from 1270 F to 1290 F with the castings at heat about 3½ to 4 hr. The softer material is produced in the same type furnace using the same time cycle but temperatures ranging from 1320 to 1340 F.

Figures 2a through 2d are photomicrographs of the structures of material drawn to some of the above described ranges. The air-quenched structures are pearlitic or spheroidized partially.

A third type of pearlitic malleable iron is produced by taking the air-quenched material and re-heating to 1600 F in a continuous-belt-type direct-fired furnace, holding at heat for 30 to 40 minutes, then quenching in oil. After oil quenching the material is tempered in a continuous-belt, recirculating-heat-type furnace. The time in the furnace is 3½ hours. The temperature will range from 1170 F to 1190 F and the hardness range is 241 to 269 bhn (3.7 mm to 3.9 mm) Figures 2e through 2h are photomicrographs of the martensitic and tempered martensitic structures. The liquid quenched structures have the nature of martensite and tempered martensite or transition products depending on section size.

Tabulated below are the specification standards and average mechanical properties of the three types of pearlitic malleable produced at Plant C.

Table 3 — Mechanical Properties, Plant C

BHN	Yield, psi Min	Ultimate, psi Min	Elong., % in 2 in.
Specification Standards:			
197 - 241	60,000	80,000	3.0
163 - 207	48,000	70,000	5.0
241 - 269	80,000	100,000	2.0
Average or Common Ranges:			
197 - 241	60,000 - 68,000	80,000 - 95,000	3.0 - 5.0
163 - 207	48,000 - 58,000	70,000 - 80,000	4.0 - 6.0
241 - 269	80,000 - 90,000	100,000 - 110,000	2.0 - 3.0

The modulus of elasticity is 26,000,000 psi to 28,000,000 psi. Density is 0.266 pounds per cubic inch. Specific gravity is 7.37. Linear coefficient of expansion is 0.0000066 in. per in. per °F.

Analysis of Plant Data

Study of the data from the three pearlitic malleable producers reveals several important facts.

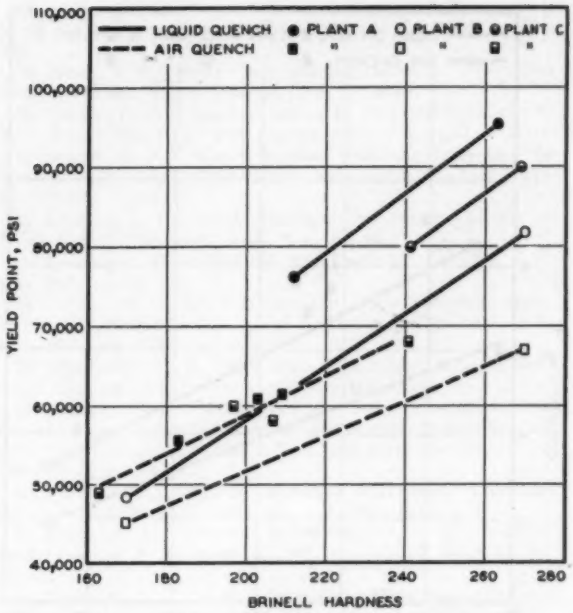


Fig. 3 — Comparison of yield point of air- and liquid-quenched irons drawn to equivalent hardness.

Yield Point

The yield point of air- and liquid-quenched irons that have been drawn to equivalent hardness is compared in Fig. 3. The data plotted is from Table 2, Fig. 1, and Table 3. In all three plants, ASTM grade 80002 iron is obtained by liquid quenching from about 1600 F and then drawing at a temperature of 1170 to 1200 F. Grade 60003 iron is liquid quenched in Plants A and B and air quenched in Plant C. Thus the latter grade appears to be a breaking point above which the liquid quenching practice is favored for obtaining higher yield strength. Figure 3 shows that the yield strength of liquid-quenched iron is higher compared with the air-quenched iron from Plants A and C after drawing to a given hardness. It is an established metallurgical principle for heat-treated steels that a tempered martensitic structure produces higher yield strength at a given hardness than does a tempered pearlitic structure of the same hardness.[2] The principle applies equally to pearlitic malleable. It is also true that the effect is more pronounced at higher hardness levels as evident in Fig. 3. The higher yield point at higher hardness of iron from Plant A compared with B and C is probably due to differences in per cent carbon, per cent silicon, and per cent sulphur in the base metal composition of the three irons.

The yield point of air-quenched and drawn pearlitic malleable should not be expected to agree between the three plants, due to the characteristics of the response of pearlitic structures to tempering. Differences in base metal composition, plant equipment, casting section, severity of air-quenching; handling, etc., will affect comparisons of air-quenched irons.

It should be recognized that Fig. 3 represents typical properties of pearlitic malleable being currently produced.

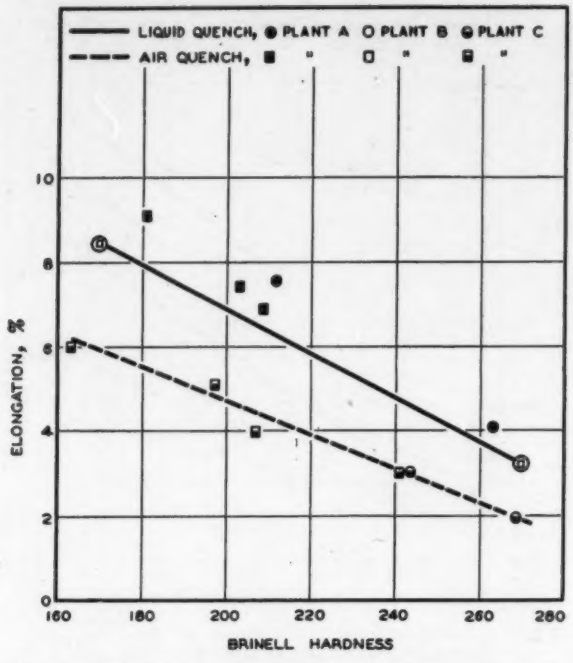


Fig. 4 — Comparison of elongation of air- and liquid-quenched irons drawn to equivalent hardness.

Elongation

Elongation of air- and liquid-quenched irons at equivalent hardness is shown in Fig. 4. There is some indication that the liquid quench produces a higher elongation at high hardness than air quenching does. However, because numerous other factors such as base metal composition, melting and heat treatment practice, graphite character, etc., affect elongation markedly, good correlation cannot be expected. The expected elongation range is indicated in Fig. 4 for the current hardness range of pearlitic malleable.

Tensile Strength

Tensile strength of air- and liquid-quenched irons at equivalent hardness is shown in Fig. 5. Tensile strength is a dependent mechanical property being determined by the combination of yield point and elongation prevailing. The relation of hardness and tensile strength shown in Fig. 5 is therefore to be expected from the results for yield and elongation depicted in Figs. 3 and 4. No clear-cut comparison of tensile strength in oil- and air-quenched irons is possible as in the case of yield point because of the mitigating effect of elongation. Of course, the higher tensile strength required of the 80002 irons is obtained by liquid quenching and drawing.

Modulus of Elasticity

Modulus of elasticity in tension values reported are 26 to 28 million psi for Plant C and 26 million psi for Plant A.

Density

Plant C reported a density of 0.266 pounds per cubic inch and a specific gravity of 7.37 for pearlitic malleable.

Linear Coefficient of Expansion

Plant C reported a linear coeffecient of expansion of 0.0000066 in. per in. per °F for pearlitic malleable.

Machinability

No specific data are provided. User reports indicate superior machinability of liquid-quenched and drawn pearlitic malleable at high hardness levels. This is attributed to the spheroidized nature of the tempered martensite structures which are usually considered to have better machinability than fine pearlite at high hardness. See reference 5 for further details.

Selective Hardening

The selective hardening characteristics of pearlitic malleable are excellent. References 3 and 4 are research reports describing the techniques of selective hardening pearlitic malleable.

Summary

Metallurgical data from three producers of air- and liquid-quenched pearlitic malleable irons have been presented. Reheating, liquid quenching, and then drawing has been shown to be successful means of producing ASTM 60003 and 80002 grades of pearlitic malleable. Air quenching and drawing is successfully used to produce ASTM grade 60003 and lower yield strength pearlitic malleable such as grades 53004, 48004, and 45007. Liquid quenching and drawing has been demonstrated as able to produce higher yield strength at a given hardness than quenching and drawing. Numerous properties of pearlitic malleable iron have been tabulated.

It should be recognized that the data presented in

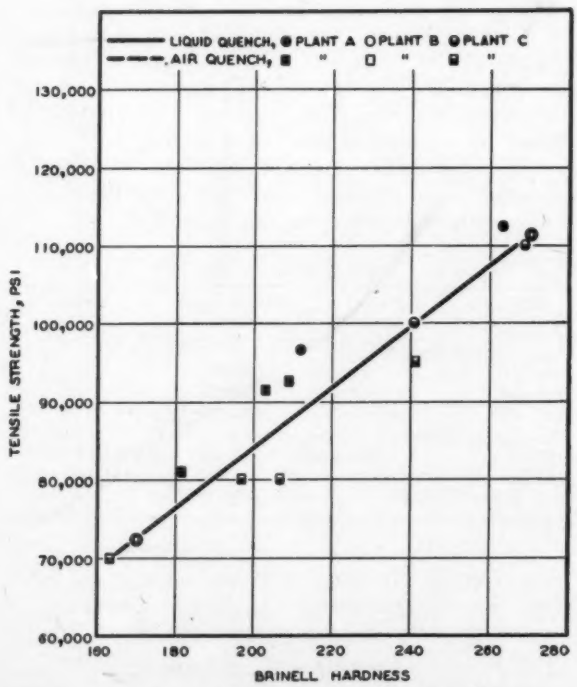


Fig. 5 — Comparison of tensile strength of air- and liquid-quenched irons drawn to equivalent hardness.

this report apply to duplexed irons melted in the cupola and of differing base metal compositions. The Committee intends to report data in the future for pearlitic malleable produced with base metal from the cold-melt process and from producers of ASTM 35018 base iron. A bibliography of papers dealing with pearlitic malleable iron is appended.

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A SURVEY OF THE CO_2 PROCESS

By

J. E. Huss*

Core Test Committee (8-C)** of the AFS Sand Division was assigned the project of objectively evaluating the CO_2 process as it has been utilized in foundries in the United States and Canada. This Committee decided that the best way of making this study would be to obtain the facts of this process from foundrymen who had actual experience with it.

Committee Used Questionnaires

In order to make a valid evaluation of this foundry process introduced from Europe as a new method for hardening cores and molds, the Committee distributed questionnaires to more than 100 foundries from coast to coast that had tried the CO_2 process. Committee (8-C) expresses its appreciation to the foundries that cooperated and to those who helped distribute, fill out, and collect these questionnaires.

Twenty-one foundries returned the CO_2 process questionnaires and nine others replied by letter for a total of 30 foundries located in 10 different states and Canada. Most foundries reporting were jobbing with some being semi-production shops. No high rate of production of cores or molds by the CO_2 process was reported on any of these questionnaires, all of which were completed by March, 1955.

The 21 questionnaires included information on 26 different castings and of these, five were of steel (both carbon and high-alloy), nine were gray iron, four were malleable iron, four were brass or bronze, three were aluminum, and one was nickel.

Steel Castings

The wide range of application of the CO_2 process for hardening cores and molds is further demonstrated when each metal is examined by itself. In steel the castings varied in weight from 1 to 100 lb and included such castings as grader arms, pump casings and trailer wheels with section sizes from 1/4 in. to 3 in. and a reported tolerance of 1/32 in.

Gray Iron Castings

In gray iron, the castings weighed from 5 lb to 5,000 lb, increasing in size from the wheel bearings and spacer spools, valve bodies and gear cases, to lathe beds and machine tool bases. The section sizes varied from 1/4 in. to 3 in. with best reported dimensional tolerance of 1/128 in.

Malleable Iron Castings

This same tolerance of 1/128 in. was given as possible for malleable iron castings which varied in section size from 1/8 to 2 in. and in weight from 2 to 100 lb. These castings include truck and railroad parts and pressure valves.

Brass and Bronze Castings

The brass and bronze castings include pressure castings, pump fittings and meter housings weighing from 3 to 20 lb. with section sizes from 1/4 in. to 2½ in.

Aluminum Castings

The aluminum castings were aircraft parts, meter bodies, and pressure castings weighing from 2 oz to 200 lb. The best reported tolerances were 0.01 to 0.03 in. with section sizes varying from 1/8 to 1 in.

How Cores Were Made

In the specific jobs reported by the foundries, 6 molds and 26 cores, of which 6 are listed as external cores, are hardened by CO_2 gas. In the cases reported the cores were made as follows: ten by hand ramming, five by blowing, and three by jolting. For the molds: two were hand-rammed, three were made on molding machines, and one was slinger rammed.

While 15 foundries reported no facings used, three shops were using regular core sand for facings. Of the foundries reporting, 59 per cent stated that no washes were needed. Of the foundries using washes, three reported using water-base washes and eight reported using solvent-base (alcohol) washes, two foundries using both types.

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Molding Sand Mixtures Used

The reported sand mixtures specified dried sand with less than 0.8 per cent clay content. Grain size numbers varied from 50 to 115 AFS GFN, while the sand temperatures were all listed as room temperature with a maximum allowable temperature of 100 F. These wide variations in the fineness of the sand and desired physical properties cause considerable differences in sand mixtures from foundry to foundry.

The amounts of sodium silicate binder used varied from 2-3/4 to 6 per cent by weight of sand. One aluminum, one steel, and four gray iron foundries reported using no other additives, while the following materials were used by one or more foundries:

Additives Used to Improve Collapsibility	Additives Used to Improve Surface Finish
Asphalt Emulsions	Gilsonite Compound
Cellulose Fibers	Silica Flour
Cereal Binder	Graphite
Gilsonite Compound	Iron Oxide
Graphite	Sea Coal
Sea Coal	

The cost of these sodium-silicate-bonded-sand mixtures also vary appreciably; a low of seven cents per 6-lb core to a high of $30.00 per 200-lb mixture were reported. However, the majority of foundries reporting on the costs of these silicate sands said they were somewhat higher in cost than their conventional sand mixtures.

Sand Mixing Cycle Used

Majority of the foundries reporting also use a wheel-type mixer for mulling these silicate-bonded sands. The average mixing cycle is as follows: mix the dry ingredients for 1 minute, add the sodium silicate binder and mix an additional 4 minutes for a 5-minute total mixing cycle. The moisture content of the mixed sands range from 2 to 4 per cent with the bench life of these sands varying from ½ to 3 hr. unless covered. Local conditions of temperature, humidity, air movement, and the amount of carbon dioxide in the atmosphere are the determining factors on bench life.

One foundry oven-dries internal cores, two others oven-dry their washed cores, but 85 per cent reported they were not oven-drying their CO_2 hardened cores and molds. Surface hardnesses of 80 to 95 scratch hardness were listed as were tensile strengths of 35 to 115 psi on the hardened cores. Other questionnaires stated the hardness and tensile strengths were less than those of corresponding conventional core sands. Six foundries mentioned there was less gas evolution from the silicate-bonded sands than from their regular core sands.

Collapsibility of Cores

The collapsibility of the CO_2 hardened sand is adequate when burn-out additives are used, and the majority stated they had no shake-out problems.

The costs of the CO_2 gas used varied widely from 4 cents to 15 cents per pound. Only two foundries were using CO_2 gas generators while 18 reported using bottled gas. Sixty-one per cent of these foundries said it was not necessary to heat their CO_2 valves. One foundry stated that the use of extra-dry carbon dioxide eliminated the need for heating the valves.

Gassing of Cores and Molds

The methods reported for gassing the cores and molds were all manual methods including the use of rods, cups, vents, gasket seals, etc. Seven foundries found it necessary to have special venting in their patterns or boxes.

Both the gassing pressures and the amounts of gas used per unit amount of sand varied widely. Gassing pressures from 8 to 50 psi were given and foundries reported using from 1 lb of gas for 8 lb of sand to 1 lb of gas for 120 lb of sand, or 15 times as much gas in one foundry as compared to another.

Storage of Hardened Cores

CO_2 hardened cores and molds are stored from 15 minutes up to 4 weeks before they are poured. Forty-four per cent of the reporting foundries have trouble with moisture pick-up on stored cores, and one foundry discontinued using the CO_2 process because of this problem.

Seven foundries reported returning the silicate-bonded sands without reclamation to their sand system, but only one of these returned the silicate sand to core sand. None of the foundries reported contamination of their sand by the Na_2CO_3 and some had been re-using the silicate sand for as long as six months. Also, none of the foundries reported using a sand reclamation unit.

Advantages Reported for the CO_2 Process

1. No baking ovens or core driers needed and more capacity from the same floor area,
2. More accurate dimensions and smoother finish on the castings,
3. Less gas evolution and good shake-out,
4. Eliminates mold drying.
5. Rapid production of finished cores to expedite rush orders,
6. Saves money and space,
7. No odor or smoke from the process, and
8. Cores can be hollowed out before gassing to make shell cores.

Disadvantages Reported for the CO_2 Process

1. Stickiness in the core box,
2. Poor workability of the sand,
3. Poor shake-out,
4. Poor handling ability and loss of hardness of cores on storage because of moisture adsorption, and
5. At the present time this process is not adaptable to high-production foundries.

Conclusions

From the information obtained from these questionnaires it is readily apparent that in its first year of introduction into America, the CO_2 process has been utilized for the production of many different types of castings, both ferrous and non-ferrous. Only time will tell if this process can be satisfactorily adapted to high-production work so that it will occupy an even more important place in American foundries.

TRANSIENT HEAT FLOW

By

Victor Paschkis*

Over the last several years the AFS Heat Transfer Committee has presented a number of papers on solidification and related problems. Solidification represents also a *thermal* problem. In order to present a unified picture, the Heat Transfer Committee suggested a series of papers by which the metallurgist and foundry engineer may be led to a better understanding of some thermal aspects of his work, which aspects he frequently can treat only in passing.

All foundry work is based on heat transfer, and almost all foundry problems are of transient nature; i.e., temperatures change with time whereas in steady-state, temperatures in the body at any position are the same at all times.

Except for specialists, heat transfer is frequently treated as if it involved only steady-state problems. This, together with the mathematical complexities of transient problems, has resulted in a certain awe in which transient problems are held. But, as will be seen, there is really nothing mysterious about transient problems.

There are *three modes* of heat transfer: conduction, convection, and radiation. All three depend on a temperature difference as driving force; but the importance of the temperature difference is different for the several modes.

Conduction takes place in solid bodies or in fluids. Heat transfer by conduction is independent of any movement of the body in which the heat transfer occurs.

Convection can take place only in liquids or gases and consists of the transport of heat by moving mass: particles of the warm fluid cool and thus transfer the heat which they have picked up elsewhere to cooler parts.

Radiation takes place between solid surfaces or, under certain conditions, between bodies of gas, or finally between such body and a solid surface.

Since heat storage is the characteristic feature of transient phenomena, heat transfer by radiation is not subject to distinction between transient and steady-state conditions. Although transient conditions can exist in convection, they are practically of little significance, and consideration of transient behavior is limited to conduction.

Prior to exploring this concept further, it is well to define some terms. *Heat* is a form of energy while *Temperature* is a potential; the two terms are frequently mistaken. Temperature is meaningful only with regard to a reference point, e.g. in the centigrade scale the freezing point of water (or melting point of ice) is used as such reference. Temperature can be compared to the level of a fluid, as shown in Fig. 1.

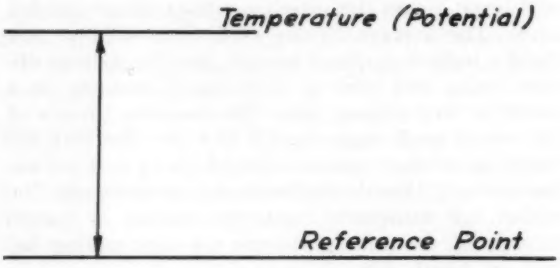


Figure 1

Heat content of a body of a given material is proportional to its temperature and its weight. In Fig. 2 the size of the square indicates heat contents; underneath each block a thermometer and a weight show the values of the factors contributing to a certain

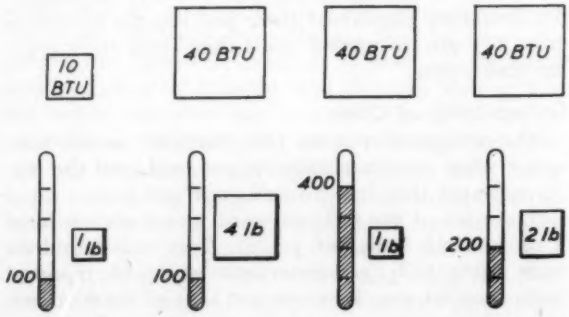


Figure 2

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heat content. All four cases refer to bodies of the same material. A 1-lb body at 100 degrees may have a heat content of 10 Btu. If a 4-lb body (same material) is at 100 degrees the heat content is 40 Btu. The same heat content obtains with a 1-lb body at 400 degrees or a 2-lb body at 200 degrees.

Wherever temperature differences exist, heat flows from the hotter to the colder point. The amount of heat transferred per unit time (Btu/hr) is called the *rate of heat flow.* The rate of heat flow can be illustrated conveniently by a tube (Fig. 3), the cross section of which indicates the magnitude of the rate of heat flow. If between two points in a body a temperature difference exists, this difference, divided by the distance of these points (F/ft), is called the *temperature gradient.*



Figure 3

Now it is possible to define the term of *thermal conductivity.* This is a property of materials and indicates the rate of heat flow per unit area, under the influence of a unit gradient of temperature. Thus the unit of conductivity is Btu/sq ft, hr, (F/ft) or, simply, Btu/ft, hr, F.

Specific heat, another important property, is the increase of heat content of a piece of 1 lb, if its temperature is increased 1 degree. Thus, in the example of Fig. 2 the body shown has a specific heat of 0.1;

specific heat x weight x temperature = heat content
0.1 x 4 x 100 = 40

In steady-state the rate of heat flow does not change with time; but of course in a given system, different rates may obtain at different parts of the body. In order to compare such rates, it is convenient to refer them to 1 sq ft. In way of example, consider Fig. 4,

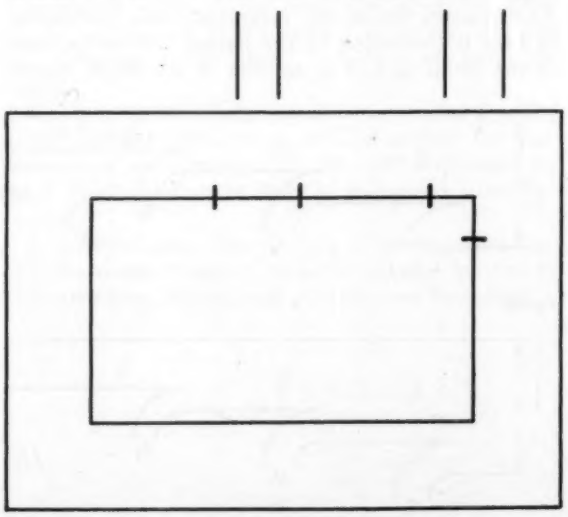


Figure 4

showing schematically the cross section through a continuous annealing furnace. The rate of heat flow through 1 sq ft of inside wall area is smaller in the center of the wall than at the corner, the difference being indicated by the two tubes of different size drawn over the respective parts of the furnace. Since this diagram is drawn for steady-state the diagram for any time would show the same picture; the two tubes do not vary with time.

In transient state, conditions are basically different and will be explained first using the simple case of heating up a slab as example. Reference is made to Fig. 5a, showing the slab, which is initially at room temperature. It is assumed that the slab is so big as to make end effects negligible. The same could be achieved by assuming that the sides are completely insulated; this is indicated in the drawing by shading the perimeter. Assume now that the slab is being heated on one side, while the other faces the shop. Such conditions would prevail in heating a ladle from the inside, while the outside faces the shop.

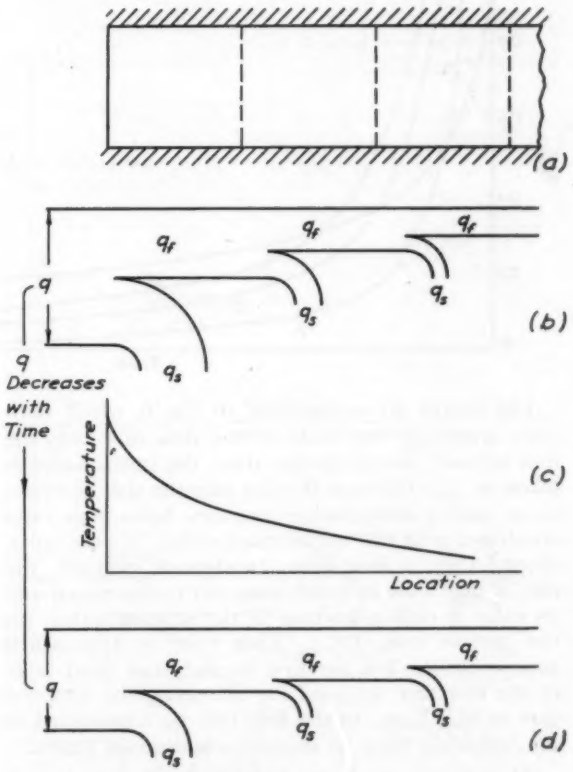


Figure 5

Consider the wall divided into a number of layers as shown. The heat flow at an early time is illustrated in Fig. 5b. The flow q coming from the heat source on arrival at the first layer is split into two parts; q_s and q_f. One part, q_s, is used to raise the temperature of this first layer, and the balance, q_f, flows to the second layer. The flow received here is again split into two parts: one to raise the temperature of the second layer and the balance flowing to the third. This continues through all layers, as illustrated in Fig. 5. This figure also indicates that the

greatest part of the heat flow received in each layer goes to raise its temperature, and only a smaller part flows on. Thus, in the example shown, practically no heat flows beyond the third layer.

Figure 5c indicates the temperature distribution at this first time step. The transient nature of the problem makes itself felt by the fact that after a very short time interval, that is to say in a second time step, the distribution of the flow changes; the second set of tubes, Fig. 5d, illustrates the new distribution. Less heat is used to raise the temperature of the first section, and thus more heat flows into the second; similarly the rate of heat flow into the third section is increased, and while more heat is used there to raise the temperature, enough is left to flow to the fourth section. The new temperature distribution is again indicated (Fig. 5e). It should be noted that the total rate of heat flow to the first section is smaller than in the first step.

of flow. Now the value of q remains unchanged for all time steps and is equal to the constant rate of heat flow introduced. It should be noted that Figs. 5b and 7a represent the same time after start of heating; Figs. 5d and 7b show conditions at a time twice as long after start of heating as in 5b and 7a.

In Fig. 8 one can summarize again the distribution. Times are plotted as abcissas, and ratios of the various q-values to the constant q total as ordinates; all ordinates lie between 0 and 1. The values of stored heat (q_{s1}, q_{s2}, etc.) have a different characteristic from those of the rates of heat flowing through the section (q_{f1}, q_{f2}, etc.). q_{s1} starts with the value of one and decreases, as q_{s1} starts with the value of one and decreases as q_{f1} increases. The curves q_{s2}, q_{s3}, etc. all start at zero and gradually increase to a maximum and then decrease again.

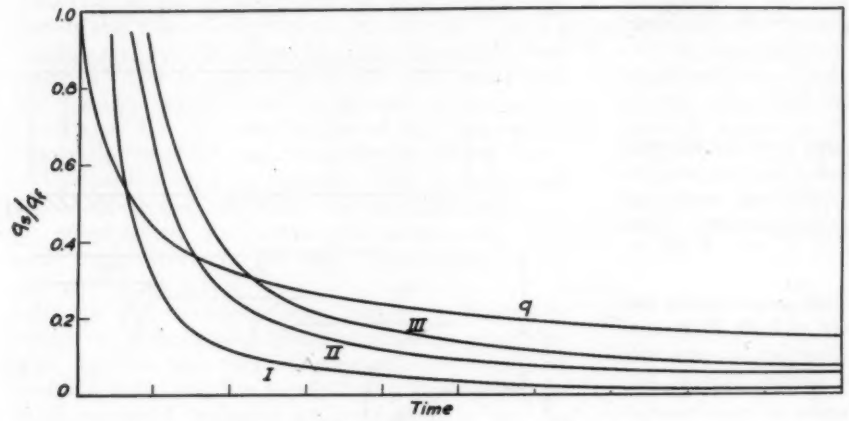


Figure 6

The results are summarized in Fig. 6, which indicates (curve q) the width of the flow tube into the first section, plotted against time; the initial value is taken as "1". The heat flowing into the slab decreases as its surface temperature increases. Since flow rates are drawn with the initial value called "1", the curve drops. After a long time, steady-state prevails; the rate of heat flow in steady-state can be computed and its value is only a fraction of the original value (in the present case, 10%). This value is approached asymptotically. For the first, second, and third layer of the flow rate q_s (used for increasing the temperature of that layer) to the flow rate q_f, transmitted to the following layer, is shown in additional curves.

At the start all heat reaching the first section is used for storage, and no heat flows to the second section. Thus the curve (I) starts at infinity and drops down rapidly. After a very long time (not shown in the graph) the section is saturated with heat; no heat is absorbed and all heat entering flows on to the next section. The curve approaches the ordinate value of zero. Similar considerations hold for other curves.

Now consider another case: the same slab to be heated, but this time by introducing a constant rate of heat flow to one surface. This can be achieved, for example, by induction heating the one side of the slab. Figure 7 illustrates the subsequent distribution

Solidification

The above concepts for transient heat flow can also be very useful in studying solidification. The rate of heat flow out of the casting q_c decreases with time. Reference is made to Fig. 9, which is similar to Figs. 5 and 7; however only one time step is shown. As explained below, the magnitude and distribution of flow is controlled by the nature and temperature of the metal as well as by that of the mold. Figures

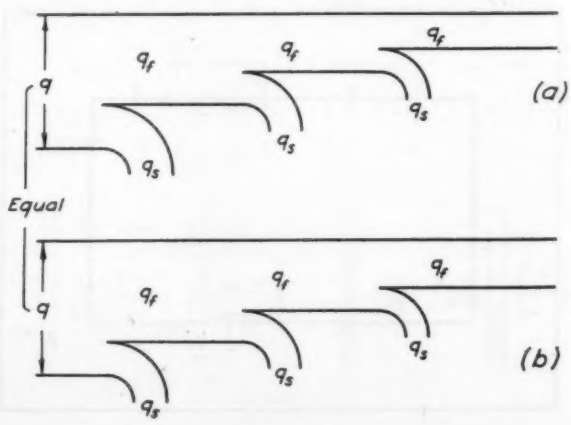
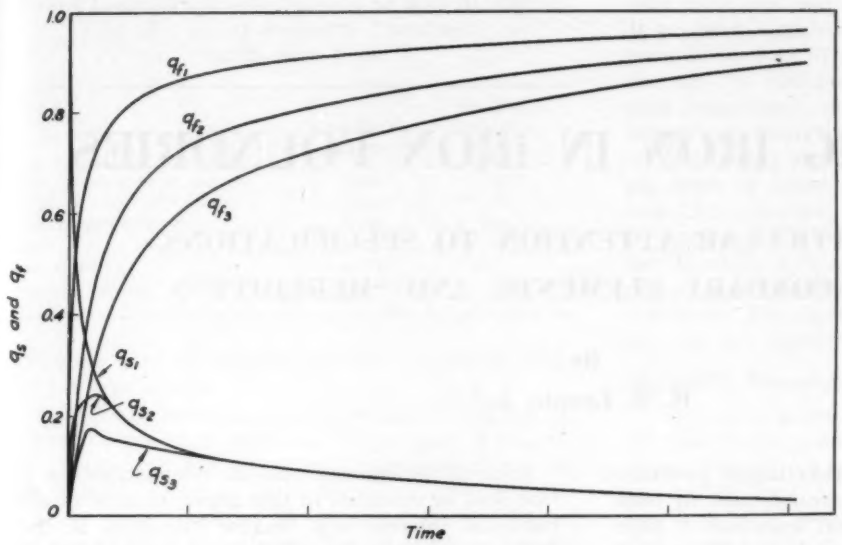


Figure 7



(Left)
Figure 8

(Below)
Figure 9

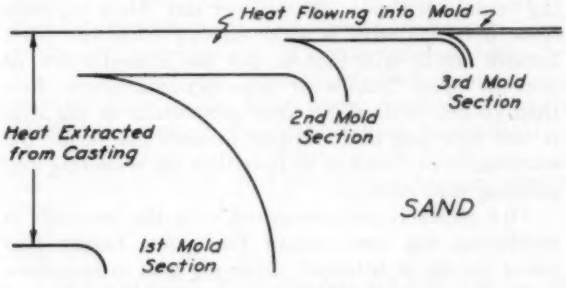
9a and 9b refer to sand mold and chill mold respectively. The flow goes into the mold; initially most of the heat is absorbed in the first layer (q_{s1}) and only little flows on (q_{f1}), etc.

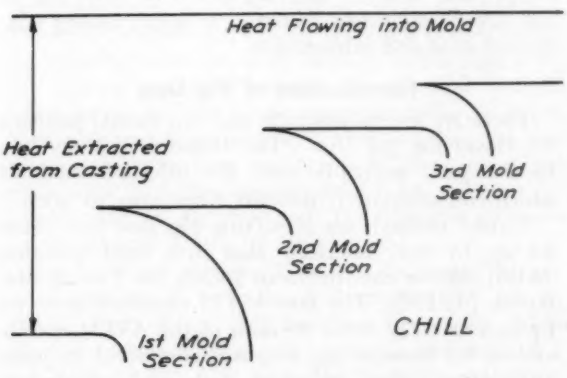
But one can now become more specific. The total heat to be extracted from the casting can be computed from the specific heat, heat of fusion, degree of superheat, weight of the casting, and the shakeout temperature. The latter is, of course, not uniform throughout the casting, but that need not be considered at present. The greater the rate of heat flow is out of the casting, the faster it will solidify.

Now, what makes for high q_{c1}? High q_{c1} obtains if the temperature of the casting is high, compared to that of the mold; with constant mold temperature high q_{c1} implies high melting metals and high superheat. The temperature of the mold will rise rapidly (thus making for a marked drop of q_c as compared to its original value), if q_{s1} is large and q_{f1} small. This occurs if the mold has a low thermal conductivity (sand-Fig. 9a): heat will not readily flow through the mold to the flask; the innermost layer of the mold, next to the casting heats up, and less heat flows from the casting: solidification times are longer.

If the mold has a high conductivity (chill-Fig. 9b) q_f will increase relatively, q_s will be smaller, the temperature of the first layer of the mold will build up more slowly and more heat is exhausted from the casting.

It is hoped that this method of presentation, applicable to any transient problem, may be helpful in understanding the thermal problems in foundries.





USE OF PIG IRON IN IRON FOUNDRIES

(WITH PARTICULAR ATTENTION TO SPECIFICATIONS, PRICES, SECONDARY ELEMENTS, AND "HEREDITY")

By

H. W. Lownie, Jr.*

The United States is the world's largest producer of pig iron. Most pig iron is manufactured in blast furnaces from iron ore, coke, and limestone. A large blast furnace will produce as much as 1500 tons of pig iron a day, but the average production per existing furnace is about 950 tons per day. Most pig iron from blast furnaces is used directly from the blast furnace while still molten for the manufacture of steel in open hearths or bessemer converters. Less than 10 per cent of the total production of pig iron is cast into pigs and shipped to steel plants for the manufacture of steel or to foundries for remelting and pouring into castings.

This paper is not concerned with the methods of producing pig iron, except for a few factors that affect the use or behavior of the pig iron in foundries.

Many sources of information have been used, some of which have been published, and some of which have not. It has been necessary at many points to compromise between differing points of view, and in the case of published material, to select among conflicting data and information.

Classification of Pig Iron

There are several informal and two formal methods for classifying pig iron. The foundry industry has, in the past, generally used the informal methods which, unfortunately, differed from area to area.

Formal methods for classifying pig iron have been set up by the American Iron and Steel Institute (AISI) and by the American Society for Testing Materials (ASTM). The first ASTM standards were set up in 1904. The latest revision of the ASTM specifications for foundry pig iron was completed in June, 1955, after a long period of study and compromise between producers and consumers.

As foundries have striven for closer and closer control of their products, the need for, and acceptance of, the formal specifications by the foundry industry has become increasingly important so that these formal specifications have almost completely superseded the outdated informal "trade practice" designations.

Some of the old and obsolete designations for pig iron will be discussed in this paper, because of their historical interest, and because discussion of them brings out the faults which have led to their obsolescence.

Sand-Cast and Machine-Cast Pig Iron

Pig iron derived its name from early casting practices in which the molten iron was run from the blast furnace and allowed to solidify in hollows molded into beds of sand. The pattern in the sand reminded some early ironmaker of suckling pigs attached to a sow, and so the name was born.

Sand-cast pigs were usually heavy (up to 150 pounds) and had sand adhering to the surface. With the advent of more-modern production methods, sand casting was replaced by casting in metal molds on a continuous conveyor.

Machine-cast (also called chill-cast) pigs are cleaner and smaller than the old sand-cast pigs. Machine-cast pigs usually weigh less than 60 pounds (sometimes as little as 10 pounds), and are convenient to handle in foundries. Because of the high cooling rate of machine-cast pigs, they usually contain more carbon than sand-cast pigs, and have a finer structure when fractured. Because of its more desirable characteristics and lower cost of production on a large scale, machine-cast pig has replaced sand-cast pig in American foundry applications.

Blast-Furnace and Electric-Furnace Pig Iron

By far the largest amount of pig iron is produced in blast furnaces. Most of the iron is charged to the blast furnace in the form of iron ore, but sometimes a small percentage of iron or steel scrap is used in the burden for non-merchant grades. Table 1 gives the average U.S. blast-furnace burden for 1955. Some producers of blast-furnace iron contend that the use of an all-ore charge without scrap gives an iron of higher quality. Iron produced without using scrap is often identified as "virgin" iron. Because of operating characteristics which limit the production of a blast furnace, it is usually not considered economical to produce blast-furnace irons containing more than about 14 per cent of silicon. Only a few blast fur-

*Battelle Memorial Institute, Columbus, Ohio.

Table 1 — Average Consumption of Raw Materials in U.S. Blast Furnaces Producing Pig Iron in 1955*

	Tons Per Ton of Pig Iron	Per Cent of Metallic Burden
Ore, briquets, sinter, etc.	1.662	89
Scrap	0.049	3
Mill cinder, scale, etc.	0.149	8
Coke	0.873	—
Limestone	0.387	—

*Source: *Annual Statistical Report*, AISI, 1955.

naces are used to produce pig iron containing over 3.5 per cent silicon.

Electric-arc furnaces are used for the production of pig iron containing more than 14 per cent of silicon. By proper selection of the charge and operation of the furnace, it is possible to prepare "irons" containing up to as much as 95 per cent silicon. These are the so-called ferrosilicons. For usual use in foundries, however, the silicon content of electric-furnace pig iron is usually limited for economic reasons to about 17 per cent. A characteristic of electric-furnace high-silicon irons is that the iron for the charge is usually added as scrap instead of as ore. For this reason, incidental elements from the scrap become of more concern than when an ore charge is used.

In Europe, much pig iron of low and medium silicon contents is produced in electric furnaces directly from the ore. Such irons are of high quality, but the electric-smelting of iron ore has never been adopted in the United States on a commercial scale because of our comparatively high cost of electric energy, and because of our need for very large tonnages which can better be furnished by large blast furnaces. The future, however, may see the introduction of iron smelting in electric furnaces on a commercial scale in the United States.

On occasion, some foundries have melted steel scrap, iron scrap, and ferro-alloys together in a cupola to produce a high-carbon material which they have pigged for later remelting. This, however, is a limited practice which is economical only under highly unusual conditions, and is given no further attention in this paper because the product does not fit the usual definitions for pig iron.

Merchant Pig Iron

Merchant pig iron is that which is produced specifically for sale in pig form. Most of the pig iron bought by foundries is produced in merchant furnaces whose entire output is for sale. Only about 10 per cent of the pig iron made in the United States is merchant iron. The other 90 per cent is made by captive furnaces in steel mills and is used directly for the manufacture of steel, usually while still in the molten state.

Fracture

Up to fifty years ago, pig iron was designated by the appearance of the fracture. The method has fallen into disrepute and is completely inadequate for today's needs. Fracture classification was subject to human error and was always a subject of controversy. It has been superseded by classifying pig iron on the basis of method of manufacture, by traditional end use, and by chemical composition. Of the old fracture designations, only one has persisted. The high-silicon pig irons which show a silvery fracture are still commonly identified as silvery pig irons.* Other pig irons of lower silicon content have a gray, or even black, fracture. The hopelessness of classification by fracture is illustrated by the fact that the present specifications of the American Iron and Steel Institute recognize 236 grades of pig iron, and the American Society for Testing Materials recognizes 243 grades of pig iron for foundries.

Method of Manufacture

Pig iron is often given a general designation on the basis of the main fuel used for smelting, for example charcoal pig and anthracite pig (both of which are extinct in U.S. practice).** If the fuel is not specified, it is assumed that coke was used as the fuel. Some producers (mainly foreign) identify their product on the basis of the type of ore used in the furnace burden, for example, hematite pig (which usually signifies a low phosphorus content), magnetite pig, titanium pig, or chromium-nickel pig. Although all U.S. pig irons from blast furnaces are smelted with a hot blast, some foreign producers identify their product as cold-blast pig. In the field of high-silicon pig irons, a distinction is made between blast-furnace silvery iron and electric-furnace silvery iron.

End Use

The most common designation for pig iron in the United States is based on the main purpose for which the particular type of iron was *originally* used; for example, basic pig iron for making basic-open-hearth steel, bessemer iron for making steel by the acid-bessemer or acid-open hearth processes, malleable pig for making malleable iron castings, and foundry pig for making gray iron castings. Today these designations do not tell the whole story. For example, much malleable and bessemer pig is used today in the production of gray iron castings, and there are other deviations from the historic uses. Table 2 shows the proportions of pig iron used for different purposes in the United States in 1954.

Pig irons are sometimes loosely designated by unusual features of their chemical composition, for example, high-silicon pig, low-phosphorus pig, high-phosphorus pig, and so on. Chemical composition is the basis of the AISI and ASTM formal classifications which will be discussed later.

Trade Practices

Informal trade practices formerly divided pig irons into several broad classes, some of which are listed in

*Modern "silvery" pig iron is graded and marketed on the basis of its chemical composition. The historic designation "silvery" is retained merely because it is just as convenient to use as "high-silicon".

**A small, but significant, amount of coal is added to the coke burden in some modern blast furnaces.

TABLE 2 — END USE OF PIG IRON IN 1954[a]

	Per Cent of Total Production
Open-hearth steel	82.9
Bessemer steel	4.9
Electric-furnace steel	0.3
Cupola	8.3[b]
Air furnaces	0.4
Direct castings	3.2
	100.0

[a] Source: Metal Statistics, 1955, American Metal Market.
[b] Includes some direct castings.

TABLE 3 — EXAMPLES OF DIFFERENCES IN PHOSPHORUS CONTENT WHEN INFORMAL TRADE PRACTICES WERE USED TO SPECIFY PIG IRON

Informal Designation for Pig Iron	Phosphorus Content, per cent	
Low-phosphorus	Agreement at 0.035 max	
Intermediate low-phosphorus	Variously reported: 0.036 to 0.075	
	0.075 max	
Bessemer	Variously reported: 0.076 to 0.10	
	0.10 max	
Malleable	Variously reported: 0.20 max	
	0.101 to 0.300	
Foundry	Many local differences, see Table 4	
Basic	Local differences about as follows:	
	Southern furnaces	0.90 max
	Northern furnaces	0.40 max
	Some Valley furnaces	0.30 max

TABLE 4 — EXAMPLES OF DIFFERENCES IN ANALYSES OF FOUNDRY PIG IRON AS SHOWN BY A COMPILATION OF TRADE PRACTICE DESIGNATIONS

Producing Area	Composition, per cent	
	P	Mn
Central West, Valley, Chicago, Buffalo, Central lake ports; "Standard"	0.40 to 0.80	0.50 to 1.00
Same districts; "extra strong"	0.30 max	0.75 to 1.25
Eastern	0.30 to 0.50	0.50 to 0.90
	0.50 to 0.70	
	0.50 to 0.90	
Southern; "standard"	0.70 to 0.90	0.20 to 1.00
	0.70 to 1.00	
Southern; "low phosphorus"	0.20 to 0.70	0.20 to 1.00
	0.31 to 0.70	
Southern; "high manganese"	0.70 to 1.00	0.50 to 1.00

TABLE 5 — OBSOLETE TRADE-PRACTICE DESIGNATIONS FORMERLY USED TO SPECIFY THE SILICON CONTENT OF FOUNDRY PIG IRON

Silicon Content, per cent	Designation Used in Local Markets		
	Midwest	East	South
Under 1.75	No. 3	No. 3	No. 3
1.75 to 2.00	—	—	No. 2 Hard
1.75 to 2.25	No. 2	No. 2 Plain	No. 2 Foundry
2.25 to 2.75	No. 1	No. 2X	No. 2 Soft
2.75 to 3.25	No. 1X	No. 1X	No. 1 Foundry

Table 3. The main difference between the classes was in the phosphorus content of the iron.

The major disadvantage of the old trade-practice designations was that the name for a grade of iron meant different things in different parts of the country. This is shown to some degree in the listing of the phosphorus content of basic pig in Table 3. There were so many local differences in the phosphorus and manganese contents of foundry pig iron, that Table 4 can be little more than an attempt to list some of them. Similarly, Table 5 is an attempt to list the meanings of the old trade-practice designations applied to the silicon content of foundry pig iron.

When it is realized that, in addition to the differences suggested in Tables 3, 4, and 5, there were other variations peculiar to specific producers and markets, it is easy to see why these confusing trade-practice designations have become obsolete and have been replaced by formal classifications which mean the same thing in all parts of the country.

AISI and ASTM Classifications

The most useful and definitive classifications for pig iron are those of the American Iron and Steel Institute and the American Society for Testing Materials. The AISI lists ten standard classes and the ASTM lists eight. AISI includes classes not normally used in foundries, and ASTM lists more grades for the classes for pig iron usually used in foundries. Both systems use a system of letters and numerals to identify each grade of iron. Some of the letter symbols in the two systems are the same, but the symbols do not necessarily mean the same thing in both systems.

The AISI system uses six or seven numerals, the first three or four to designate the average silicon content of the iron, and the last three to designate the average manganese content. ASTM simply numbers the grades and never uses more than two numerals. For example, AISI Grade LP 113088 is the same as ASTM Grade LP 11. Both designations indicate that the iron contains a maximum of 0.035 per cent phosphorus, a maximum of 0.035 per cent sulfur, between 1.00 and 1.25 per cent silicon, and between 0.76 and 1.00 per cent manganese.

AISI and ASTM classifications differentiate between classes of iron mainly on the basis of phosphorus content, because phosphorus is the most difficult element to lower in either steelmaking or in remelting for foundry uses. Within each class, grades differ in silicon and manganese content. For each class except silvery pig, grades usually differ in steps of 0.25 per cent silicon and 0.25 per cent manganese.

AISI and ASTM classifications are summarized in Tables 6 and 7. The detailed lists can be obtained at small cost from AISI* as Volume 44 of the AISI "Contributions to the Metallurgy of Steel," and from ASTM** as Specifications A43, Foundry Pig Iron. All producers of pig iron are familiar with both systems.

It is important to recognize that there are significant differences between grades in the AISI and ASTM coding systems which might at first glance

*American Iron and Steel Institute, 350 Fifth Avenue, New York, N. Y.

**American Society for Testing Materials, 1916 Race St., Philadelphia 3, Pa.

Table 6 — Summary of AISI Designations for Pig Iron*

Symbol	Class	No. of Grades	Composition, per cent P	S	Mn	Si
LP	Low phosphorus	26	0.035 max	0.035 max	To 1.25	0.50 to 3.00
LPi	Intermediate low-phosphorus	24	0.036 to 0.075	0.05 max	To 1.25	1.00 to 3.00
Bes	Bessemer	16	0.076 to 0.100	0.05 max	To 1.25	1.00 to 3.00
M	Malleable	33	0.101 to 0.300	0.05 max	0.50 to 1.25	0.75 to 3.50
S	Silvery	72	0.300 max	0.05 max	To 2.00	5.00 to 17.00
B	Basic-Northern	6	0.400 max	0.05 max	1.01 to 2.00	To 1.50
Bs	Basic-Southern	3	0.700 to 0.900	0.05 max	0.40 to 0.75	To 1.50
Fl	Foundry, Northern low-phosphorus	24	0.300 to 0.500	0.05 max	0.50 to 1.25	To 3.50
Fh	Foundry, Northern high-phosphorus	24	0.501 to 0.700	0.05 max	0.50 to 1.25	To 3.50
Fs	Foundry, Southern	8	0.700 to 0.900	0.05 max	0.40 to 0.75	To 3.50

*From AISI Steel Products Manual, Section 1, May, 1951.

Table 7 — Summary of ASTM Specifications for Foundry Pig Iron, ASTM Designation A43-55T*

Symbol	Class	No. of Grades	Composition, per cent P	S	Mn	Si
LP	Low-phosphorus	26	0.035 max	0.035 max	To 1.25	0.50 to 3.00
LPi	Intermediate low-phosphorus	24	0.036 to 0.075	0.05 max	To 1.25	1.00 to 3.00
Bes	Bessemer	16	0.076 to 0.100	0.05 max	To 1.25	1.00 to 3.00
M	Malleable	33	0.101 to 0.30	0.05 max	0.50 to 1.25	0.75 to 3.50
S	Silvery	72	0.30 max	0.05 max	0.50 to 2.00	5.00 to 17.00
Fl	Foundry low-phosphorus	24	0.31 to 0.50	0.05 max	0.50 to 1.25	To 3.50
Fm	Foundry intermediate phosphorus	24	0.51 to 0.70	0.05 max	0.50 to 1.25	To 3.50
Fh	Foundry high-phosphorus	24	0.71 to 0.90	0.05 max	0.25 to 1.00	To 3.50

*Latest revision as accepted June, 1955.

appear to be the same. For example, Grade Fh, Foundry Northern high-phosphorus pig iron in the AISI system calls for 0.501 to 0.700 per cent phosphorus. Grade Fh, Foundry high-phosphorus pig iron, in the ASTM system calls for 0.71 to 0.90 per cent phosphorus. For this reason, it is not a good practice, even when using these formal systems, to identify a pig iron by anything less than its full identification consisting of both symbol and number.

In addition to dividing the possible compositions for pig iron into a number of specific standard ranges of composition for the convenience and guidance of both producers and consumers, the AISI and ASTM specifications perform further useful services by providing information on the method of manufacture, sampling, and inspection of the iron, and provide for manufacture of special grades for special purposes.

Carbon in Pig Iron

The carbon content of pig iron usually is not specified. Fortunately, however, there is a fair relationship between the silicon content of machine-cast pig iron and its carbon content. Such a relationship for a wide range of silicon contents is shown in Fig. 1. The relationship is practically independent of variations in manganese content between 1 and 3 per cent, but is affected by the phosphorus content of the iron. For a given silicon content, irons of lower phosphorus content will tend to be slightly higher in carbon content.

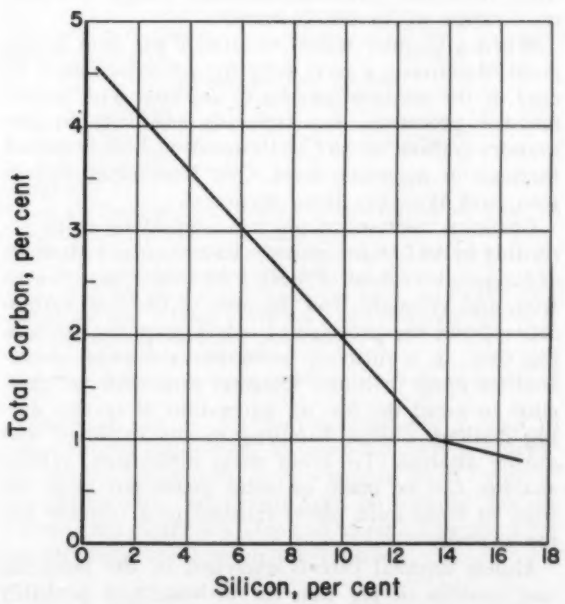


Fig. 1 — Approximate relationship between carbon and silicon contents in chill-cast pig iron. (Source: "Carbon in Pig Iron," W. H. White, Foundry, June, 1946.)

Figure 1 is for chill-cast or machine-cast pig. For sand-cast pig (which is not produced in the United States), the carbon content can be expected to be a

little lower than shown in the curves, because slower cooling in the sand permits more graphite to separate from the iron as kish.

Because molten pig iron is saturated with carbon which is rejected from the iron during cooling, the techniques of ladle practice, pouring practice, pouring temperature, and pigging practice at the blast furnace can have an appreciable effect upon the final carbon content of individual pigs. Because of the appreciable length of time that it takes to pig a large cast of iron from the blast furnace, some variation in the carbon content of individual pigs is unavoidable, but it is one duty of the producer to keep such variations at a minimum.

Sampling and Analysis of Pig Iron

Producers usually sample merchant pig iron from the ladle used for each cast. The pigged iron may be loaded directly into railroad cars for shipment, or may be diverted to storage where it is stockpiled with pig iron from other casts of similar composition. If the iron is loaded directly from the pig machine, the vendor's analysis furnished to the customer with each carload of iron is the analysis of the particular cast. If the iron is shipped from a stockpile, the vendor's analysis furnished with each carload of pig iron is the weighted average analysis of the stockpile. When the pig iron is shipped without stockpiling, the vendor's analysis is representative of the composition of the iron in the shipment, but when the iron is shipped from a stockpile, the vendor's analysis may differ slightly from the average analysis of the pig iron in the shipment. In either case, however, the iron from reputable producers will be within the contract range set by the customer.

When a foundry wishes to analyze pig iron at the point of delivery, a good sampling procedure must be used or the analysis may be misleading. The recommended procedure for sampling pig iron in the foundry is given in ASTM Designation E59, Standard Methods of Sampling Steel, Cast Iron, Open-Hearth Iron, and Wrought Iron; Section 3.

Chemical analyses of pig iron should be made according to ASTM Designation E30, Standard Methods of Chemical Analysis of Steel, Cast Iron, Open-Hearth Iron, and Wrought Iron. Because of the high carbon content and the presence of coarse graphite flakes in pig iron, it is difficult to obtain a correct carbon analysis using drillings. Unusual care must be exercised to avoid the loss or segregation of carbon during drilling, during handling of the drillings, and during analysis. To avoid these difficulties, carbon analyses can be made on solid pieces cut from the iron, or from solid pieces crushed to a suitable size for analysis.

Unless unusual care is exercised in the sampling and analysis of pig iron for carbon, it is probably preferable to estimate the carbon content from a curve such as Fig. 1.

Foundrymen recognize that different analysts cannot be expected to agree precisely on the composition of pig iron or cast iron, even when the same sample is used. AISI has prepared a guide to "reasonable differences" to be expected between different analysts. This guide is shown in Table 8. It should be noted that the guide is based on the use of a "properly divided" set of drillings from the same sample.

TABLE 8 — REASONABLE DIFFERENCES BETWEEN ANALYSTS WORKING ON PROPERLY DIVIDED SET OF DRILLINGS FROM THE SAME SAMPLE OF PIG IRON*

Element, per cent		Reasonable Difference, per cent
Manganese		
	to 0.80	0.03
	over 0.80 to 1.35	0.04
	over 1.35 to 1.75	0.05
Phosphorus		
	to 0.040	0.005
	over 0.040 to 0.075	0.006
	over 0.075 to 0.300	0.010
	over 0.300 to 0.400	0.015
	over 0.400 to 0.600	0.020
	over 0.600 to 0.900	0.025
Sulfur		
	to 0.05	0.005
Silicon		
	0.50 to 1.50	0.06
	over 1.50 to 2.50	0.07
	over 2.50 to 3.50	0.08
	over 3.50 to 5.00	0.10
	over 5.00	0.25

*Source: AISI Steel Products Manual, Section 1, 1951.

Principal Uses for Different Grades of Pig Iron

The principal modern uses for the most important grades of pig iron are summarized below.

Basic Pig Iron

Basic pig iron is the largest class of iron produced in blast furnaces. Table 9 shows that 80 per cent of the pig iron produced in the United States in 1955 was basic iron. Little of this iron is used in foundries because its particular combination of silicon, manganese, and phosphorus contents usually makes

TABLE 9 — PRODUCTION OF PIG IRON IN THE UNITED STATES IN 1955*

	Production, net tons	Shipped for Other Than Own Use, net tons
Basic	62,484,889[a]	1,076,229[c]
Bessemer and low phosphorus	7,699,390	788,406
Malleable	3,531,420	3,661,403[d]
Foundry	2,754,641	2,027,185[d]
Silvery pig and ferro-silicon	803,281	375,000*
All other pig iron	387,077	200,000*
	77,660,698[b]	8,128,223

[a] Basic pig iron accounted for 80 per cent of total production.
[b] Production of pig iron in 1955 was at 92.6 per cent of capacity. On January 1, 1955, there were 198 furnaces in blast, 52 idle.
[c] Only 1.7 per cent of basic pig was shipped for other than own use.
[d] Foundry and malleable pig iron accounted for 70 per cent of all pig shipped for other than own use.
* Estimated.
*Source: Annual Statistical Report, 1955, AISI.

it undesirable for most foundry applications. Most basic iron is used while still molten directly from the blast furnace into a basic-open-hearth furnace for steelmaking. Phosphorus can be removed in the basic-open-hearth process. Basic iron produced from Southern ores is higher in phosphorus and lower in manganese than basic iron produced from Northern ores. (See Table 6.) Less than 2 per cent of the basic pig iron is pigged and shipped as merchant iron.

Bessemer and Low-Phosphorus Pig Irons

These irons, which are low in phosphorus content, are used in the manufacture of acid-open-hearth steel and acid-bessemer steel because the acid steelmaking processes do not permit phosphorus to be removed. Low-phosphorus pig iron, intermediate low-phosphorus iron, and bessemer pig iron are used in gray iron foundries and steel foundries for the manufacture of rolls, heavy-duty brake drums, glass molds, cylinder liners, large engine castings, automotive dies, chilled iron castings, nodular iron castings, and other castings where it is desirable to maintain phosphorus at a low level.

Bessemer pig iron contains a little more phosphorus than the low-phosphorus classes, but its phosphorus content is still low enough to permit its use in the acid bessemer process. Most bessemer pig is used as hot metal as a charge for bessemer converters. Comparatively little bessemer pig is marketed, but its use in foundries is increasing. The main merchant use has traditionally been for making cast iron ingot molds and rolls.

Because phosphorus is not lowered during smelting in a blast furnace, the production of any of these low-phosphorus grades presents a major problem in selection of the ore, coke, and limestone. The quantity of these pig irons which can be produced is limited by the amount of suitable raw materials available. For this reason, the low-phosphorus grades of pig iron are the most expensive, as shown in Table 10.

Malleable Pig Iron

Malleable pig iron is widely used for the production of malleable iron and gray iron castings of all types where the phosphorus content may be higher than that of the expensive low-phosphorus grades, but where phosphorus must be lower than that of foundry pig iron. Because of its convenient level of phosphorus and wide ranges of silicon content, the largest tonnage of pig iron shipped from producers and merchant furnaces is in this class, as shown in Table 9. Malleable pig and foundry pig together account for about 70 per cent of the merchant pig produced in the United States.

Table 10 — Examples of Effect of Phosphorus Content on Price of Pig Iron as of March 15, 1956*

Base Grade	Producing Area	Phosphorus, per cent	Price fob Producing Point, $/GT for base grade[a]
Basic	Southern	0.70 to 0.90	54.50
Foundry	Southern	0.70 to 0.90	55.00
Foundry	Southern	0.31 to 0.69	56.00
Foundry	Northern	0.40 to 0.80	59.00
Malleable	Northern	0.10 to 0.30	59.00
Bessemer	Northern	0.076 to 0.10	59.50
Intermediate low phosphorus	Northern	0.036 to 0.075	63.50
Low phosphorus	Northern	0.035 max	66.50
Low phosphorus	Southern	0.035 max	72.50

[a] Does not include price increases of about $1.50 per gross ton expected in late March or early April.
*Source: *Steel*, March 26, 1956.

Foundry Pig Iron

These irons give a wide selection of fairly high phosphorus contents, and a wide range of silicon contents for making a large variety of castings where phosphorus content is not critical. Foundry pig iron is usually remelted in cupolas with cast iron scrap and steel to produce castings for machine tools, dies, rolls, and all kinds of heavy and light castings.

Silvery Pig Iron

Common grades of silvery pig iron contain 5 to 17 per cent of silicon. It is used for adding silicon to iron and steel. Because of the high cost of producing pig iron high in silicon content, silvery pig iron must be sold at a higher price than pig irons of lower silicon content. Some typical prices are shown in Table 11.

Table 11 — Typical Grades of Silvery Pig Iron as of March 15, 1956*

Producing Point	Price fob Producing Point, $ per GT[a]
Blast-Furnace Silvery (6.0 to 6.5% Si)	
Jackson County, Ohio	67.50[b]
Buffalo, New York	68.75[b]
Electric-Furnace Silvery (14.01 to 14.5% Si)	
Niagara Falls, New York	91.00[c]
Keokuk, Iowa	95.50[c d]
Keokuk, Iowa (12½-pound piglets)	98.50[c d]

[a] Does not include price increases of about $1.50 per gross ton expected in late March or early April.
[b] Add $1.25 per ton for each additional 0.50% Si and $0.75 per ton for each additional 0.50% Mn over 1%.
[c] Add $1.00 per ton for each additional 0.50% up to 18%. Add $1.00 per ton for each additional 0.50% Mn over 1%.
[d] Freight allowed to normal trade area.
*Source: *Steel*, March 26, 1956.

Although silvery pig iron is available in pigs of conventional size (about 50 pounds), there is a growing tendency to market silvery iron in small "piglets" of closely controlled weight (say, 10 pounds) for the convenience of the foundry and to eliminate the need for weighing the pigs in the foundry.

Charcoal Pig Iron

Charcoal was widely used as a blast-furnace fuel in the United States until about 1840. By 1870, only about 20 per cent of U.S. pig iron was made with charcoal. The last charcoal furnace in the U.S. was taken off blast in 1943 or 1944. Charcoal iron was characterized by low sulfur and low silicon contents.

Similar compositions of iron are obtainable today using coke as a fuel.

Because no charcoal pig iron is now produced in the United States, neither the AISI nor ASTM recognize it is as a classification for pig iron. By special arrangements, however, charcoal pig iron can be imported.

Pig Iron for Special Uses or With Special Characteristics

Some producers go to special efforts to produce pig iron especially suited to use in the manufacture of high-ductility nodular iron. These are low-phosphorus pig irons with low contents of manganese, sulfur, and trace elements, and often carry special trade names. ASTM is considering preparation of specifications to cover such pig irons specifically, but as yet there are no formal industry-wide specifications on such pig irons. The range of chemical composition of such irons is about as follows:

Composition, per cent					
P	S	Mn	Si	Ti	Other Secondary Elements
0.035 max	0.035 max	0.15 max 0.25 max 0.30 max	Ranges from about 0.75 to 2.00	0.07 max 0.10 max	May be specified by special arrangement

Most merchant pig iron is produced from general shipping ores which lose their identity by the time they are blended and delivered to the blast furnace. Some pig iron is produced mostly from more-or-less local ores which may impart some special characteristics to the iron. For example, one low-phosphorus pig iron has been produced for many years mostly from New York magnetite ore, and, because of the titanium content of this ore, contains up to 0.25 per cent titanium in some grades. Another low-phosphorus pig iron is produced in the South from the residue of sintered pyrrhotite (iron sulfide), rather than from conventional oxidic ores.

Another merchant pig iron is unusual in that titanium-bearing ore is purposely added to the blast furnace to add about 0.40 to 0.50 per cent of titanium to the pig iron.

The American market used to include one brand of pig iron produced from Cuban ores. This pig iron contained up to about 1.25 per cent nickel, up to about 2.30 per cent chromium, and small amounts of vanadium and titanium. Although this pig iron was popular at one time for adding nickel and chromium to cupola charges, it has not been manufactured since 1941, and resumption of production is not expected.

Price of Pig Iron

The price of pig iron is an important item in the cost of operating an iron foundry. To a large extent, the cost of pig iron to the consuming foundry is determined by the manufacturer's cost of assembling the raw materials and producing the desired grade of iron at a reasonable profit, and freight costs between the manufacturing plant and the foundry.

Pig iron is sold in the United States on the basis of the gross ton or long ton, both of which are 2240 pounds.

Effect of Common Elements on Price

The common elements that affect the price of pig iron are phosphorus, silicon, and manganese. Phosphorus is so important that it is used as the basis for the classification of pig iron. The effect of phosphorus content on market price was shown in Table 10. Because of the lower cost of higher phosphorus irons, foundries should use them whenever metallurgically possible.

Silicon and manganese in pig iron both cost money. At present, the price of blast-furnace silvery pig is increased $1.25 per ton for each additional 0.50 per cent silicon over the base grade, and other standard irons are increased $0.50 per ton for each additional 0.25 per cent silicon over the base grade. Base prices for non-silvery grades are increased $0.50 per ton for each 0.50 per cent of manganese over 1.00 per cent.

Effect of Freight Rates on Price

When a foundry determines the minimum cost of pig iron for their charges, freight rates must be taken into consideration. Pig iron from the nearest producer establishes the price at the point of delivery, but the nearest producer may not market the particular grade desired. Table 12 gives some examples of railroad freight rates. Railroad freight rates are determined from rather complicated schedules, but a rough approximation of freight costs between usual points of shipment is about as follows:

Distance, miles	Approximate Carload Rate, dollars per gross ton
1	1.60
50	2.70
100	4.80
200	7.20
400	8.50
600	9.30

Relative Prices of Pig Iron and Other Raw Materials

Foundries are usually interested in the price of four main raw materials for melting; pig iron, cast iron scrap, steel scrap, and coke.

Market prices for the melting stock for foundries

TABLE 12 – TYPICAL RAILROAD FREIGHT RATES FOR PIG IRON*

From	To	Carload Rates, dollars per gross ton
Milwaukee, Wis.	Chicago, Ill.	2.30
Canton, Ohio	Cleveland, Ohio	2.91
Newark, N. J.	Steelton, Pa.	4.64
Cincinnati, Ohio	Toledo, Ohio	6.11
Cincinnati, Ohio	Cleveland, Ohio	6.68
Cincinnati, Ohio	Birmingham, Ala.	8.16
San Francisco, Calif.	Geneva, Utah	9.39
St. Louis, Mo.	Birmingham, Ala.	9.51

*Source: American Metal Market, March 20, 1956.

are affected by many factors, many of them on the local level. These factors include general business conditions, competition of demand by the steel industry, and changes in local market conditions.

By paying attention to the changing local delivered price of each raw material, furnace charges can often be cheapened. For example, the amounts of pig iron, steel scrap, and coke in a cupola charge are related. When pig iron is expensive in relation to coke and steel, the cheapest charge will usually be low in pig, high in steel, and will use a rich coke ratio. When pig iron is cheap in comparison with steel and coke, the most economical charge will often be high in pig, low in steel, and use a lean coke ratio. There are mathematical methods for calculating the charge of lowest cost under any given set of market conditions and average charge composition desired, but these methods require solution by proficient mathematicians.

Pig iron and foundry coke are both at their highest prices in 1956, but there has been fair consistency in the relation between the prices of these materials since 1930. For example, although the cost of foundry pig iron at Chicago has increased from $15.50 per gross ton in 1933 to $60.50 in 1956, and the price of foundry coke at Chicago has increased from $7.00 per ton in 1933 to $27.00 in 1956, the price of pig iron at Chicago each year has fluctuated around an average close to 2.2 times the cost of foundry coke. The extremes were in 1944 and 1945 when pig iron cost only 1.91 times as much as coke, and in 1930 and 1951 when pig iron cost 2.50 times as much as foundry coke. The changing prices of pig iron and foundry coke at Chicago are shown in Fig. 2. The

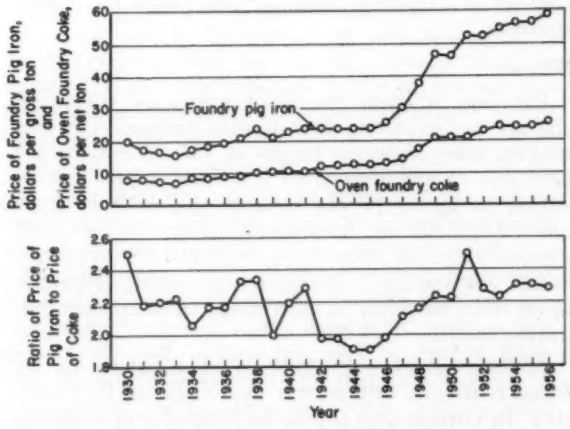


Fig. 2 — Prices of foundry pig iron and oven foundry coke at Chicago in January of years shown.

ratio chart given in the lower part of Fig. 2 is a good way of showing when coke is cheap with respect to pig iron, and vice versa. The chart must be used, however, with consideration to the fact that a gray iron foundry may use only one ton of coke to each 2 to 8 tons of purchased pig iron.

The relationship between the prices of foundry pig iron, cast iron scrap, and steel scrap has been much more erratic than the relationship between pig iron and coke. Therefore, with due regard to metallurgical considerations, it is in adjustment of pig iron, cast iron scrap, and steel scrap in charges that the foundryman has a great opportunity to reduce his costs.

Foundry pig iron often costs about 50 per cent more than good cast iron scrap. This was the situation in 1954. In 1948, however, foundry pig at Chicago was selling for almost half the price of cast scrap. This was obviously the time for high-pig charges (if you could get pig iron!). At the other extreme, in 1933 pig iron at Chicago cost over twice as much as good cast scrap and low-pig charges were indicated so long as the metallurgical quality of the resulting iron could be maintained. Major variations like this suggest obvious changes in metal charges at the time, but foundrymen often overlook the fact that local changes in prices are constantly occurring, and that continual cooperation between the purchasing agent and the metallurgist can sometimes save the foundry real money.

In 1954, steel scrap was selling at a comparatively low price, much lower than in 1951, 1952, or 1953. It is obvious that this was a time for high steel charges, where they could be used. Since 1954, however, the price of steel scrap has increased steadily nearer to the price of pig iron, because of increasing exports of steel scrap and increased demand from steelmakers.

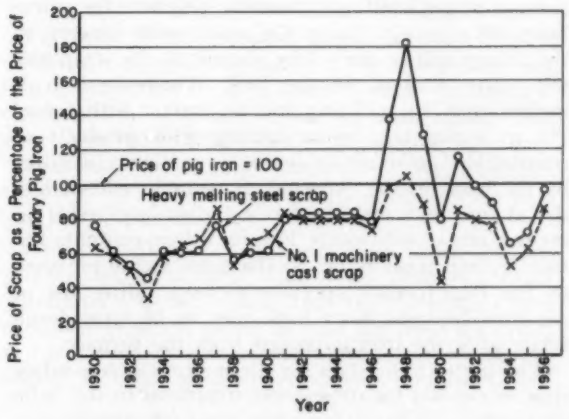


Fig. 3 — Relative prices of foundry pig iron, cast iron scrap, and steel scrap at Chicago in January of years shown.

Figure 3 shows the relation between the prices of foundry pig iron, cast scrap, and steel scrap at Chicago since 1930. It is obvious that there were drastic swings in the relationship between these prices. The two types of scrap do not even bear a constant relationship to each other. As an example of fast local changes, in 1948 steel scrap sold at Chicago for more than pig iron, and two years later steel scrap at the same location sold for less than half the price of pig iron.

An intelligent metallurgist with the ability and authority to adjust his cupola charges can save a foundry appreciable money by close observation of price changes in his local market. To be of greatest value, however, constant attention must be given to

this (rather than yearly as in Fig. 3), and the method should be based on delivered costs at the foundry (rather than on market prices as in Fig. 3).

Primary Elements in Pig Iron and Their Behavior During Remelting

In addition to iron, the primary elements in pig iron are the so-called "big five"; carbon, silicon, manganese, sulfur, and phosphorus. These are the elements which receive most of the foundryman's attention when he selects his raw materials for melting. The amount of each element desired in the charge will be determined by the nature of the castings to be made (gray iron, malleable iron, etc.), their section thickness, their desired mechanical properties, and the nature of the melting unit to be used. The amount of pig iron which the foundryman will decide to blend with scrap to make up his charge depends upon the quality, chemical composition, and cost of the scrap available to him. The selection of actual charges and blends of scrap and pig iron to satisfy the numerous variations of conditions desired by foundrymen is outside the scope of this paper. A few comments will be made, however, to point out some of the major factors that must be considered.

Sulfur

Sulfur is not much of a problem in pig iron because it almost invariably contains less than 0.05 per cent sulfur. Only under unusual conditions will the sulfur content of pig iron be a deterring factor in the calculation of a cupola charge. Of much more concern in the charge will be the sulfur content of the scrap and the sulfur content of the fuel. Whenever iron is melted with an acid slag and in contact with a fuel (as in a cupola), or in contact with products of combustion (as in an air furnace), a pickup of sulfur by the iron can be expected. If suitable selection of the charge and melting method does not result in melted metal sufficiently low in sulfur content, the foundryman must resort to the same technique used by the blast-furnace operator to keep sulfur low in pig iron—he must use a basic slag, or he must desulfurize after the iron is tapped from the furnace.

The use of basic slags for the melting of low-sulfur iron in cupolas for subsequent treatment in the ladle to produce nodular iron has received considerable attention within the last several years. Also, within the last several years, methods for desulfurizing iron in the ladle with injections of calcium carbide have been worked out both mechanically and chemically so that the method is practical and rapid. The desulfurization of iron in the ladle with sodium carbonate (soda ash) has been practiced in foundries for many years. Desulfurizing treatments for cast iron are discussed and explained in Chapter 10 of the AFS publication THE CUPOLA AND ITS OPERATION, 1954.

Phosphorus

The phosphorus content of pig iron can be controlled only by selection of the raw materials used in the blast furnace. Because of the scarcity of low-phosphorus raw materials, low-phosphorus pig irons command a premium price. When selecting a charge for a remelting operation, for reasons of economics the phosphorus content of the charge should be as high as possible consistent with metallurgical requirements. When considering the phosphorus content of the charge, foundrymen realize that most melting operations do not permit a reduction in the phosphorus content of the charge. In both acid and basic cupolas, the phosphorus content of the metal at the spout is about the same or slightly higher than that of the charge. The manufacture of gray iron castings permits much more leeway in the use of high-phosphorus charges than can be tolerated in the manufacture of malleable iron, nodular iron, or steel castings.

Manganese

Pig iron is a good source of manganese for ferrous castings. Manganese is furnished economically in the usual ranges encountered in pig iron. When a foundryman is using a charge that requires him to add ferromanganese or manganese bearing briquets to the charge, he should take a second look at the manganese content of his pig iron. In many cases it will be found that a higher manganese content can be furnished in the pig iron at no additional cost to him, or at a cost lower than that of other types of manganese additions. The usual metallurgical function of manganese, of course, is to tie up the sulfur as manganese sulfide so as to avoid the undesirable effects of sulfur. As a general rule in malleable iron, the manganese content of the iron is at least two and one-half times the sulfur content. In malleable iron, the ratio of manganese to sulfur must be controlled quite closely because of its effects on the quality and annealability of the iron. In gray iron the manganese content is frequently four to five times the sulfur content.

Silicon

Pig iron is often the main source of silicon in charges for remelting into ferrous castings. For this reason, many different ranges of silicon content are available in pig iron from less than 1 per cent of silicon to about 17 per cent of silicon. For larger amounts of silicon; ferrosilicon, silicon carbide, and various types of ferrosilicon briquets are available. When calculating a charge for remelting, consideration must be given to the change in silicon content during melting.

Melting under acid slags, as in a cupola or air furnace, usually results in a loss of silicon by oxidation. In normal acid cupola melting of gray iron, the loss amounts to about 10 per cent of the silicon charged. With tapping temperatures of about 2850 F and higher, however, often enough silicon is reduced from the acid slag in a cupola to cause a slight increase in average silicon content during remelting. This, in fact, is part of the technique used to produce high-silicon pig iron in blast furnaces. When melting iron under basic slags, silicon losses become much higher, even as high as 40 per cent under some conditions. This high loss of an expensive element, silicon, is sometimes one of the major disadvantages of basic melting in a cupola.

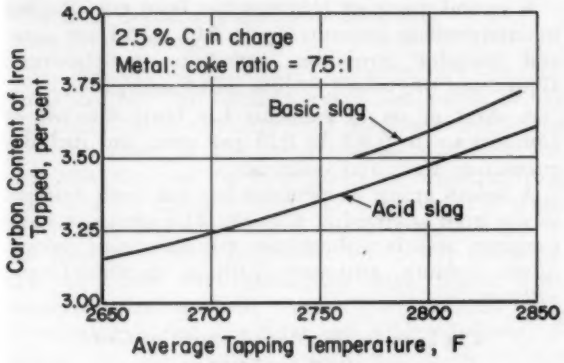


Fig. 4 — Effect of tapping temperature on carbon pickup during melting in a cupola.

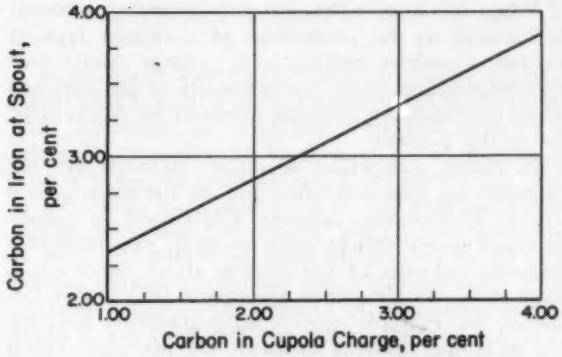


Fig. 5 — Effect of carbon content of cupola charge on carbon content of iron for specific conditions described by Levi.

Carbon

The carbon content of pig iron is related to its silicon content, as was shown in Fig. 1. When pig iron is blended with scrap and remelted in a cupola, the same type of relationship holds, but to a lesser degree. The higher the silicon content of a cupola charge, the lower will be the amount of carbon picked up during remelting. Most cupola charges consisting of a mixture of pig iron and scrap with an average carbon content of less than about 3.40 per cent will increase in carbon content during melting. This effect, however, is small when compared with the effect of tapping temperature upon the carbon content of cupola-melted iron. A typical effect is shown in Fig. 4*, The higher the temperature to which the molten iron is exposed in a cupola, the higher will be its final carbon content. Time of contact between the molten iron and the coke also plays an important part in the amount of carbon picked up during melting. For this reason, cupolas that are tapped continuously produce iron with a lower carbon content than those tapped intermittently at the same temperature.

The nature of the fuel and the basicity of the slag each have an important effect on the carbon content of remelted iron. Some fuels, such as petroleum coke and electrode carbon, are more soluble in iron than others. Basic slags apparently flush away the acid ash from coke and permit the coke to come into better contact with the iron; thus, irons melted in a cupola with a basic slag pick up carbon faster than irons melted with an acid slag (see Fig. 4).

In general, for cupola melting when a low carbon content is desired in the iron at the spout, (1) use a low percentage of pig iron in the charge, (2) use as much steel as possible, (3) melt fast at lean coke ratios, (4) melt on the cold side, (5) keep the silicon and sulfur contents as high as possible, (6) get the iron out of the cupola as fast as possible, (7) avoid fuels or additives known to promote carbon pickup, and (8) keep the slag acid. For high carbon pickup, the opposite conditions should be used. When melting in air furnaces, the same factors apply, but to a much more limited degree because the iron is not in contact with coke during and after melting.

An excellent picture of the effects of some melting conditions on the change in carbon content of iron during melting in a cupola was given by Levi* and is summarized in Fig. 5. From this figure, it can be seen that iron does not always pick up carbon during melting in a cupola. When the charge contains an average of more than about 3.70 per cent carbon, it will often lose carbon during remelting.

Secondary Elements in Pig Iron

Pig iron always contains a number of chemical elements that are often not considered when the pig iron is purchased or used. Some of these "secondary" elements are unavoidable in pig iron made from some ores, and, hence, are often called "residual," "incidental", or "tramp" elements. Other secondary elements are sometimes purposely added to pig iron for specific purposes.

Some secondary elements are those which are frequently used as alloying additions. Examples include chromium, copper, nickel, and molybdenum. Other secondary elements found in some pig irons are rarely or never used as intentional alloying additions. Examples include antimony, tin, lead, cadmium, and cobalt.

The influence of secondary elements in pig iron upon the properties and quality of the castings depends upon the type of castings being made. For example, secondary elements which are harmless in gray iron castings can have an important effect on malleable iron castings or nodular iron castings.

The nature and amount of secondary elements in pig iron are subject to some control by the producer of pig iron. He can, for example, change the type of ore being used so as to avoid harmful amounts of specific secondary elements. Such a change may, however, increase the cost of production of the pig iron, and be reflected in higher prices to the foundry. This situation is illustrated in the price for low-phosphorus pig iron.

*D. E. Krause and H. W. Lownie, Jr., "Blast Humidity as a Factor in Cupola Operation," Transactions, AFS, vol. 57, pp. 111-132 (1949).

*W. W. Levi, "Variables Affecting Carbon Content in Cupola Operation," Transactions, AFS, vol. 55, p. 626 (1947).

When it is known that a specific secondary element is harmful to the production of a certain type of casting, foundries making such castings should pass this information on to the producers of pig iron and make arrangements with the producer for limits to be set on this element.

In many cases where secondary elements are important, pig iron will often not be the main source of the troublesome element. Unstripped or poorly stripped motor blocks used as scrap can introduce harmful amounts of non-ferrous alloys. Most scrap, even steel, contains some amounts of undesirable elements. For example, steel scrap is gradually building up in copper content to about 0.1 per cent. When this amount of copper is harmful, steps must be taken to reduce the amount of steel scrap, or to neutralize the effects of the copper.

An insidious thing about many of the most troublesome secondary elements is that they cannot be economically removed from the iron by the foundry—they can only be avoided or neutralized.

Secondary elements in pig iron are best detected by spectrographic analysis. These analyses are inexpensive and indicate the approximate amount of many elements from a single analysis. The results of spectrographic analyses of some typical pig irons that have been analyzed at Battelle Memorial Institute are given in Table 13.

It is interesting to observe the amounts of various secondary elements which have been encountered in pig irons analyzed at Battelle. Many of these irons were analyzed because the presence of contaminating elements was suspected. Obviously, however, the irons made available to this one laboratory do not indicate the maximum amounts of secondary elements that might be encountered in every pig iron. For example, pig irons analyzed spectrographically at Battelle have not contained more than 0.07 per cent chromium, but it is known that some foreign pig irons contain more than 0.2 per cent of chromium.

Of the irons analyzed at Battelle, the secondary elements encountered in the largest amounts were nickel, copper, and titanium, each of which has been found in amounts greater than 0.1 per cent.

A second group of elements has been encountered in intermediate amounts up to 0.01 to 0.1 per cent, and includes chromium, vanadium, molybdenum, aluminum, tin, cobalt, boron, and magnesium.

A third group of elements has been detected in amounts up to 0.001 to 0.01 per cent, and includes zirconium, lead, and cadmium.

A fourth group of elements has not been detected in pig irons analyzed at Battelle. This group includes tungsten, arsenic, columbium, tantalum, zinc, germanium, bismuth, antimony, gallium, beryllium, and calcium.

Pig Iron in the Manufacture of Gray Iron Castings

Most gray iron is melted in cupolas using coke as a fuel to melt a metal charge of pig iron, cast iron scrap, and steel scrap. The amount of coke is usually between 7 and 15 per cent of the weight of the metal charge. Typical charges for gray cast iron contain from 10 to 40 per cent of pig iron, and up to 40 per cent of steel scrap. About 35 per cent of the metal charge in many foundries consists of foundry returns such as scrapped castings, gates, and risers.

Cupola charges are compounded on the basis of the amount of the major elements in each component, the cost of each component, the composition of the iron desired after remelting, and the loss or gain in each element during melting.

Effects of Primary Elements in Gray Iron Castings

This paper has already included a discussion of the amounts of primary elements found in various grades of pig iron, the influence of the composition of the pig iron on its price, and the behavior of the primary elements in pig iron during remelting. Further detailed discussion of the compounding of metal charges for the numerous types of gray iron produced in cupolas and other melting units is outside the scope of this paper. For additional information on this subject, reference should be made to THE CUPOLA AND ITS OPERATION, Second Edition, Chapters 2, 9, and 27; American Foundrymen's Society, Des Plaines, Ill. (1954).

TABLE 13 — TYPICAL SECONDARY ELEMENTS FOUND IN PIG IRONS ANALYZED AT BATTELLE BY SPECTROGRAPHIC METHODS

	Source of Pig Iron												
Element	USA	USA	USA	USA	USA	USA	USA	USA	Brazil	Brazil	Belgium	Turkey	Germany
	Composition, per cent												
Cu	0.02	0.02	0.1	0.2	0.005	0.01	0.1	0.01	0.002	0.005	0.04	0.07	0.2
Ni	0.02	0.02	0.01	0.1	0.005	0.01	*0.02	0.01	0.02	0.005	0.04	0.2	0.1
Cr	0.01	0.02	0.01	0.05	0.01	0.01	0.02	0.01	0.01	0.002	0.02	0.07	0.01
V	0.005	0.005	0.02	0.004	0.03	0.03	0.01	0.03	0.002	0.002	0.03	0.07	0.03
Mo	0.01	0.02	0.005	0.02	0.001	0.002	*0.005	0.004	*0.001	0.001	0.02	*0.001	0.005
Al	*0.002	*0.002	0.002	0.03	*0.005	*0.005	*0.01	*0.005	*0.002	0.002	0.005	0.005	0.005
Ti	0.10	0.05	0.3	0.05	0.01	0.1	0.1	0.001	0.1	0.09	0.09	0.08	0.1
Sn	0.005	0.005	0.01	*0.005	*0.005	*0.005	*0.005	*0.005	*0.005	*0.005	*0.005	0.03	0.03
Co	*0.005	*0.005	0.02	0.02	0.007	0.007	0.01	0.007	0.005	0.005	0.005	*0.005	*0.005
B	*0.0001	0.002	0.004	0.002	0.0006	0.0007	0.03	0.004	*0.0001	*0.0001	*0.0001	*0.0001	*0.0001
Zr	*0.005	*0.005	0.005	0.005	0.005	0.005	*0.01	*0.005	*0.005	*0.005	*0.005	0.005	0.005
Mg	*0.001	*0.001	*0.001	*0.001	0.02	*0.001	*0.001	*0.001	*0.001	*0.001	*0.001	0.001	0.005
Pb	*0.001	*0.001	*0.001	*0.001	*0.001	*0.001	*0.001	*0.001	*0.001	*0.001	*0.001	0.005	0.005
Cd	*0.001	*0.001	0.001	*0.001	0.005	0.005	*0.001	*0.001	*0.001	*0.001	*0.001	*0.001	*0.001

* = Less than

Effects of Secondary Elements in Gray Iron Castings

If undesirable secondary elements are encountered in gray iron castings, every component of the charge should be examined as a possible source of the contamination. In the usual scheme of things, undesirable elements are more likely to originate from sources other than pig iron. For example, it is possible that at any one time a pile of cupola scrap in the foundry yard may include an unholy combination of plated steel, galvanized iron, brazed iron, porcelain-enameled iron, painted steel, stainless steel, unstripped leaded bearings, and soldered electrical parts. The use of any of these materials, or the use of an unsuitable pig iron, can introduce objectionable amounts of harmful elements into gray iron.

Most secondary elements usually cause only infrequent trouble in general-purpose gray iron castings. The exceptions that are most likely to cause trouble in small amounts are bismuth, lead, and tellurium.

Bismuth is a deoxidizer that sharply reduces the impact strength of gray iron and increases its chill depth when present in amounts of about 0.01 per cent. About 0.001 per cent of bismuth is usually considered the top safe limit. The presence of bismuth causes the iron to develop a sooty fracture and makes the graphite flakes short and stubby. Although weak, iron containing bismuth is said to be highly machinable. The most likely source of bismuth is bearing metal or unstripped motor blocks. It is rarely found in harmful quantities in pig iron.

Lead is retained in gray iron in only very small amounts,* but as little as 0.005 per cent is often troublesome. Lead in an amount of more than 0.010 per cent has been reported to definitely lower impact strength. About 0.007 per cent of lead has caused the formation of "Widmanstätten-like" graphite distributions in heavy-walled castings, and has reduced their tensile strength. It is unlikely that this amount of lead will be introduced by the pig iron, so other sources should be suspected. Bearing metals and enameled scrap are two likely suspects. Lead increases the stability of iron carbide, rapidly raises chill depth, lowers impact and tensile strength, and has been blamed a number of times for severe porosity in iron castings.

Tellurium is a powerful carbide stabilizer and can be expected to cause trouble if it exceeds 0.001 per cent in gray iron. Tellurium rapidly hardens the castings, reduces tensile and impact strength, and increases the chill depth. Although chemically tellurium belongs to the same group as sulfur, the carbide-stabilizing effect of tellurium cannot be neutralized with manganese, as is the case for sulfur, even though the tellurium will combine with manganese. If there is sufficient manganese present, tellurium exists as $MnTe_2$. Without manganese, the tellurium is present as FeTe. Tellurium cannot be detected in pig iron by conventional methods of spectrographic analysis.

Aluminum is a strong deoxidizer and small amounts act as potent graphitizers. Graphite flakes are coarsened, combined carbon is decreased, and the iron is softened and weakened. Small amounts of aluminum have a strong inoculating effect. About 0.1 per cent of aluminum is considered to be the maximum that can be tolerated, but trouble with severe pinholing in thin sections can be expected as soon as the aluminum content approaches about 0.05 per cent. This is about ten times the amount of aluminum usually found in pig iron. In much larger amounts, say, 2 to 10 per cent, aluminum is sometimes used for alloying gray iron for heat resistance. Aluminum is sometimes also used in gray iron that is to be nitrided.

Antimony is usually not found in pig iron. It appears to be harmless up to about 0.01 per cent, but as antimony approaches about 0.1 per cent, it begins to lower impact strength. At 0.35 per cent, antimony drastically lowers tensile strength and impact strength, and increases chill depth and hardness. Bearing metals and enameled scrap are the most likely sources of antimony, and a large percentage of the antimony from these sources will come through a cupola as antimony in the iron.

Arsenic is thought to be harmless in gray iron up to at least 0.1 per cent.

Boron is a very potent stabilizer of iron carbide. Fortunately for the manufacturers of gray iron castings, boron is not encountered in pig iron or in clean scrap in residual quantities large enough to cause trouble. Less than about 0.1 per cent of boron will cause no trouble in most gray iron castings. In special applications, however, it is conceivable that some pig irons might add objectionable amounts of boron to gray iron. For example, as little as 0.03 per cent of boron can significantly increase the chill depth of some irons, especially in thin sections, and make the chill extremely resistant to decomposition by annealing.

As with some other elements, however, very small amounts of boron in the range of thousandths of one per cent can have a slight graphitizing effect on gray iron. Porcelain enamel is a source of boron in cast iron. About half the boron in the enamel in a cupola charge can be expected to be found in the melted iron. In heavy castings with walls 2 to 5 inches thick, about 0.001 to 0.05 per cent boron in the iron has been reported to cause fine "mesh" graphite and a "Widmanstätten-like" graphite which lowered mechanical strength.

Cadmium is not frequently encountered in pig iron, but can be presented in amounts up to about 0.005 per cent or more. If the pig iron or other components of the charge add sufficient cadmium to gray iron to bring the residual cadmium content to about 0.005 per cent, chill depth will increase and tensile strength may decrease appreciably.

Cerium is not found in pig iron, but has a strong influence on the mode of graphitization of all cast irons. If small amounts of cerium, up to, say, 0.02 per

*Reported solubility in cast iron of 0.02 per cent at 2370 F, 0.063 per cent at 2460 F, and 0.11 per cent at 2550 F. (*Gjuteriet*, July, 1955).

cent are added to hypo-eutectic irons, there will be a strong tendency for the iron to mottle, or for the graphite flakes to precipitate in undesirable interdentritic distributions with an accompanying impairment in mechanical properties. About 0.06 per cent of cerium can cause the iron to solidify white. In low-phosphorus hypereutectic irons, about 0.04 per cent of residual cerium will produce the spheroidal graphite characteristic of nodular cast iron. (Nodular iron is discussed in a later section of this paper.)

Cobalt frequently occurs in amounts of up to 0.02 per cent in pig iron. Aside from a mild effect in increasing chill depth, cobalt is not a troublesome tramp element in gray iron castings.

Chromium is a major alloying element in gray iron and can be present in large amounts in pig iron. Chromium is a powerful carbide stabilizer which increases chill depth and hardens the iron. The best strengthening effect with minimum increase in hardness is encountered at about 0.50 per cent chromium. Incidental amounts of about 0.05 per cent and less, as commonly found in pig iron have little effect on gray iron castings. In larger amounts, chromium as a tramp element may have undesirable effects of increasing the chill depth of gray iron. Such an effect is particularly harmful in the manufacture of cast iron pipe in metal molds. Chromium is widely used as an intentional alloy in many irons to improve wear resistance, improve heat resistance, and improve the uniformity of hardness and strength in thick sections of gray iron castings, but these effects are obtained by sacrificing machinability.

Copper is a weak graphitizer with about one-tenth to one-fifth the graphitizing potency of silicon. Copper alloys with gray iron up to about 3.75 per cent copper, progressively strengthens the matrix of the iron, improves machinability, and reduces section sensitivity. In the amounts usually found in pig iron (0.2 per cent or less), copper has practically no effect on gray iron castings.

Molybdenum is ordinarily found in pig iron in only very small amounts which have no significant effect on gray iron castings. Molybdenum is one of the most useful alloying elements in gray iron, and when properly used has no technical disadvantages. It is a mild carbide stabilizer, and as little as 0.2 per cent in gray iron will strengthen it and improve impact strength. In the range of 0.2 to 1.0 per cent, molybdenum will also improve wear resistance and decrease section sensitivity.

Nickel in any amount up to about 4 per cent is usually beneficial in most gray iron castings. It is a mild graphitizer with about one-third the potency of silicon. By alloying with the matrix, nickel strengthens the iron and hardens it. As the nickel content of gray iron is increased up to several per cent, machinability is frequently improved and section sensitivity reduced. It is extremely unlikely that harmful amounts of nickel will be present in pig iron.

Tin has a bad reputation in gray iron because it is said to embrittle the iron and to lower its impact strength. Tin is readily picked up by gray iron from bearing metals and tin plate, and should be held to a maximum of about 0.01 per cent in gray iron castings. Some pig irons contain enough tin to have a decidedly depressing effect on the tensile and impact strengths of gray iron castings.

Titanium is a powerful deoxidizer and reducing agent in gray iron. In amounts up to 0.2 per cent it strengthens high-carbon soft irons and weakens low-carbon irons. Its general effect in small amounts is to refine the graphite flakes, promote the formation of ferrite, and soften the iron. Titanium is a touchy element because too much of it can cause overgraphitization of the iron. Most pig irons contain 0.10 per cent titanium or less, and in these amounts the titanium is not harmful in gray cast iron.

Some pig irons contain much larger amounts of titanium, and there is a controversy whether these larger amounts are beneficial or harmful. For example, one high-carbon low-phosphorus pig iron with a silicon content of about 0.50 per cent contains a normal amount of about 0.07 per cent titanium, but another grade of the same pig iron with about 2.50 per cent silicon contains up to 0.35 per cent titanium. Another commercial pig iron contains 0.40 to 0.50 per cent of titanium intentionally added for the claimed purpose of refining the graphite flakes of both the pig iron and the castings made from it. It is common to find that pig irons which are high in titanium are also high in vanadium.

Tungsten is rarely found in pig iron. It has little effect on the strength or hardness of gray iron up to about 1 per cent of tungsten, and it is unlikely that tungsten will be encountered in pig iron in anything approaching this concentration.

Vanadium is a powerful carbide stabilizer widely used as an alloying element in gray iron. It is encountered in pig iron in about the same amounts as chromium, say, 0.01 to 0.2 per cent, and has somewhat the same effects as chromium on gray iron. Vanadium is sometimes intentionally added to gray iron in amounts up to about 0.40 per cent, at which level it progressively improves tensile strength, impact strength, wear resistance, and heat resistance. Even as little as 0.1 per cent of vanadium, however, will have an effect upon the properties of gray iron castings. For many applications, the effects will be beneficial. For other applications, especially where it is desired to soften the iron by annealing, the effects of vanadium will be considered harmful. Pig irons that contain substantial amounts of vanadium are likely also to contain substantial amounts of titanium.

Zinc volatilizes so readily that it is not encountered in pig iron in detectible amounts. Up to about 0.01 per cent, it appears to be harmless in most gray iron castings.

Zirconium is a powerful deoxidizer in the same class with titanium. It is often found in pig iron in amounts of about 0.005 per cent. At this level it has little effect on gray iron castings, and little has been published about its effects in larger amounts.

For additional information on the effects of secondary and residual elements in pig iron and gray

iron, the reader is referred to the ALLOY CAST IRONS HANDBOOK (AFS), and to "The Effect of Copper Addition Contaminants on Mechanical Properties of Gray Cast Iron" by K. Rose and C. H. Lorig, TRANSACTIONS, AFS, 1947.

Pig Iron in Manufacture of Nodular Iron Castings

Compounding of metal charges for the manufacture of nodular iron castings follows the same general pattern as for gray iron, but closer attention must be given to several factors involving the primary elements carbon, manganese, phosphorus, and sulfur. These factors are such that it is common practice to use a higher percentage of pig iron in the charges for nodular iron than in the charges for gray iron.

When fairly high ductility is desired in nodular iron, it is possible to approach this by two routes. (1) as cast, or (2) by annealing. When specific properties are to be obtained without heat treatment, the degree of control during melting and processing must be quite close. Small variations in chemical analysis can have an important effect on as-cast properties, especially on ductility. When the iron is to be annealed (for ductility), or otherwise heat treated (for strength or hardness), larger variations in chemical composition can be tolerated, and can be somewhat offset by the thermal treatment.

Effects of Primary Elements in Nodular Iron Castings

Common ranges of the primary elements in nodular cast iron are given in Table 14.

TABLE 14 — TYPICAL COMPOSITION OF NODULAR IRON CASTINGS

	Composition, per cent				
	T.C.	Si	Mn	P	S
Ferritic high-ductility*	3.5-4.2	1.2-2.0	0.35 max	0.08 max	0.005-0.015
Pearlitic high-strength*	3.2-4.0	1.2-2.7	0.6-0.8	0.10 max	0.005-0.015
Heat resisting	3.2-4.0	2.7-3.5	0.3-0.4	0.10 max	0.005-0.015

*High-strength grades are usually used as-cast. When high ductility is desired, the irons are usually annealed, but reasonably high ductility can be obtained in the as-cast condition if the composition of the iron is properly selected, and if certain carbide-stabilizing secondary elements are avoided.

Carbon in the form of tightly packed spheroids in nodular iron has a much smaller weakening effect on the iron than graphite in the form of sprawling flakes in gray iron. For this reason, nodular iron castings often contain more carbon than gray iron castings, so that advantage can be taken of the higher castability and fluidity imparted by high carbon content. To obtain high carbon content in the melted iron, it is common to use large percentages of high-carbon pig iron, or to melt in a cupola with a basic slag so as to promote carbon pickup by the iron.

Magnesium is the usual additive made to the molten iron to produce the spheroidal graphite. Magnesium is effective in either hypo-eutectic or hypereutectic irons, if added in suitable amounts. The spheroidal graphite is also produced by treatment with other metals, notably cerium. If cerium is used as the main nodulizing element, the iron must be hypereutectic (that is, high in carbon content). This requirement, of course, makes control of the carbon content of cerium-treated irons much more important than for magnesium-treated irons.

Sulfur is an extremely important element in the manufacture of nodular iron. Both of the common "nodulizing" elements (cerium and magnesium) are potent desulfurizers. They do not produce their desired metallurgical effects until the sulfur content of the iron is lowered to about 0.01 per cent. If a small amount of cerium or magnesium is added to a molten iron containing, say, 0.10 per cent sulfur, the effect is first to desulfurize the iron to about 0.01 per cent sulfur. Only then will the remaining amount (if any) of cerium or magnesium become effective in promoting the nodular-graphite structure.

Although it is possible to add sufficient magnesium or cerium to desulfurize the iron and leave enough residual magnesium or cerium to nodulize the graphite, this is very difficult to do in practice because the residual amount of magnesium or cerium must be closely controlled. From a practical standpoint, it is much more satisfactory to desulfurize the iron to a consistent low value, say, 0.02 per cent sulfur, and then to add a controlled amount of cerium or magnesium.

Because of the value of low-sulfur irons in the manufacture of nodular iron, the iron is melted in basic cupolas, or melted in acid cupolas and then desulfurized at the spout with calcium carbide or soda ash. Today, the latter practice probably accounts for more tonnage of nodular iron than melting in basic cupolas. The use of low-sulfur pig iron is not so beneficial as one might expect when melting in a cupola, because the pickup of sulfur from the coke during acid melting, or the decrease in sulfur during basic melting exert much more influence than the composition of the raw material.

Phosphorus is an embrittling agent in nodular cast iron. For this reason, it is commonly maintained below 0.10 per cent. Under some conditions, it may be desirable to use phosphorus to obtain increased fluidity and heat resistance at the expense of ductility. Under these conditions, phosphorus contents as high as 0.6 per cent in the melted iron might be desirable. Because it is not practical to reduce the phosphorus content of iron during foundry melting operations, close attention must be given to the phosphorus content of each raw material.

Manganese is a strengthening agent in nodular iron, but is commonly maintained at a maximum of about 0.35 per cent when high ductility is desired. Because of the low sulfur content of nodular iron, and the effects of cerium and magnesium in lowering the sulfur content of the iron, it is not necessary to consider manganese in ductile iron as a neutralizer for sulfur. Almost all the manganese that is present can be considered as a carbide stabilizer which strengthens and hardens the iron while reducing its ductility.

As with any carbide stabilizer, excess manganese also makes it more difficult to produce a completely ferritic matrix by annealing.

Silicon in nodular iron does not have as great an effect upon the section sensitivity as silicon in gray iron, and is usually present in a range of about 2.0 to 2.5 per cent. Because silicon raises the strength of the iron (by solid solution in the matrix), the silicon content of high strength grades of nodular iron is usually toward the high side of the range. Another advantage of silicon contents on the high side of the range is that more primary graphite is precipitated during freezing, and the solidification shrinkage of the iron is reduced.

Because silicon lowers the ductility of the iron, it is usually kept on the low side of the range for high-ductility irons. An increase in the silicon content of nodular iron also raises the transition temperature at which notch brittleness occurs.† A ferritic nodular iron containing 2 per cent silicon will be ductile below 0 F, but a ferritic nodular iron containing 3.5 per cent silicon may be notch-brittle at 100 F.

Effects of Nodulizing Elements in Nodular Iron Castings

Magnesium is the most common element used to produce the spheroidal graphite structure. Magnesium is sometimes found in amounts of up to 0.005 per cent in pig iron, but is rapidly eliminated from the iron during remelting because of the low boiling point of magnesium. About 0.03 to 0.10 per cent of residual magnesium is usually required in the iron to produce the spheroidal structure. Small amounts usually have only a desulfurizing effect, or produce the fine so-called "quasi" graphite flakes. Larger amounts of residual magnesium permit it to exert its very strong carbide-stabilizing effect, and may cause the iron to solidify white. Such over-treated irons can often be reclaimed by suitable annealing treatments, and can produce a nodular iron just as satisfactory as a properly treated iron.

The main metallurgical disadvantage of the use of magnesium in nodular irons is that it is sensitive to the influence of certain "subversive" elements to be discussed later. The main metallurgical advantage of magnesium in nodular iron is that it can be used successfully to treat both hypo-eutectic and hypereutectic irons. Another advantage of magnesium is that it is generally less expensive to use than most of the other nodulizing materials.

Cerium was the nodulizing element first disclosed publicly.* It rapidly fell into disuse, however, because of the advantages of magnesium. Subsequent research has shown, however, that cerium is an extremely valuable element, even in magnesium-treated irons.** Although the ability of magnesium to produce spheroids of graphite is offset by certain "subversive" elements, these subversives can be effectively neutralized by small amounts of cerium. By adding sufficient cerium to give a residual of about 0.005 per cent, the effects of the subversives are so neutralized that full advantage can be taken of the magnesium treatment. To take advantage of the strong points of both cerium and magnesium, it is common practice to add both elements at the same time, for example as cerium-bearing magnesium ferrosilicon, or as an alloy of nickel, magnesium, and cerium.

Other Elements which have at one time or another been reported to produce at least a limited amount of spheroidal graphite in suitable base irons, or under special conditions, include calcium (a residual content of at least 0.04 per cent), titanium, boron, lithium, strontium, barium, sodium, potassium, bismuth, selenium, thorium, tellurium, and other rare earths in addition to cerium. To some extent, these elements have been found to have a cumulative effect on graphitization. The effect of these elements is, however, not always consistent. For example, under some conditions, tellurium and titanium interfere with the formation of spheroidal graphite.

Effects of Secondary Elements in Nodular Iron

Some "subversive" elements which have been shown to interfere *with the formation* of nodules of graphite when the iron is treated with magnesium will be discussed below. It is important to realize, however, that all the subversive elements may not yet have been recognized.

Antimony in amounts of more than 0.004 per cent is detrimental to the magnesium treatment. It is a subversive element but it is readily neutralized by proper treatment with cerium. As with all subversives, it is more damaging in thin castings and less harmful in heavy sections. It is important to realize that the effects of the subversives are additive, so that a very small amount of each of several subversive elements can add up to a damaging total. Many pig irons contain enough subversive elements to make the production of nodular iron very difficult unless the cerium treatment is used.

Lead is a subversive element in magnesium-treated nodular iron when present in an amount greater than about 0.009 per cent. The effect of lead can be neutralized by cerium.

Bismuth is a subversive element when present in amounts over 0.003 per cent, but it can also be neutralized by cerium.

Titanium is a subversive element found in fairly large amounts in some pig irons. Above about 0.08 per cent of titanium will drastically interfere with the attainment of good spheroidal graphite, unless the titanium is neutralized by cerium. It has already been pointed out that some U.S. pig irons contain as much as 0.50 per cent titanium.

Aluminum has an effect similar to that of titanium, but up to 0.10 per cent can be tolerated before becoming subversive. It too is neutralized by cerium. In

†"Notch Ductile High Strength Nodular Irons," G. A. Sandoz, H. F. Bishop, and W. S. Pellini; *Transactions*, ASM, Preprint No. 21, 1955.

*H. Morrogh, "Production of Nodular Graphite Structures in Gray Cast Irons," TRANSACTIONS, AFS, vol. 56 (1948).

**H. Morrogh, "Influence of Some Residual Elements and Their Neutralization in Magnesium-Treated Nodular Cast Iron," TRANSACTIONS, AFS, vol. 60 (1952).

one respect, aluminum is less troublesome than titanium—aluminum is not usually encountered in pig iron in amounts greater than about 0.005 per cent.

Tin is one of the most troublesome subversive elements when making a ferritic nodular iron. About 0.04 per cent of tin is a powerful pearlite former, stronger than 0.9 per cent of manganese. The subversive effect of tin on graphite formation can be neutralized by cerium, but the carbide-stabilizing effect of tin on pearlite cannot be neutralized. Because some pig irons contain about 0.04 per cent tin, the pig iron must be carefully selected if high-ductility ferritic iron is desired.

Arsenic is similar to tin in its behavior but is not considered subversive until it reaches about 0.09 per cent. Like tin, its powerful carbide-stabilizing effect cannot be neutralized with cerium. Unlike tin, it is rarely found in pig iron.

Selenium and tellurium are both subversive elements that can be neutralized by proper treatment with cerium.

Copper is often considered a subversive element, but the most recent work shows that copper itself is not subversive—it merely intensifies the effects of subversive elements. If subversive elements are absent up to 3 per cent copper is not believed to be usually harmful in nodular iron, but there have been reports that over 2 per cent of copper will sometimes cause the graphite spheroids to become ragged. In the presence of 1 per cent of copper, however, as little as 0.03 per cent of titanium exerts a strong subversive effect. The role of copper is important because some addition agents containing magnesium also contain large amounts of copper.

Zinc, cadmium, and calcium are not subversive elements. In the small amounts usually encountered, they are considered beneficial to the attainment of good spheroidal graphite in magnesium-treated iron.

Chromium is a useful alloying element in ductile iron. It should probably be limited to about 1 per cent if good ductility is to be developed. The usual amounts encountered in pig iron will usually not be harmful to nodular iron.

Nickel is probably the most desirable alloying element in nodular iron. It is thought to be beneficial to at least 35 per cent.

Molybdenum up to at least 0.6 per cent is not subversive, and is a useful element for strengthening nodular iron.

Pig Iron in Manufacture of Malleable Iron Castings

The main problem in manufacture of malleable iron castings is to control primary and secondary elements so as to prevent formation of primary graphite during solidification of the casting, and to promote graphitization of iron carbide during annealing. As in the case of nodular iron castings, the close control required in the manufacture of malleable iron often calls for the use of fairly large percentages of pig iron in the charge. The advantages of pig iron are that it has a known composition, something that most scrap does not have. Pig iron is also low in undesirable elements such as phosphorus, sulfur, and carbide stabilizers that can cause trouble during the annealing operation.

Carbon and silicon are the main elements which effect the final properties of malleable iron. The influence of other elements is mainly on the speed of graphitization of the iron. In the discussion below, some elements encountered in pig iron will be identified as retarders of graphitization and other elements as promoters of graphitization. Some elements perform dual roles as graphitizers if present in small quantities and as retarders if present in larger amounts. In the manufacture of malleable iron, it is often possible to offset the retarding effect of one element by the graphitizing effect of another. These combinations, however, must be intelligently used and carefully controlled.

The usual range of composition for malleable iron is as follows:

	Per Cent
Carbon	2.00 to 2.70[a]
Silicon	0.80 to 1.60[b]
Manganese	0.55 max[a]
Sulfur	0.18 max[a]
Phosphorus	0.20 max

[a] Higher in cupola malleable iron.
[b] Lower in cupola malleable iron.

Most malleable iron today is produced by duplex processes. The initial melting is usually done in a cupola, and finishing is usually done in an air furnace or an electric furnace. One large producer which manufactures about one-third of the malleable iron made in the United States duplexes the iron from a cupola to an electric furnace. A cupola charge for duplex malleable may contain up to 15 per cent of pig iron.

"Cupola malleable" is melted in a cupola and is cast without further treatment in a second furnace. Substantial amounts of malleable iron are still produced from cold melts in an air furnace (especially in older and smaller shops). The construction and operation of an air furnace resemble those of a small open-hearth furnace. A typical metal charge for an air furnace might contain 40 per cent pig iron, 50 per cent remelt from the foundry, and 10 per cent purchased malleable iron scrap.

The pig iron for producing malleable iron is selected for low phosphorus content (usually 0.20 per cent maximum) and freedom from harmful secondary elements. The components of the metal charge are selected and blended together after taking into consideration the changes in carbon, silicon, and manganese content expected during melting.

Effects of Primary Elements in Malleable Iron Castings

Carbon in properly made malleable iron exists entirely as ragged nodules of temper carbon. Poorly made malleable iron may contain flakes of primary graphite which greatly harm the mechanical properties of the iron. Special grades, such as pearlitic malleable, contain some carbon as iron carbide, but

strictly speaking, such irons are not true malleable irons. The lower end of the carbon range in malleable iron is governed by the fact that irons containing less than about 2.00 per cent of carbon are more difficult to cast, have high melting temperatures, and have relatively poor fluidity and castability.

With more than about 2.70 per cent carbon, it is difficult to control the formation of primary graphite, and the strength and ductility of the iron are lowered because more graphite is present after annealing. Within this range of 2.00 to 2.70 per cent carbon, the carbon content of the iron is selected mainly on the basis of the type of casting to be made and details of the manufacturing procedure. Thin castings are often made from iron with carbon contents on the high side of the range, and thick castings are often made with carbon contents on the low side of the range.

Silicon in malleable iron must be carefully controlled on the basis of the carbon content of the iron and the section size (cooling rate) of the casting. Silicon is a strong graphitizer in malleable iron and is responsible for the fact that white iron can be graphitized by an anneal to form malleable iron. High silicon content is desirable to speed the anneal. The silicon content, however, cannot be so high that it will cause the iron to graphitize during solidification and form flake graphite.

In addition to its effect on graphitization, silicon is useful because it alloys with the ferrite in the matrix and strengthens it. This advantage is largely offset by the fact that the higher silicon contents tend to make the temper carbon more sprawly, and thus reduce the strength of the iron. The silicon content of malleable iron, therefore, is a compromise which usually puts it in the range of 0.80 to 1.60 per cent. Because of its effect in promoting primary graphitization, silicon cannot be used as an alloying addition unless the iron is poured in a metal mold to produce a very fast cooling rate. The effect of silicon in promoting primary graphite can be partially offset by a high degree of superheat of the iron before casting.

Phosphorus is an element which cannot be avoided in malleable iron. A maximum of 0.18 or 0.20 per cent is usually set because above this amount it begins to lower impact strength and to make the malleable iron sensitive to blue brittleness and galvanizing embrittlement. Up to about 0.20 per cent, phosphorus is beneficial in strengthening the matrix. In this respect, it is about six times as potent as an equal amount of silicon. Phosphorus is a mild graphitizer, so that up to at least 0.35 per cent it has no detrimental effect on the speed of annealing.

Sulfur is another unavoidable element in malleable iron, because it is present in the raw materials and in most fuels. It is a very potent carbide stabilizer which seriously interferes with annealing unless it is neutralized with manganese to form manganese sulfide. To avoid the strong retarding effect of sulfur on graphitization, the iron should contain at least twice as much manganese as sulfur.

Manganese by itself is a retarder of graphitization, but the first manganese that is added combines with another retarder, sulfur, to form a nonmetallic inclusion that does not retard graphitization. Thus, two elements which individually will each have a powerful effect on lengthening the annealing time will combine to produce a compound which is quite harmless.

The excess manganese over that required to neutralize the sulfur, however, shows a powerful retarding effect on the rate of graphitization. For example, 0.70 per cent manganese and 0.06 per cent sulfur will require more time to anneal to a ferritic matrix than will 0.30 per cent manganese and 0.06 per cent sulfur. Because of its effect in stabilizing pearlite, manganese in the range of 0.50 to 0.80 per cent is widely used in the manufacture of pearlitic malleable iron. Manganese is one of the few elements used as an alloy in malleable iron. The alloying effect of manganese in strengthening the matrix is about twice that of silicon, and, of course, the stabilizing element manganese can be used in many cases where the graphitizer silicon cannot be used.

Effects of Secondary Elements in Malleable Iron Castings

Secondary elements found in pig iron can have major effects in malleable iron, because of the need for close and precise control of the graphitization of the iron. Only three elements are considered as commercial alloying materials (manganese, copper, and molybdenum) and each of these must be carefully controlled. Very small amounts of some other elements, such as chromium, vanadium, sulfur, tin, and tellurium, can have potent and often disastrous effects on malleable iron.

Copper is a mild graphitizer that has the beneficial effect of strengthening the ferritic matrix. In amounts of about 0.04 per cent, copper has a strengthening effect with little influence on the annealing time. Copper up to 1.25 per cent is a common alloy in standard malleable iron to increase its strength and resistance to corrosion. About 1.0 per cent of copper has a graphitizing power equivalent to about 0.10 per cent of silicon. It is unusual to find more than 0.10 per cent of copper in pig iron, so it is not a critical element so far as pig iron is concerned.

Molybdenum is a moderate retarder of graphitization. It is usually found in pig iron in amounts up to about 0.03 per cent. It is believed that three or four times this amount are necessary to have an appreciable effect in lengthening the annealing cycle. Molybdenum is often used as an alloying element in malleable iron, usually in combination with copper which offsets the carbide-stabilizing effect of the molybdenum.

Chromium is a strong carbide stabilizer that has the undesirable effect of greatly lengthening the annealing of malleable iron. Chromium is rarely found in excess of 0.04 per cent in pig iron (except in special types of pig iron), and in this amount is not too troublesome. With about 0.03 per cent chromium in the white iron, however, a definite lengthening of the first and second stages of the anneal may occur. At 0.30 per cent chromium, annealing is seriously impaired.

Vanadium behaves similarly to chromium and produces a very stable carbide. Less than 0.10 per cent of vanadium will seriously retard annealing. Most pig irons contain no more than about 0.03 per cent vanadium but some pig irons contain vanadium in dangerous amounts.

Tin is another strong retarder of graphitization whose effect on annealing will be felt at less than 0.09 per cent. This is about three times as much tin as usually encountered in pig iron, but tin pickup from poorly selected scrap can be a major problem in a malleable iron foundry. Up to about 0.05 per cent, tin appears to have no effect on the first stage of the anneal, but above about 0.02 per cent it begins to reduce the rate of second-stage annealing. One effect of tin is to combine with sulfur and free manganese to act as a retarder.

Boron has an inconsistent but very important effect in malleable iron. Small amounts of 0.001 to 0.004 per cent reduce the annealing time. About each 0.001 per cent of boron will offset the effect of about 0.03 per cent of chromium. About 0.001 per cent of boron can be expected to reduce the annealing time by about 10 per cent and to have little effect on tensile strength. Above about 0.01 to 0.05 per cent, boron, boron carbides appear in the microstructure and boron begins to retard graphitization during annealing. At 1.0 per cent boron, graphitization is completely inhibited.

Boron, therefore, is a useful and potent graphitizer in small amounts, but a dangerous and strong retarder in large amounts. If metallographic examination shows that a boron-bearing iron has very fine nodules of temper carbon near the surface of the casting, the chances are that there is sufficient boron present to have materially lengthened the annealing time.

Aluminum in small amounts is a strong graphitizer, as little as 0.1 per cent can cause the formation of primary graphite. At this level or below, there is usually little measurable effect on the annealing time. At about 0.20 per cent, aluminum improves first and second stage graphitization. At about 0.60 per cent, aluminum becomes a powerful carbide stabilizer and interferes with fast annealing. Probably the safest maximum for aluminum is about 0.06 per cent. This is much higher than the aluminum content of most pig irons. It has been reported that an addition of 0.05 per cent aluminum to malleable iron will neutralize the effect of 0.08 per cent chromium on annealing time.

Titanium behaves similarly to aluminum, but is of considerably more concern because it is encountered in pig iron in relatively large amounts. Pig irons often contain about 0.10 per cent of titanium. At this level, it is a mild graphitizer, but is on the borderline approaching the level where it will begin to interfere with annealing. As with aluminum, a few hundreths of one per cent of titanium are potent as a graphitizer.

Nickel is in the same class with copper. It graphitizes mildly and strengthens ferrite without lengthening the annealing time with amounts up to as much as 0.50 per cent. Except for special pig irons, little attention need be given to the nickel content of pig iron.

Zirconium is a potent graphitizer in amounts of a few hundreths of one per cent, and a mild graphitizer in amounts of about 0.02 per cent. It is found in pig iron in very small amounts (say, less than 0.01 per cent), and has no adverse effect on malleable iron up to at least 0.08 per cent.

Zinc is a strong retarder of graphitization and will lengthen the annealing operation.

Tungsten is a moderate retarder of graphitization. Its effect is not important because it is found in pig iron in quantities much lower than the 0.10 per cent believed necessary to influence annealing appreciably.

Cerium and lanthanum in the amounts likely to be encountered in simple remelting without alloying are not likely to be harmful in malleable iron. They are mild graphitizers themselves, but by combining with sulfur can free manganese to act as a retarder of graphitization.

Selenium and tellurium behave similarly to sulfur and promote dense temper carbon. In very small amounts, say, 0.001 per cent, they are mild graphitizers. In larger amounts, they are carbide stabilizers. About 0.003 to 0.005 per cent of tellurium can be used to prevent mottling in heavy sections without greatly increasing the annealing time. This is a very useful range for tellurium. At about 0.15 per cent, either tellurium or selenium lengthens the annealing time.

Antimony is rarely found in pig iron. Very small amounts (such as 0.003 per cent) are mild graphitizers, but as little as 0.05 per cent (and perhaps as little as 0.02 per cent) will retard second-stage annealing.

Cobalt is a mild graphitizer which is not likely to cause any trouble in the amounts found in pig iron.

Uranium is a mild graphitizer rarely found in pig iron.

Calcium is a mild graphitizer that has little effect in the amounts found in pig iron.

Arsenic is a rarity in pig iron, and is not expected to cause trouble in malleable iron up to at least 0.15 per cent.

Bismuth is rarely found in pig iron, but small amounts are sometimes intentionally added to white iron to retard primary graphitization and mottling. The usual amount added is about 0.04 per cent. At this level, bismuth mildly retards second-stage graphitization.

Lead has been found to have no effect on either first or second-stage annealing in amounts up to about 0.03 per cent. Although large amounts of lead are considered to retard graphitization during the anneal, quantitative data are lacking.

Much of this information on the effect of minor elements in malleable iron has been abstracted and correlated from the works of Schwartz.*

*H. A. Schwartz, "Principles of Graphitization," TRANSACTIONS, AFS, vol. 50, p. 1 (1942).

H. A. Schwartz, "Solved and Unsolved Problems in the Metallurgy of Blackheart Malleable," *Foundry Trade Journal*, vol. 84, June 17, 1948.

"Heredity" in Pig Iron

Over a period of many years, foundrymen have observed that there is a tendency for the microstructure and the properties of some pig irons to be repeated in iron castings, even though the composition and cooling rate of the castings are quite different from those of the pig iron. This effect has been called "heredity."

Only one example of "heredity" will be given here, but many examples could be cited. Some molten pig iron in a ladle directly from a blast furnace can be chill cast so as to produce pigs with fine graphite flakes, or no flakes at all (white iron). Other molten pig iron from the same ladle can be cast in refractory molds so that large coarse graphite flakes are formed in the pigs.

These two types of pigs differ in their structure because of their difference in cooling rate. This is easy to understand, but when these two types of pigs are remelted, say in a cupola with added steel scrap and iron scrap, the behavior we call "heredity" shows up. By our normal standards, the chemical composition of these two types of pigs would be considered identical because they came from the same ladle of iron. On remelting to give cast irons of what we think is identical composition at the cupola spout, the cast iron melted from the chill-cast pig will produce castings with a finer graphite structure than castings made from the slow-cooled pig. This behavior is one of the effects we call "heredity."

The selection of the word "heredity" with respect to pig iron is unfortunate because foundrymen and metallurgists are not biologists. The word "heredity." however, does not imply anything mysterious to a biologist, who uses it simply to refer to the resemblance between an organism and its ancestors. True enough, the reasons for the many hereditary characteristics of plants, animals, and humans were once mysterious, even to the biologist, but the mystery has been torn away by years of research which have shown the effects and importance of many biological factors called "genes."

In the same way, some of the "hereditary" effects of pig iron have already been explained by the "genes" of chemical composition. The "genes" of pig iron which are best understood today are those of solids (such as graphitic carbon, boron, and titanium, to mention a few), because these factors are the easiest to measure and to observe. Just as in biology, however, the full explanation of all heredity effects in pig iron has not yet been developed. This does not mean that we are dealing with *unexplainable* phenomena. It merely means that we are dealing with phenomena which are *unexplained* as of now, but which can probably be explained later after more research is conducted and after our methods of measurement and observation are improved.

The list of names of men who have tried to determine the causes of heredity in pig iron reads like a "Who's Who" of the foundry industry. The list includes the names of Piwowarsky, Boegehold, Jominy, Hurst, Norbury, and Jungbluth, each of whom recognized the existence of the effect. There were also others (for example, MacKenzie, Bolton, and Portevin), who had the idea that because there were so many elements and compounds which could be present in pig iron, and not be detected by conventional analyses, it was not surprising that the effects of some of these materials could be carried through remelting operations.

The history of the study of heredity in pig iron is a bewildering story. Different experimenters have performed similar experiments and reached conflicting conclusions. Out of years of experimentation, however, have come two conclusions which stand out:

(1) There are often characteristics or properties of pig iron which carry through remelting operations.

(2) All of these "hereditary" effects have not yet been completely explained by chemical composition or by any existing theories.

Three main theories have been developed. Each has its adherents, but none explains all the effects that have been observed.

These theories, briefly stated, are as follows:

(1) The nature and behavior of the graphite in different pig irons differ. Some pig irons contain graphite which is little affected by remelting operations, and other pig irons contain graphite which dissolves during similar remelting so that it will not nucleate the precipitation of graphite during solidification (or during annealing in the case of malleable iron).

(2) Some pig irons contain a "slime" of ferrous silicates which pass through remelting operations and nucleate the formation of graphite flakes, while other pig irons do not contain this "slime," or contain less of it, and hence, do not graphitize so readily.

(3) The behavior of pig iron during melting is influenced by the presence or absence of dissolved gases.

Fortunately for everyone interested in this subject (and every foundryman should be interested), an eminent American metallurgist prepared a review, which remains a classic today, even though it was prepared 22 years ago.* After a complete study of all the evidence available at that time, Dr. Horace Gillett reached the following conclusions:

"It is quite evident that the heredity problem is quite involved. The theory of persistent kish and its solution, a theory attractive because of its simplicity, seems none too well grounded. That of submicroscopic silicate or oxide nuclei is difficult of direct proof and it is by no means certain that it explains all of the phenomena.

"Certainly the basic facts need to be soundly established by careful chemical analysis, and before we theorize about "differences in behavior without change in chemical composition" we ought to be sure that the composition is the same, and not be satisfied with stating that the same raw material was used."

Following Dr. Gillett's paper, extensive experimentation was conducted at Battelle on this subject over a period of seven years. Pig irons from a number of sources were melted and remelted in different types

*H. W. Gillett, "Heredity in Cast Iron," *Metals and Alloys*, vol. 5 (1934).

of melting units. The pig irons were used to produce gray irons, white irons, and malleable irons. Tests were made by conventional methods of chemical analysis, spectrographic analysis, and vacuum-fusion analyses for gas contents. The pig irons and their products of remelting were tested and examined for metallographic structure, shrinkage characteristics, strength, annealability, chill depth, response to inoculation, tendency to hot tear, castability, density, and about every other test that anyone could think of at the time.

Attempts were made to reproduce or influence the characteristics of the pig irons by adding to them many different types of materials. The program involved over a dozen pig irons, including charcoal pig iron and some other special types. Literally thousands of tests were conducted, and the results were subjected to statistical evaluation. Out of this program came the indisputable fact that pig irons can have characteristics which persist to some extent regardless of the type of melting unit employed or the type of cast iron produced. Furthermore, the study showed that at least some of these characteristics were established in the pig iron *before* it was cast into pigs. Although the study revealed methods by which the behavior of the pig iron could be influenced during remelting, it was not able to explain conclusively the reason or reasons for all hereditary effects.

In 1944, Dr. Gillett presented the AFA Annual Lecture. It was entitled "Cupola Raw Materials."* In the ten years between 1934 and 1944, Dr. Gillett had concerned himself with this problem, both in the work at Battelle and elsewhere. But, unfortunately, the situation in 1944 was just about the same as in 1934, except that now there were much more data to wade through, and many more conflicting interpretations from similar experiments.

In his 1944 lecture, Dr. Gillett had the following to say about heredity in pig iron:

"Cupola products do show considerable non-uniformity, some of which results from known and understandable causes affecting composition as to elements normally analyzed for, but some of which arises from causes less well understood and not yet subject to adequate control.

"Read any one article and it will be found that the author has a hypothesis to fit some of the facts, but in discussion, someone is sure to bring up equally well-established facts that do not fit the hypothesis.

"Despite all the observed facts, the reasons for this much-discussed "heredity" of pig iron—differences that persist after remelting and affect the type and distribution of graphite in the cast iron—remain obscure."

Today, in 1956, we are not much further along in our knowledge of this subject than we were 26 years ago. Meanwhile, however, steady pecking away at the problem has led most investigators, including the author, to believe that, although the hereditary characteristics of pig iron are not completely understood, they will be understood some day if experimentation is continued. When the explanation does come, it may be on the basis of chemical composition, or something equally simple. Many of us will wonder why we were so blind as to miss an obvious explanation, and we will realize that we were right on top of an answer all the time.

Remember what happened in the metallurgists' search for an iron in which the graphite would be in a spheroidal form in the as-cast iron. For many years, metallurgists dreamed of the superior properties which such an iron would probably have. The main problem was that no one seemed to know how to make it. Then in 1948 at the AFS Annual Meeting in Philadelphia, we were told how to make this iron. Many of us were surprised at first, but as we sat back and took another look at our own work, we saw that it was our own fault that we had not found the answer sooner. Many of us before 1948 had produced irons containing nodules of graphite in as-cast iron—our experiments were trying to tell us how to do it reproducibly, but we misinterperted the results of the experiments and failed to find the solution to the problem. Then, in 1948, with the answer in front of us, it was possible to go back over our experimental data and find our mistakes.

We found that, like most puzzles, the answer was simple and right in front of us all the time, but that we had lacked that certain "spark of genius" which eventually led to the solution of the problem. The answer to the problem of "heredity" in pig iron may well come the same way, and just as suddenly, but it will not come unless we continue to work for it. We know for sure that we have a tough problem to tackle, and we know just as surely that we will not lick it by ignoring it.

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*H. W. Gillett, "Cupola Raw Materials," TRANSACTIONS, AFA, vol. 52 (1944).

DISCUSSION

Chairman: W. W. LEVI, Lynchburg Foundry Co., Lynchburg, Va.

Co-Chairman: E. J. BURKE, Hanna Furnace Corp., Buffalo, N. Y.

Secretary: C. F. WALTON, Gray Iron Founders' Society, Cleveland.

E. C. MATHIS[1] *(Written Discussion):* Mr. Lownie has performed a commendable service in documenting the current status of those phases of pig iron which are of concern to the foundrymen. It represents so faithfully the current situation that except for a few minor items which are scarcely worthy of mention, the writer can make no critical comment.

However some of the early history concerning grading and classifying pig iron might be a worthy supplement to this paper. There are very few alive today who have first-hand knowledge of the early practice but from some of them a partial story has been learned.

Up to the turn of the century all pig iron was graded and selected for use on the basis of the appearance of the fracture, augmented by the 'ring' or 'timbre'. It was considered that the coarser the grain, the more suitable the pig iron for producing soft, machineable gray iron castings. The coarsest grain was called No. 1 and as the grain grew finer, higher numbers were assigned. No. 6 was 'white' iron; No. 5 was partially white or "mottled"; No. 4 was called "gray forge" and was probably fine grained with a light gray background. No. 2 had further subdivisions of No. 2-X and No. 2 plain. No. 3 was also broken down further into No. 3-X and No. 3 plain. Salesmen in those days would carry with them, sometimes in fancy, plush-lined cases, sections or pieces of pig iron showing representative fractures.

Admittedly today the clean fracture of a casting affords a rough appraisal of its mechanical properties when consideration is given to its cooling rate. It was the lack of knowledge of the cooling rate of any particular pig iron that was one of the weaknesses of the fracture grading system. In the old sand pig beds, sometimes additional sand was thrown over the pigs after the open molds were filled to reduce radiated heat and sometimes a hose played water over them to speed up the freezing.

Around 1900 some pioneering chemists began to determine silicon in pig iron and cast iron and in the relatively short span of eight or nine years the silicon content of pig iron became the principal basis for grading. It was a natural consequence that attempts were made to correlate the then popular numbered grades with a silicon range, but the correlation was not the same country-wide. Identification by number has virtually disappeared as the ranks of the old-timers become thin. Reference to No. 2 grade is still occasionally encountered, meaning a base grade silicon range of 1.75 to 2.25 per cent and still more rarely the terms of No. 1 and No. 3 pig iron come up.

Following closely the practice of determining silicon came manganese, phosphorus and sulfur determinations. By the middle teens, pig iron was almost universally graded on the basis of the content of those elements. Today, the standard grades defined by the American Iron and Steel Institute and American Society for Testing Materials are universally used in this country, as Mr. Lownie appropriately points out in his paper.

1. Pickands Mather & Co., Chicago.

SOME CONSIDERATIONS ON THE TENSILE AND TRANSVERSE STRENGTH TESTING OF SHELL MOLD AND CORE SANDS

By

P. J. Ahearn*, F. Quigley*, J. I. Bluhm** and J. F. Wallace***

ABSTRACT

Photoelastic evidence indicates excessive non-uniformity of stress distribution both across the critical section and at the gripping points in the standard tensile test briquet currently used for baked core and cured shell mold sand. This non-uniformity and consequent stress concentration is a direct result of the geometry of the specimen and of the relative radii of curvature at the point of contact of the grips and the specimen.

In an attempt to retain the currently standard specimen steel clips were designed to fit around the end of this specimen; it was expected that these clips would minimize the tendency of the grips to crush and cut into the sides of the specimen producing low and erratic strength values. These expectations were only partially realized. Somewhat higher tensile strengths and slightly better reproducibility were attained in the thin shell mold specimen but little improvement in testing performance was obtained for thicker core specimens.

An alternate design of briquet is demonstrated in which the stress distribution across the critical section is considerably more uniform. Tests conducted on cured shell mold and baked core mold briquets produced to this alternate design demonstrate higher tensile strengths with no significant loss of reproducibility over the standard specimen.

Large radius (5-in.) supports were manufactured for load application in the transverse testing of shell mold and core specimens to reduce the stress concentration and cutting into the specimens at these locations.

Utilization of these large radius support points produces higher modulus of rupture readings on the thin dumped, shell mold transverse test but has little influence on the results of the standard core transverse test.

Introduction

Reproducibility of baked tensile results has long been recognized as a difficult problem.[1] Some of this difficulty has been shown to be the result of variations in the absolute humidity and changes in the temperature and atmospheric conditions in the baking oven.[2] Differences in the core temperature during testing and the cooling rate and atmosphere can also introduce significant variation in the results.

In addition to these processing variables, it appears that some mechanical features of the test may influence the results. The lack of reproducibility became evident to the authors in tensile testing briquets of resin-bonded, shell-molded sand. In attempting to determine the cause of the erratic results obtained, it was observed that the literature reports the non-uniform stress distribution existing in this type of tensile specimen. The conventional type of "dog-bone" briquet[1] employed for tensile testing of baked core sand is similar to the tensile briquets frequently employed for testing cement.

Coker and Filon[3] point out, in discussing a cement tensile briquet of similar shape, that an intense contact pressure occurs in the small areas of contact where gripping takes place and that an extremely variable longitudinal tensile stress accompanied by a variable cross stress exists in the narrow central section. Not only will this variable stress produce results that are not indicative of the true strength but also lack of reproducibility will result because of the inhomogeneous nature of bonded sand.

Since the tensile strength of baked core oil or resin-bonded sand can also be measured by a transverse bend test,[1] employment of this test was also investigated. It was again found that considerable scatter in results was experienced. A review of the pertinent literature[4] indicates that some of these irregularities could be attributed to stress concentration, which occurred under the knife edge or sharp cornered supports that are employed for some tests.

In view of these difficulties with reproducibility, it was decided to undertake an investigation directed toward developing tensile tests for baked oil or resin-bonded sands that would more nearly comply with the stress analysis requirements for accurate testing. Both the briquet tensile test and transverse bend test were investigated for these materials. The work on each type of test is described separately in the following pages. This paper is solely concerned with a mechanical analysis of the tests and makes no attempt to investigate the existing and important problems of duplicate processing conditions for specimens or interpretation of actual casting performance from the results.

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Fig. 1 — Photoelastic analysis of present standard sand tensile test briquet.

Briquet Test Development

Procedure. Stress conditions existing in the standard briquet specimen were first investigated photoelastically. A photoelastic specimen was prepared from allymer CR-39 to the same face dimensions as the standard core tensile test briquet. This specimen was, however, only 3/8 in. thick compared to the standard 1-in. thickness. The result of a conventional photoelastic analysis of this test under a load applied from standard grips[1] is shown in Fig. 1.

It was evident from a study of this photograph that a considerable concentration of stress occurs at the contact points of the grips and that a non-uniform stress distribution exists across the 1-in. central section. While this variation in stress condition is probably not as severe in the actual sand briquet because the material is in the semi-elastic condition, rather than complete elastic, variations in stress of considerable magnitude undoubtedly do exist in the sand test.

Two different methods were experimentally employed to obtain a more optimum stress distribution in this tensile specimen. First, a clip of hardened steel was placed around each end of the specimen so that the grips would bear upon the outside surface of this clip rather than cut directly into the side of the sand briquet. Second, a new tensile briquet was designed such that the resulting stress concentration did not significantly influence the tensile strength.

The first method required the fabrication of hardened steel sheet, 0.031 in. thick into the contour of one-half of the specimen and a width corresponding to the full thickness of the briquet as shown in Fig. 2. The new tensile briquet was developed from a consideration of the stress pattern obtained in the specimen under load. The dimensions of this new tensile briquet are shown in Fig. 3 and the improved stress pattern obtained under tensile load is demonstrated photoelastically for a 3/8-in. thick allymer CR-39 model in Fig. 4.

These two improved methods of determining the tensile strength of the baked sand briquet were compared with the standard briquet for the baked oil-bonded core sand and for cured dry-resin-bonded shell mold sand. The baked oil-bonded core specimens were produced from a standard mixture consisting of:

90 lb washed, dried, and screened sand of AFS GFN 80; 14½ oz of cereal; 4 oz of wood flour; 1⅛ qt of heavy core oil; 3 qt of water.

The dry resin-bonded mixtures were as follows:

93½ to 97 per cent was washed, screened and dried sand of AFS GFN 140.

3 to 6½ per cent commercial thermal setting shell molding resin. Standard[1] mixing procedures in a laboratory muller were employed.

The usual type, oil-bonded, core specimens were rammed according to the established[1] procedure; the new core briquet was rammed by a contour rammer to the same density as obtained by ramming the standard specimen three times. The number of rams re-

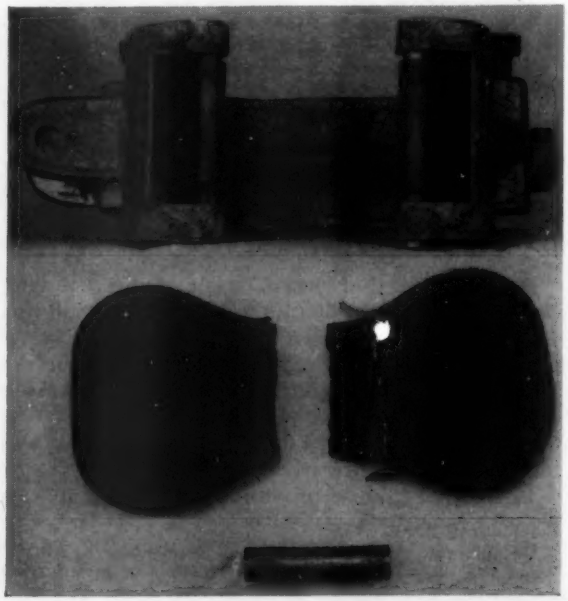


Fig. 2 — Clips and ¼-in. diameter specimen grips used with clips.

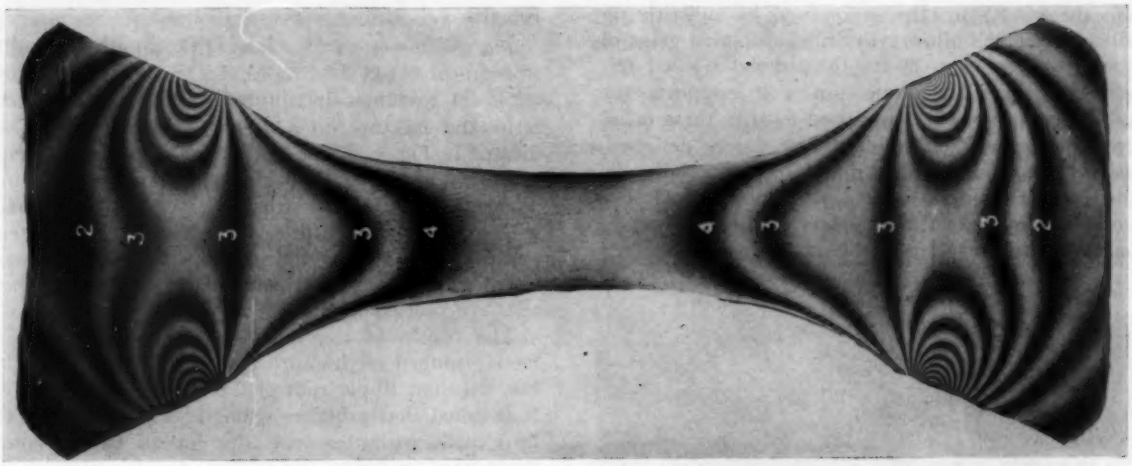


Fig. 4 — Photoelastic analysis of newly designed sand tensile test briquet.

quired for the new specimen was a function of the weight of the rammer and the friction developed between the rammer and core box. Approximately five rams were required to obtain the desired density with the equipment employed in this investigation.

The resin-bonded shell mold specimens were prepared by two methods; first, by ramming to either ½-in. and 1-in. thickness and equivalent density as previously described for the core samples, or second, by dropping the dry sand resin mixture from a dump box a height of 4 in. onto a specially machined ¼ in. high matchplate preheated to 450 F. The matchplate utilized is shown in Fig. 5. The dumping method is similar to that described in previous work in this field.[5] It is appreciated that this method of shell test dumping is at variance with the tentative standard[6] and is not completely satisfactory as discussed later.

The rammed specimens were then baked or cured in a closely controlled, laboratory, recirculating core oven for the same time-temperature cycle for each batch. The dumped, shell mold specimens were cured on the matchplate for a standard time-temperature cycle before stripping. The details of the baking and curing cycle are described with the results of each type. While dumping methods provided satisfactory ¼-in. thick shell mold briquets of standard face dimensions, some difficulty was experienced in obtaining a complete fill in the new type specimens by this method. For this reason, both dumped and rammed specimens were employed to study the tensile testing of resin-bonded sand mixtures.

The strength tests were conducted in a dead-weight type of strength test apparatus using the standard tensile test grips and procedure insofar as possible. The standard procedure had to be altered slightly

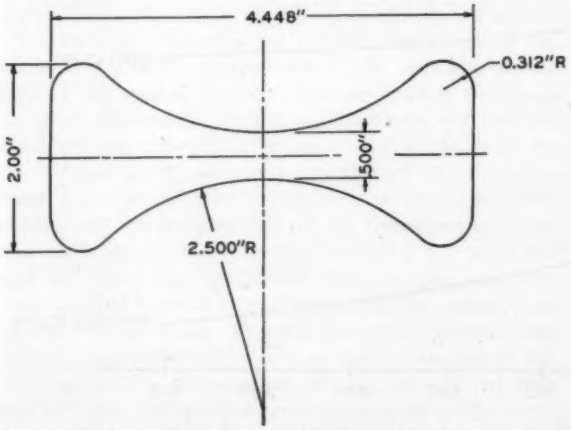


Fig. 3 — Dimensions of newly designed sand tensile test briquet developed at Watertown Arsenal.

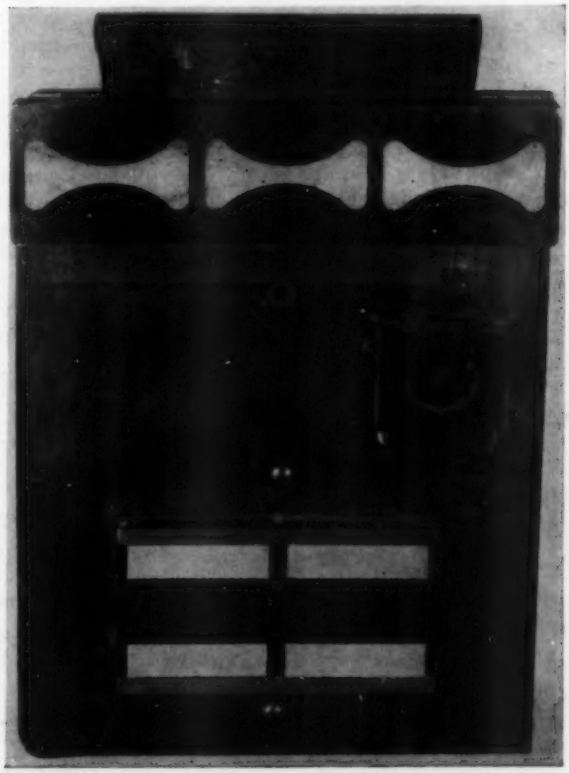


Fig. 5 — Matchplate used for manufacture of new type tensile test briquet and transverse test for shell mold.

when the steel side clips were used by substituting smaller diameter hollow cylindrical specimen grips of ¼-in. diameter (Fig. 2) for the usual flat-sided test specimen grips. The same pin that supports the standard gripping pieces was used to affix these cylinders to the specimen holder. The newly designed, tensile briquet required the use of a specially designed set of holders to grip this specimen closely. The arrangement of these holders in the dead-weight strength testing machine is shown in Fig. 6.

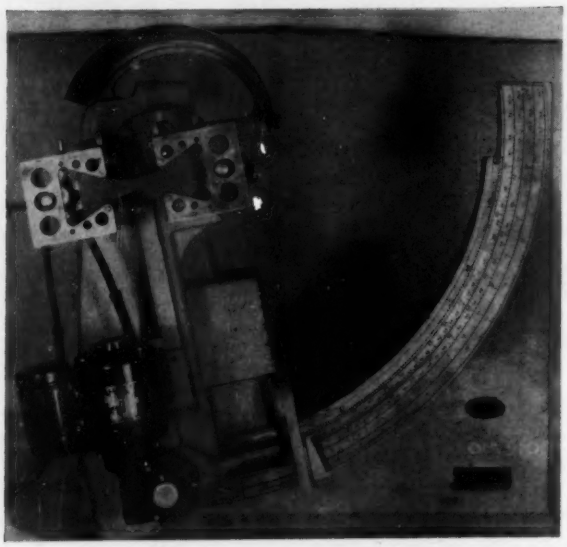


Fig. 6 — Arrangement of newly designed tensile briquet in dead weight testing machine.

Results

The influence of the side clips upon the tensile strength of cured shell mold specimens manufactured to ¼-in. thickness by dumping a dry 6.5 per cent resin-sand mixture on a preheated matchplate is indicated in Fig. 7. These results are for standard shape specimens tested in the dead-weight type of machine. The curing time was varied from 40 to 600 seconds with a plate temperature of 450 F and an oven temperature of 525 F. A total of about 15 specimens were tested at each curing time and an average of these tests are plotted in Fig. 7.

The results of any specimens which appeared defective judged on fracture appearance were discarded but less than 10 per cent of the total were eliminated. It is noted that a higher strength is exhibited by the tests made with the steel clips for all baking conditions from under to over baked. Examination of the fractured specimens indicated that a considerable number of the specimens tested without clips broke through the contact point of the grips but all of the clip tested specimens fractured through the central section of the test.

The investigation of the tensile strength of ¼-in. thick, dumped shell mold specimens obtained with various tests was continued for dry 5 per cent resin-sand mixtures utilizing the standard shape briquet tested with and without clips and briquet produced to the new dimensions. This work was confined to a large number of tests produced with a single curing treatment. (A curing time of 3 minutes at 450 F was employed). The results of specimens which contained obvious defects were discarded. Little difficulty was experienced in obtaining satisfactory specimens of

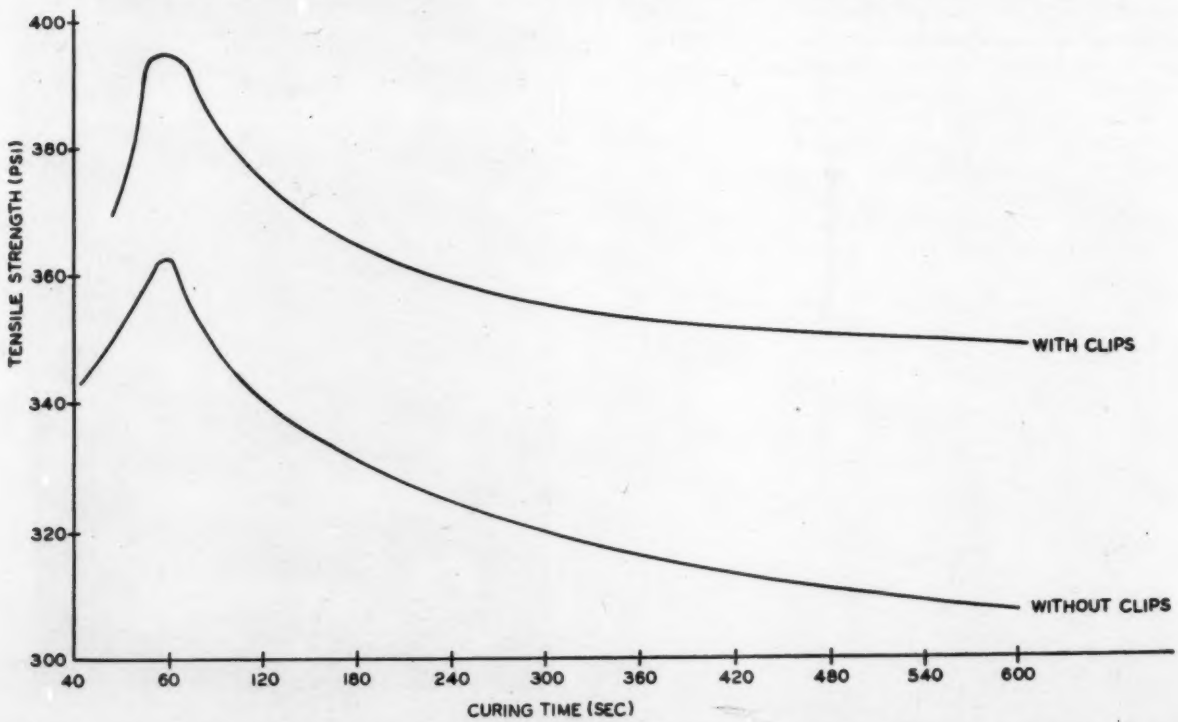


Fig. 7 — Effect of clips on strength of standard shell mold dog-bone tensile specimens.

the standard dimensions but approximately one-half of those produced to the new shape were defective and discarded because of improper filling of the ½-in. narrow section during dumping. The results of these tests are shown in Table 1.

TABLE 1 — RESULTS OF DUMPED SHELL MOLD TENSILE BRIQUETS

Type of Test	No. of Tests	$\bar{x}$-Average Tensile Strength, psi	Standard Deviation, θ psi	per cent Deviation, % θ
Standard shape without clips	101	352	36.5	10.38
Standard shape with clips	97	372	32.17	8.63
New specimen (No clips)	50	516	59.00	11.44

These results again indicate that a somewhat higher tensile strength is obtained when the clips are employed on the standard shape specimens. It is also evident that a considerably higher tensile strength is indicated by the new type of specimen. This is undoubtedly the result of the more uniform stress distribution obtained across the narrow width of this briquet.

The high strength results may also be accompanied by a reduction in scatter, provided the scatter is attributable to defects in the sand and that the spacing of these defects is large compared to the critical dimension of the specimen. On the other hand, if the spacing of these defects is minute then, neither the use of clips nor the modified specimen can be expected to significantly reduce the scatter.

In order to evaluate the applicability of this concept, the average results in Table 1 were subjected to the statistical analysis indicated below. The results demonstrate that the clips have a minor effect in reducing the per cent variation of the standard deviation. The considerable number of new type briquets discarded before testing indicate that further investigation of methods of producing dump-type specimens is necessary.

The tensile briquet results were treated by established[7] statistical methods. The arithmetic average of all results is expressed by $\bar{x}$, and the standard deviation, θ, is the square root of the average of the sum of the squares of the difference of each test from $\bar{x}$. The percent deviation, θ, compares the test results on the same relative scale and is obtained by dividing θ by $\bar{x}$ and multiplying by 100. Statistics have been previously employed[8] to evaluate sand testing and are frequently used for foundry problems.

Tensile strengths were also obtained for standard and modified test specimens using 3 per cent dry resin-bonded sand briquets compacted by ramming as opposed to dumping. The reduced resin content lowered the strength sufficiently to permit testing of all briquets within the capacity of the dead-weight testing machine. All specimens were rammed to the same density and were cured for 25 minutes at 450 F. The standard briquets were rammed to ½-in. thickness and the new tensile tests to 1-in. thickness. In this manner, it was possible to cure the same ½-sq in. cross section at the reduced section of each briquet. The selection of a thickness of briquet for each test to assure that the highly stressed section has the same cross-sectional area is necessary for both resin- and oil-bonded specimens and to avoid a variation in the curing or baking treatment.

The results obtained from these standard and new type tensile briquets are listed in Table 2.

TABLE 2 — RESULTS OF RAMMED, RESIN-BONDED TENSILE BRIQUETS

Type of Test	No. of Tests	$\bar{x}$-Average Tensile Strength, psi	θ-Standard Deviation	Per cent Deviation, % θ
Standard Briquet ½ in. thick, 1 in. wide No clips	28	343	61.56	17.9
New Briquet 1 in. thick, ½ in. wide No clips	22	644	65.99	10.2

The average of these results show the higher strength obtained in the newly designed briquet because of uniform stress in the reduced section. Statistical evaluation of the results demonstrates by comparison of the % θ, that the new type briquet has considerably better reproducibility than the standard test. Manufacturing difficulties experienced with the shell mold tensile briquets were not encountered with rammed specimens and the better reproducibility of the new type of test became evident.

Results of the tensile tests on baked core specimens indicate considerable improvement in the tensile strength measurement from the use of the new type of briquet compared to the standard type. The average tensile strength obtained with conventional tensile briquets ½-in. thick, 1 in. wide and new type briquets 1 in.-thick, ½-in. wide prepared from the same batch of core sand and baked in intermixed lots for 60 minutes at 450 F is shown in Table 3.

TABLE 3 — RESULTS OF RAMMED, BAKED OIL-BONDED TENSILE TEST BRIQUETS

Type of Test	No. of Tests	Average Tensile Strength, psi, $\bar{x}$	Standard Deviation, θ, psi	Per cent Deviation, % θ
Standard Briquet ½ in. thick, 1 in. wide No clips	48	177	33.29	18.8
New Briquet 1 in. thick, ½ in. wide No clips	49	317	36.51	11.5

Considerably higher average strength is again shown by the new type specimen because of the improved stress distribution. This improved stress

distribution also yielded better reproducibility in the new type specimen as demonstrated by a comparison of the % standard deviation.

Clips were used on the ½-in. and 1 in.-thick standard briquets but failed to produce any significant difference in either tensile strength or reproducibility. A carefully controlled group of these tests were conducted by two cooperating laboratories. Practically identical strength and reproducibility were obtained by both groups. Apparently, the grips do not severely cut into these thicker specimens because of the larger contact area and no significant variations in testing conditions are produced.

It is interesting to note, however, that the use of clips will produce a considerably higher strength and better reproducibility if the specimen grips are not properly aligned. Some earlier work indicated that improved results were obtained with clips but when the grips were re-aligned, this difference disappeared. Apparently, the misalignment caused the grips to cut into the side of the core test and produce a more erratic and lower strength when the clips were not employed.

Transverse Test Development

Procedure. The investigation was continued to determine whether a reduction in the stress concentration under the loading points of the standard transverse test bar would produce results more representative of the resin- or oil-bonded material. Special loading points were developed with the standard shank and a small section of lucite machined to a 5-in. radius on the top surface set into this shank. The loading points are shown in use in Fig. 8.

Standard loading fixtures with a flat top and sharp side corners were utilized in comparison tests. All transverse testing was conducted in the dead-weight type of testing machine with a single point load applied over a total span of 6 in. in core transverse tests and 3 in. in shell mold transverse tests (Fig. 8).

Transverse testing work was conducted on both cured resin-bonded shell mold bars and baked oil-bonded core samples. The shell mold specimens were prepared by dumping a dry resin sand mixture on a preheated matchplate as shown in Fig. 5 to obtain ¼-in. thick specimens, 1 in. wide and 4 in. long. The sand-resin mixture and dumping and curing technique are similar to those described for the shell mold tensile briquets utilized to obtain the data listed in Table 1. The sand core specimens were rammed to 1-in. x 1-in. x 8-in. dimensions by standard methods.[1] The core mixtures and baking methods were the same as for the core tensile briquets previously described.

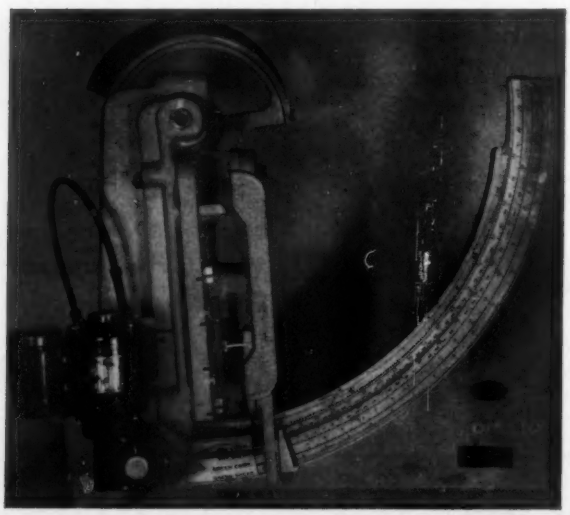


Fig. 8 — Transverse shell mold test in a dead weight testing machine with 5-in. radius loading points.

Results

The average modulus of rupture test results, standard deviation and per cent deviation obtained from cured shell mold specimens tested with standard and 5-in. radius loading pieces are shown in Table 4.

TABLE 4 — RESULTS OF DUMPED SHELL MOLDED TRANSVERSE TESTS

Type Loading Fixtures	No. of Tests	Average Modulus Rupture of in psi $\bar{x}$	Standard Deviation θ in psi	Per cent Deviation $\%\theta$
Standard	99	877	86	9.85
5-in. Radius	166	936	77	8.2

It is apparent that the 5-in. radius supports have resulted in a higher strength and a somewhat reduced scatter. These improved results must be attributed to the lower stress concentration under the 5-in. radius loading points, compared to the standard points. This reduced stress concentration obtained by large radius supports compared to sharp cornered supports has been amply demonstrated by previous work.[4] The better reproducibility is to be anticipated in this material since the resin-bonded sand contains voids that act as defects and cause erratic, low strength results when they occur in the highly stressed section of the transverse bar.

The two different types of supports were also utilized to obtain the modulus of rupture of a number of rammed and baked core sand transverse bars. The average modulus of rupture ($\bar{x}$), standard deviation (θ), and per cent deviation ($\%\theta$) for these tests are shown in Table 5.

TABLE 5 — RESULTS OF RAMMED CORE SAND TRANSVERSE TESTS

Type Supports	No. of Tests	Average Modulus of Rupture, $\bar{x}$	Standard Deviation, θ, psi	Per cent Deviation $\%\theta$
Standard	47	166	21.61	13.0
5-in. Radius	48	162	19.87	12.3

Analyses of $\bar{x}$, and $\%\theta$ results obtained for these transverse core specimens indicate that no significant difference in strength or reproducibility is obtained by the employment of the 5-in. radius supports rather than the standard supports. The fact that improved

results were obtained with the large radius supports on the shell mold bars but not on the core specimens appears to be principally associated with a difference in the material. The shell mold specimens are considerably stronger and contain fewer defects than the oil-bonded core sand. The difference may be also somewhat attributed to the fact that the span to thickness ratio is much higher for the shell mold specimen (six for the shell mold to three for the sand core).

Conclusions

1. The new design of tensile test briquet demonstrates higher tensile strength with all sands tested and better reproducibility in the case of rammed resin- and oil-bonded sands.

2. The use of steel clips at the location of the grips on the present standard tensile test briquet slightly improves the strength of dumped shell mold specimens, but does not affect the tensile testing of rammed resin- or oil-bonded sand briquets.

3. Utilization of large radius supports for load application on the transverse test slightly improves the strength and uniformity of the results of dumped shell mold sands, but has a negligible influence on the data from standard baked core transverse tests.

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DISCUSSION

Chairman: CHARLES WILLIAMSON, Columbia Malleable Iron Co., Columbia, Pa.

Co-Chairman: R. H. OLMSTED, Whitehead Bros. Co., New York.

Secretary: D. C. WILLIAMS, Ohio State University, Columbus, Ohio.

A. H. FREEDMAN[1] *(Written Discussion):* This paper on the tensile and transverse strength testing of shell mold and core sands describes a well directed and unique approach to the difficult problem of obtaining reproducibility in baked tensile specimens.

While processing variables such as baking time, baking temperature, humidity, etc., are probably responsible in part for the scatter obtained in test results, the authors have conclusively shown that the mechanical aspects of the tests may have a considerable effect on the reproducibility. This was particularly evident in the case of the rammed resin- and oil-bonded tensile test briquets.

Although the modified tests produced higher values, except in the case of rammed core sand transverse tests, these higher values per se are not important because most present day sand testing data cannot be compared in a direct quantative manner to actual foundry practice.

The ideas advanced in this paper should stimulate an approach to the design of new sand testing equipment and revision of existing equipment.

L. J. PEDICINI[2] *(Written Discussion):* The authors have, for the most part, done an excellent job of scientifically analyzing the problem of our present standard tensile specimen. There can be little argument with their photoelastic analysis and their conclusions based on statistical techniques. It can definitely be said that their new specimen shape is better for resin- and oil-bonded cores than our present specimen of ½-in. thickness. However, the authors did not indicate the correlation between a ½-in. and 1-in. thickness for the present standard shape.

Referring to past results in our own laboratory did not verify the authors' range of standard deviation or percent deviation. In fact, the last batch of 89 specimens tested from the same batch gave the following results:

No. of Tests	Average Tensile Strength, psi	Standard Deviation, psi	Per Cent Deviation, %
89	262.75	11.65	4.44

It seems possible that the notch sensitivity of the present standard specimen may be exaggerated by reducing the thickness to ½-in. However, work with the authors' recommended specimen should certainly be furthered, possibly by one of the AFS Research Committees.

R. E. MOREY[3] *(Written Discussion):* The purpose of sand testing is twofold. We may test to determine whether the sand being used in a shop is within established limits of strength and other properties, and if it is not, we may take corrective measures. This is a type of quality control. Statistical methods assist us in determining proper low and high limits for the properties and to determine when the properties fail to satisfy the established limits by an amount large enough to justify corrective action.

The second purpose of sand testing is to study the change in properties which results as a function of the change in some controllable variable. For instance, we may be interested in how compressive strength changes as a function of the water content in a sand mixture. A great many other things, of course, affect sand properties, some by large amounts, and others by relatively small amounts.

If a variable has a pronounced effect on sand properties, the general effect can usually be seen even though the testing technique is very poor. If the variable has a relatively small effect on sand properties, the actual effect may be masked by the spread of the test values unless the testing is very precise.

This is a very important point. The testing must be precise. What is precision? In a statistical sense, it is the exactness with which a measurement or test can be reproduced. Standard deviation is one measure of precision. It may be expressed in the same dimensions as the test value or it may be expressed as a percentage of the test value. This is sometimes called the coefficient of variation.

Molding and core sands may be tested in compression, in shear, in tension, or in transverse bending. If we test a relatively large number of samples by each method and determine standard deviation in percent for each method, it will usually be found that compressive tests are the most precise and tensile strengths the least precise method.

In each of the tests precision can certainly be improved by varying the techniques used and selecting techniques which produce the smaller standard deviation, δ in percent can be reduced to 5 percent, and 1 percent is not uncommon. In such a case the sand technician may be confident that his methods are good.

The authors' show values of δ in the order of 8 to 18 percent. These are so high that it seems likely the testing technique could be greatly improved without changing the speci-

1. Research Assistant, Case Institute of Technology, Cleveland, Ohio.

2. Foundry Supervisor, General Motors Corp., Process Development Section, Detroit, Mich.

3. Metallurgist, Naval Research Laboratory, Washington, D.C.

men shape. As sources of error are discovered and eliminated the standard deviation will become smaller and smaller and the test becomes a more reliable tool for the measurement of the physical properties of molding sand.

If the standard deviation cannot be reduced to a relatively low value in the tension, as well as shear and bending tests, they are useless, and we should then replace them with compression test.

In our work we have already replaced tensile, shear and bending tests with compression tests in the belief that they tell us as much or more, and give us the answers simply, quickly and reliably. We use the pendulum type machine for intermediate strengths and also have two other machines, for testing very high and very low strengths. A modified laboratory press is used for testing compression specimens from 100 to 2000 psi which covers most core and shell mold sands and a shot loaded machine is used for strengths from 0.08 to 2.0 psi which covers the green strength range of most core sands.

J. F. WALLACE *(Authors' Closure)*: The authors are very grateful to Messrs. Freedman, Pedicini, and Morey for their constructive comments. The tensile results were compared on standard specimens, ½-in. thick and new design specimens, 1 in. thick to permit a similar cross section of specimen for baking and curing. The authors agree with Mr. Pedicini that the notch sensitivity and standard deviation of the standard specimen may well have been increased by the reduced thickness. Even greater notch sensitivity and a higher standard deviation should be anticipated, then, with the ¼-in. thick shell-mold tensile tests.

The authors suggest that, in addition to the purposes of sand testing described by Mr. Morey, knowledge of the actual properties of bonded sand (including the actual tensile strength) could well have value in establishing foundry practices. The tensile tests were conducted with the usual foundry testing equipment under what was believed to be a reasonable attempt at precision. The core taken was believed to be greater than would normally be used for production sand testing and the precision is thought to be better than obtained under many operating conditions. If the present standard test specimens must be broken under rigidly controlled conditions to obtain precision, possibly a less sensitive test should be considered.

Mr. Morey is undoubtedly correct in stating that the compression tests yield greater precision than tensile tests. This is true both because of the nature of the material and the type of test. However, there appear to be two reasons why restriction of sand testing to only compression is undesirable. First, the special equipment necessary for the higher strength, bonded sands may not be available in many foundries. Second, and most important, the tensile properties of bonded sands is of direct interest to the foundryman, since some bonded sands are subjected to tensile stresses in use. The relationship between the tensile and compression properties may well vary considerably depending on the composition and treatment of the bonded sands. Actual tensile test results are believed necessary to predict the behavior of bonded sands under many conditions.

INCENTIVES IN A SMALL JOBBING FOUNDRY

By

Charles C. Erhart, Jr., and Herbert Wagers*

The foundry in which the authors are employed is a small jobbing foundry employing approximately 65 to 70 people. We have approximately 1500 to 2000 pattern changes per month. Our daily tonnage is between 20 to 25 tons and the castings poured range in size from 1 lb to 5 tons. Machine tool castings are predominately produced as well as a complete line of roadway castings.

The types of iron produced are gray iron and nodular iron. Our equipment includes a mobile sand-slinger in a main floor where large size work is produced, a side floor with one squeezer, a 6,000-lb capacity jolt rollover, and a set of 1,500-lb capacity cope and drag machines. We have a 38-in. cupola which will melt 5 tons per hour. It is hand charged and intermittently tapped.

Need for an Incentive System

The situation in which our company found itself in 1952 when the decision was made to adopt incentives was roughly as follows: Labor costs were excessive and constantly climbing. High hourly rates were being paid, but output per man-hour was low. Casting costs were high and resulting in high prices to our customers. As a result, customers were being lost and new ones not being obtained.

Dollar volume and net income were decreasing. Prices were set on the basis of time and material cost on each piece, resulting in inequities and inaccuracies in pricing. Indirect labor costs were not readily available nor apparent. Attempts were made to control costs, but no tool was available to guide management in their efforts. In a situation such as this, some definite and positive management action seemed imperative.

All signs pointed toward an incentive system with built-in controls to guide and advise management in their decisions. However many factors had to be taken into consideration before a final decision was made. Among them was the expense involved in an actual incentive installation. Without trained personnel in our own organization it was necessary to hire consultants. This, added to already high non-productive expenses, gave many doubts as to whether an organization of our size could actually afford the initial cost, and whether the results obtained would amortize such a large investment in a reasonable period of time.

Another factor was the maintenance cost. Could we afford to increase our non-productive labor costs even more with the additional people who would have to be hired to run such a system? Another big factor was the labor view. It was not known whether labor would even approve the installation of incentives, nor whether such an installation might not foment labor trouble or upset labor-management relations.

And finally there was some doubt as to whether an incentive system could actually measure our type of work and come up with standards which would be accurate and truly reflect the amount of work involved in such a wide diversity of castings which we were making. After much thought, discussion, and consulting with firms in our line of business who already had an incentive system, the decision was made to hire consultants and attempt a measurement.

Our labor standards department originally consisted of three men. One was a time-study man, another an office clerk, and the third man was a shop checker. The function of the time-study man was to take time studies, build new data, work out new methods, control labor costs, and to supervise the overall department. The clerk's job was to compute specifications and standards and to take care of premium calculations. The shop checker's job was to write all specifications and to set standards, as well as to approve any checkouts or downtime for the workmen, and to keep a record of the work each man accomplished.

As the number of patterns coming out of storage upon which standards had not been set decreased, it was found that by letting each man in the shop make out his own time sheet, we could function with but two men in the standards department. Thus we have no shop checker but depend upon each individual to keep a record of the work which he accomplishes each day, and to keep track of the amount of downtime he may have. We have found that these sheets

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are fairly accurate, and that the small amount of deviation which may occur will not compensate for the salary of a full time checker. In addition we feel that the men have more confidence in the incentives if they are allowed to have a part in reporting their own work accomplished. This does not mean however, that we do not check at all.

Anytime that unusually high premium earnings occur, a check is made for inaccuracies, but it is surprising the small number of errors which do occur. We admit that some errors occur, that some actual cheating may occur, that it is not the most ideal of situations, but in the light of our own peculiar circumstances, we feel that any savings we might make with a full time checker would more than be offset by the salary he would be paid.

During the past three years we would estimate that we have set in excess of one hundred thousand separate standards. Considering the number of standards set and dividing them by our salaries will give an idea of the dollar cost per standard. However, even this figure would probably be misleading because of the fact that many more contributions are being made to the good and betterment of the company by the standards department besides the function of setting standards. So on this basis we would consider that the cost of an individual standard is very low if one will weigh the different contributions being made by the standards people.

Experiences With Incentive System

We would now like to analyze and describe a few of our own experiences and results in the past three years.

First, our direct and indirect labor costs have declined approximately 20 to 25 per cent. This has been accomplished through increased production per man and by detecting areas which were overstaffed previously, as well as greater utilization of existing labor saving devices and the acquisition of new equipment where cost studies showed the equipment was needed.

Second, our prices are now set on the basis of standard data and directly reflect the actual amount of work in each casting, not an average cost which would cause some castings to be overpriced and some underpriced. This has led to more satisfied customers who realize they are not paying part of the cost of someone else's casting in their purchase price. By utilizing our standard data on the type of pattern equipment furnished we are able to forecast accurately and estimate our costs on a particular casting, and allow us to estimate closely casting costs on any inquiries for prices. This same standard data can also be used as a basis for a production control system to schedule accurately and balance the work flow through the shop.

Third, one of the most important features of an incentive system is the control attained. By collecting and analyzing our direct, indirect, and non-productive labor costs each week, we are able to pinpoint and correct any costs which are either excessive or out of line. Any increase or decrease in costs in a particular department can be noted at a glance and remedial action taken to correct any such condition. The trouble does not remain hidden and unsuspected for months as could happen without some means of bringing to light conditions which should be corrected.

These same figures can be used as a basis for management planning to forecast what labor costs will occur at a particular volume of operation. Then too, they should be used to determine just how much the incentive system is accomplishing in terms of dollar savings. This is in direct contrast to the practices made before incentives were adopted when only casual cost controlling was accomplished. This is understandable however for no current figures were available to spotlight sore spots and trouble areas.

Effectiveness of Supervision Gaged

A gage of supervisory effectiveness is available with the weekly report, for as the costs fluctuate, the supervisor's utilization of his men, material, and machinery is reflected. By comparing our current cost figure in a particular week with a week chosen before incentives, we can determine progress being made and remedial action necessary, as well as reward our supervisors for their part in affecting savings in their particular department. It is the view of the authors that the controls attained with an incentive system are the single most important feature of incentives.

When any incentive standard is applied, the standard assumes that a particular method, machine, or equipment will be used to accomplish the work described by the standard. Only when the method described in the standard is used, is the time allowance valid. This brings to light another aspect of incentives, standardization of methods and equipment. There are those who argue that methods should be improved and standardized before an incentive application, but an attempt of this kind will often delay the application of the standards to a prohibitive point.

Methods and Procedures Improved

Our foundry chose to apply the standards and then to attempt to improve our methods and procedures, and to adjust our standards as the amount of work in a given operation changed. Thus we were admitting that methods improvements were desirable and necessary. With our data as a guide, we then had a means of comparing alternative methods and equipment and a basis to choose the better for our particular operation. We believe that incentives have been invaluable to us in this field, if for no other reason than we became aware of the necessity for method improvements.

One operation where we have been particularly pleased with our methods improvement is on our large jolt rollover, where the flask sizes range from 38 x 38 in. to 84 x 40 in. As an example, when incentives were first applied, three men were operating this machine, hand shoveling all sand, with a daily output of 12 to 13, 48 x 48 in. green sand molds. After time studying this operation, we found that the area needing most improvement was the hand shoveling of the sand.

We considered running overhead sand to this machine, determined the cost of the equipment, and estimated what our savings would be by using overhead sand. The estimates were promising and showed that the installation would pay for itself within a reasonable period of time. However, the only element which we were eliminating was hand shoveling of the heap sand. The elements of bumping, and butting off were still present and were a large part of the work involved in the operation. We then took a look at our mobile sandslinger, and after time studying this operation found that the sandslinger was only being used a small percentage of the day and that ample time was available to ram molds produced on this machine.

By using the sandslinger to ram molds on this machine, we could then eliminate the element of shovel heap sand, bump, and butt off without any additional investment in new equipment, merely by more fully utilizing present equipment on hand. As a result of this change, the man-power used on this machine was reduced from three men to one man. The output per day rose from the 12 to 13 molds with the old method to 20 plus molds with the new method. In addition, mold quality was improved as we got a more uniform ram with our sandslinger, and the amount of scrap was materially reduced.

The success of this operation caused us to look at an adjoining and similar operation where two men were operating a set of cope and drag machines with bumpers and strippers, and like the jolt-rollover, all sand was being shoveled by hand. The flask sizes range from 28 x 28 in. to 34 x 30 in. We reasoned that if successful ramming with our sandslinger on the jolt-rollover could be accomplished, then it could also be done on this other operation with smaller green-sand molds. This too called for a one-man operation and the same proportionate increase in production resulted.

As the men became more proficient in this new method, however, we found that our sandmixer could not mix facing sand as fast as the men were using it. This seemed to indicate that a larger and faster sand-muller was needed to keep up with the increased demand for facing sand. However, then it occurred to us, why not mull all of our sand before it went into our sandslinger and eliminate the need for facing our green-sand molds? The machine was installed and cost estimates indicated that the device would pay for itself not only with increased production, but also by affecting savings in new sand, and give us a much better controlled sand and hence fewer scrap losses on all molds rammed by our sandslinger.

This new sand system has been in use for six months at this writing and while we do not feel that we have yet tapped all the savings possibilities, the results obtained have been highly gratifying, not only because of the savings involved but because we were able to eliminate the greater part of the back breaking work and drudgery which is inherent in shoveling operations. We feel that the more of this type of work eliminated, the better a foundry will be as a place to work.

Standards are particularly invaluable to the small jobbing foundry, because often times the records maintained and the information gathered by the standards department would not otherwise be available nor kept on file. For instance, we maintain a complete description of all castings made, whether one casting is made from the pattern described or 100. At a later date when the pattern is likely to be reused, we still have data concerning pattern description, flask size, casting weight, number of cores, and any other particular information concerning the casting to be made. The files of the standards department become a store-house of pertinent information readily tapped and available for instant use.

Summary

It is the view of the top executives of our company that without incentives and the controls attained, and the savings made, that the competitive position of our company might be entirely different. It is felt that incentives have been an invaluable tool in not only reducing our operating costs but also in taking the guesswork out of estimating costs for quotation purposes. It has taken the guesswork out of the evaluation of an operator's output as well as the supervisor's accomplishments. It has removed much of the arbitrary judgment from our plant and replaced it with scientific management, a solid and substantial basis for management decisions and planning.

With competition steadily getting keener it behooves every management to make their organization as efficient as possible, and to get the maximum return out of every dollar spent for plant operation. Incentives are one of the most valuable tools which management can choose to gain these ends.

Since incentives were adopted in our own plant, labor costs have been substantially reduced in each department. While we have not reached the point of maximum savings possibilities, at least we have made a good start. We know how our labor dollars are being spent, we have a goal for which to strive, and we can chart our progress in attempting to reach our goal of maximum savings possibilities. It is in such a manner that costs can be reduced.

SOME ASPECTS OF DUST SUPPRESSION IN FOUNDRIES

By

C. M. Stoch*

ABSTRACT

The various methods of dust suppression, which fall into two main groups, namely dust prevention and dust control, are outlined and briefly reviewed. A summary is given of the research and development projects in the industrial health field which have been included in the program of the British Steel Castings Research Association and which have been carried out with financial support from the British Steel Founders' Association.

There follows a comprehensive account of the Association's development work on local exhaust systems applied to grinding machines. Firstly, modifications to ensure improved dust control on pedestal grinding machines are described. Secondly, dust flow characteristics of swing frame grinding machines are discussed and various exhaust systems are critically examined as a preliminary to experimental work on the integral exhaust system, and on a combined integral and booth system.

Introduction

It has been recognized for a long time, that dust generated during foundry operations constitutes a health hazard to the personnel inhaling it and that it gives objectionable working conditions. What is more important, however, is that definite steps have been taken in order to reduce this health hazard in foundries, and a considerable amount of money has been spent by the industry, not only on introducing improvements but also on supporting development and research work in this field.

A great deal of work still has to be done before the problems involved are fully understood and solved, in particular, regarding the question of assessment of the health hazard in relation to the working conditions in foundries.

In order to answer this question the close cooperation of several different professions is required. The medical evidence of the effect of dust on human beings must be tied up with the conditions in which the men work and with the knowledge of how to alter or adjust these conditions, which lies within the scope of foundrymen, design and ventilating engineers, metallurgists and other scientific workers. The assessment of the danger of pneumoconiosis is obtained by measurements of dust concentrations, which have been carried out in various industries for many years. In South African gold mines, for example, they have been made regularly for about 40 years. In British coal mines regular dust sampling on a national scale has been carried out for several years. In foundries dust surveys have been made by various investigators but no widely co-ordinated and organized sampling has been introduced.[1]

Problems Involved and Research Projects

The considerations outlined in this paper apply mainly to steel foundry problems, with which the author is connected. The majority of these problems are common to both iron and steel foundries, with variations in the degree of their importance. Some of them appear to be more severe when making steel castings, due to higher pouring temperature, higher contraction figures, greater toughness of metal, which in turn cause extra defects, such as sand adherence, tearing, etc., and require more drastic processes of cleaning and fettling of steel castings when compared with iron castings.

In order to provide a clear picture of the terminology used and at the same time to present, in a simple form, the methods of dealing with dust, Table 1 has been drawn up.

It would be beyond the scope of this paper to consider in detail all dust suppression methods that are available, developed or being developed at present. To give a comprehensive survey, however, a brief reference is made to most of them and only the projects on which the B.S.C.R.A. have carried out research or development work are presented in more detail. The industrial health projects, which have been included in the program of investigation of the B.S.C.R.A. are given in Table 2. Regarding the priority to be given to the projects, the Association is guided by Advisory Committees representing the steel foundry industry, and a close contact is maintained with other research organizations in order to avoid, as far as possible, duplication of effort.

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TABLE 1 — SUMMARY OF DUST SUPPRESSION METHODS

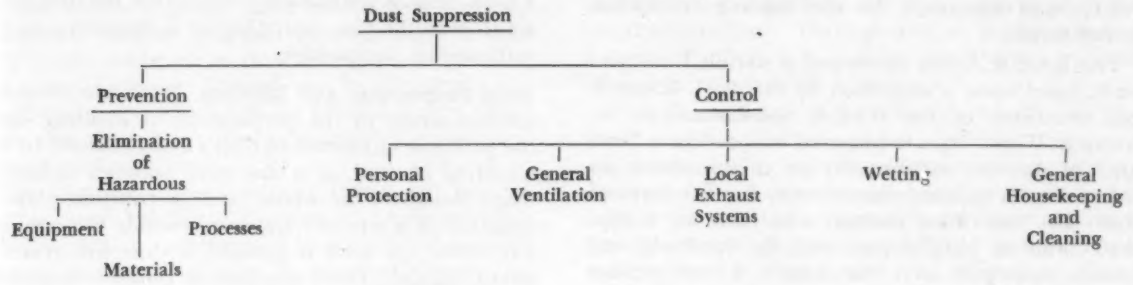


TABLE 2 — INDUSTRIAL HEALTH SUBJECTS INCLUDED IN THE INVESTIGATION PROGRAM OF B.S.C.R.A.

Research Field	Research Objectives
Suppression of Dust	Prevention of the adherence of mold materials to the surface of steel castings. Limitation in use of pneumatic chisel by introducing "hot fettling" methods (powder washing, air-carbon-arc torch). Limitation in use of siliceous materials. Development of local exhaust systems on pedestal and swing frame grinding machines. Development of "air-feed" mask. Study of steel foundry dust collection systems.
Dust Sampling and Estimation	Development of continuous dust recording equipment and dust sampling device for taking samples for nephelometric assessment. Development of automatic particle counting apparatus employing "flying-spot" microscope.
Properties of steel foundry dusts	Fundamental research on the toxicity of silica dust.

Dust Prevention

The prevention of dust formation is seldom possible, although it provides the best solution to the problem. The elimination of hazardous materials means in fact the elimination of materials producing *silica* dust, and it might be more correct to refer to the prevention or elimination of silica dust and not dust in general.

Several examples can be mentioned of minimizing the generation of silica dust in the foundry industry. Hot fettling methods, such as the powder washing process and the air-carbon-arc torch, have in many cases replaced or at least reduced the use of grinding operations and pneumatic chisels, which are recognized as dangerous dust producers. The development work carried out by the Association relating to the powder washing process has contributed to improvements facilitating the regulation of powder flow and maintenance of the equipment and resulted in wider application of this process in the industry.

Regarding the air-carbon-arc torch, the Association has been successful in assisting in the introduction of this kind of tool into its member foundries by developing and demonstrating to members a satisfactory operating technique. There are many of these torches in use and there is general agreement that in addition to the reduced health hazard they are superior in speed to the pneumatic chisel.

The elimination of hazardous materials includes the prohibition of sand in shot-blasting and of the use of silica flour as a parting powder. Some steel founders went so far as eliminating silica flour from their shops altogether and substituted for it in mold paints and fillers non-siliceous materials such as zircon, silimanite, etc. Other steel foundries, however, continue to use silica flour for mold paint and filler, justifying it by the superior strip and surface finish of castings, which results in reducing the dust problem in the cleaning room. In many cases silica flour is delivered and handled in the foundry in a wetted state.

In connection with the preventive measures leading to a reduced amount of sand adhering on castings, which has to be removed by dressing, the research at Cambridge University on the causes of sand adherence, which is sponsored by the B.S.C.R.A. has given valuable information and will no doubt contribute to the solution of dust problems. This investigation has elucidated many of the factors governing metal penetration and sand "burn-on" both of which are related closely to the degree of sand compaction. The latter problem is now under active investigation with the object of improving the sand compaction in machine molding.

Dust Control

The various methods of dust control outlined in Table 1 are not necessarily arranged in order of their importance or effectiveness. Their application will depend on many factors such as regulations, conditions of operations, adaptability to these conditions, capital available, running cost, etc.

It should be stressed at this point that the majority of steel foundries in Great Britain are jobbing foundries with varied degrees of mechanization and this must be kept in mind when considering the possible application of dust suppression methods.

Personal Protection from Dust

This is not in fact a method of dust control as protective masks and respirators have to be used because dust is not controlled by other means. This is clearly indicated by the regulations, which require that the occupier shall provide approved respirators for workers carrying out any operations creating a

heavy dust concentration which cannot be dispelled quickly and effectively by the existing ventilation arrangements.

The B.S.C.R.A. has developed a so-called *air feed mask*, based upon a suggestion by Mr. R. J. Richardson, Chairman of the B.S.F.A. Sub-Committee for Practical Work. This is intended to provide a "curtain" of dust-free conditioned air to the wearer in such a way as to avoid the necessity for any form of filter. The nose hood through which the air is supplied forms an integral part with the eye-shield and extends downward over the mouth. Extensive tests have shown that a suitably designed air feed mask can be made to provide a high degree of protection against the inhalation of airborne dust, whether breathing is by mouth or by nose.

General Ventilation

It is not always possible to control dust and fumes at the points of their generation, in particular in jobbing and heavy foundries. The whole atmosphere of the foundry consequently requires some general ventilation in order to avoid excessive build-up of the dust level and contamination.

General ventilation systems very often rely on roof ventilators. The air flow through the ventilators depends upon the direction of the wind and the temperature difference between the inside and outside atmospheres. These factors vary between wide limits, and make ventilation unreliable and changeable. Under unfavorable conditions there might even be downdrafts from the ventilators in place of upward ventilation.

A recent research carried out by the Building Research Station of the Department of Scientific and Industrial Research in a floor molding foundry gave interesting observations.[2] It was shown that dust and fumes instead of going to the ventilating fans provided in the roof span, travelled along the building at a height of about two-thirds of the distance from floor to roof and partially dispersed through a doorway. The conclusion from these observations was that dust-bearing currents will not necessarily be controlled by adding height to the building and that the height of a foundry should be based on cranage requirements rather than ventilation. This would lower the initial cost of the building in many cases.

There are foundries ventilated on the principle of downward movement of ventilating air, which is introduced from above by means of overhead trunks and is drawn into the local exhaust units placed near the floor level, such as knock-outs, pouring stations, etc.[3] If applicable, this system provides a more logical way of dust control than upward ventilation, in which all airborne dust has to pass the operator's breathing level on its way to the ventilators.

Local Exhaust Ventilation

Fine dust, of respirable size, is dispersed from the point of its generation by air currents which may be extraneous or may be produced during the operation. In some cases, such as grinding, where fine dust and large particles are produced, the latter induce some of the fine dust to follow their path. The function of a local exhaust ventilation is to capture the dust-laden air close to the dust source before it disperses into the surrounding atmosphere.

Sand Preparation and Molding. With the increased mechanization in the preparation of molding sand, the problem of control of dust clouds released in the handling of sand in a dry state becomes easier. In many foundries the whole sand preparation plant is installed in a separate enclosure outside the molding bay. Once the sand is damped it does not create a health hazard. There are several possible sources of dust in a molding bay. Parting powders were mentioned previously.

Cleaning of molds by a jet of compressed air or using a T-piece suction pipe operated by compressed air, removes the dust from the mold and disperses it into the surrounding atmosphere. A potential source of dust here is the dry sand accumulated on the floor, which becomes airborne during normal working procedures. The right way to deal with the latter two problems is the application of exhaust systems in the form of vacuum cleaning of molds and floors, wherever possible.

Knock-out. There are local exhaust installations operating in foundries, which provide an effective control of the heavy clouds of dust and fumes normally associated with knocking out. It is obvious that the more complete is the enclosure of this operation to guide the air flow where it is needed most, the easier it is to control the dust employing lower exhaust volumes. The difficulties connected with enclosing the knock-out grid become more pronounced when dealing with larger boxes handled by a crane.

There are three main well-known systems of applying exhaust ventilation at knock-out grids: a) Upward exhaust into an overhead hood; b) side-draft into a hood placed close to the knock-out grid, and c) down-draft exhaust through the grid. *Upward exhaust* is comparatively simple to design and operate, but has the disadvantages that the operators must often bend into the dust cloud and that it cannot be used when boxes are handled by an overhead crane. Upward exhaust is often employed in completely enclosed knock-outs, which are rather limited to small boxes and used mainly in mechanized foundries.

Side-draft hoods are placed close to the knock-out grid, either without a roof extending over the grid, or hoods are inclined and provide a roof over ⅓ to ½ of the grid width. Inclined side hoods are a compromise between a straight side-draft and an up-draft and they require a much greater volume of exhaust air, higher power costs and heat losses than the up-draft. *Down-draft* system is suitable only for small boxes, not higher than 12 in., but it has the advantage that all four sides of the knock-out are accessible to the operators, and there is no hindrance to the use of a crane or hoist.

Cleaning of Castings. The cleaner the castings, that is the less sand adheres to them, when they reach the cleaning room, the lower will be the silicosis hazard for dressers. Mechanical methods of cleaning castings

by hydro-blast or metallic abrasives are well-known and are widely used in foundries. Abrasive cleaning operations, whether it is shot blasting, using a high pressure air stream or an impeller type unit utilizing centrifugal force, are carried out in completely enclosed chambers, which can be ventilated very effectively providing the exhaust equipment is maintained in good working order.

The cleaning equipment, which does not require any personnel inside the chamber, is being installed in an increasing number of foundries. In shot blast cabinets where an operator is employed inside the chamber during the cleaning operation suitable protective helmets supplied with clean air must be worn. Regulations require provision and maintenance of protective helmets for the use of all persons who are employed in a blasting chamber, whether in blasting or in any work connected therewith, or in cleaning inside the chamber.

Portable Fettling Tools. When cleaning small castings with portable grinders or pneumatic chisels, a local exhaust can be provided by suitably designed benches or cabinets connected to a suction unit.[4] A considerable volume of air must be exhausted to control the dust efficiently. Appropriate air-curtains can often be arranged to improve the performance.

Dust control on portable tools is more difficult when dressing larger castings, on which more adhering sand can be expected. Heavy dust clouds can be observed and if exhausted cabinets of suitable size were considered, large volumes of air would have to be handled. Observations have shown that this method cannot be entirely satisfactory as in most cases heavy dust clouds contaminate the operator's breathing zone before dispersing into the current of ventilating air. Experimental work is in progress exploring the possibility of using a small portable exhaust hood designed so that it could be conveniently moved by the operator as the cleaning operation progresses.

Recently a different method of approach to the problem of dust control on portable tools has been publicized and the equipment introduced on the market. This is referred to as a *"low volume high velocity exhaust system"*[5] which was first developed for percussive rock drills. Dust produced by drilling was extracted through the hollow drill steel, at a velocity of 16,000 to 18,000 feet per minute, at a static vacuum of 10 in. of mercury. When applied to a *pneumatic chisel,* the suction was brought close to the point of dust generation through a hollow chisel and through a specially designed hollow pneumatic hammer. The hammer was connected to a suction unit through a flexible hose of approximately ¼ in. in diameter, exhausting 4 cu ft/min, at a vacuum of 10 to 15 in. of mercury.

The hollow chisel had some mechanical disadvantages and the design was replaced by a conventional solid chisel on which a rubber sleeve was fitted. Ducts pass through the sleeve and extract from both back and front surfaces of the chisel. Another alternative method was developed on the basis of the so called "mason's glove," because it was used originally in the stone cutting industry. A suction hose was attached to the left-hand glove of the dresser, so that the inlet of the hose can be brought close to the point of the chisel. Through a ½ in. diameter hose a volume of 13.5 cu ft/min was exhausted at a vacuum of approximately 5 in. of mercury.[6]

The low volume high velocity system was also fitted to a 6-in. diameter *portable grinder.* The guard was designed with a double wall on its edge, forming a peripheral duct of rectangular section. On the inside surface of this duct, intake ports were provided through which approximately 24 cu ft/min of air was extracted at a vacuum of 5 in. of mercury. Good dust control was obtained, but some inconvenience was experienced in practice due to the coverage of the grinding wheel by the hood, and modifications are in progress to reduce the size of the hood.[5]

Stand or Pedestal Grinding Machines. It was reported some years ago that an orthodox exhaust system on stand or pedestal grinding machines does not provide an effective control of the dust generated during grinding.[4] Since then a great deal of experimental work and observations have been carried out and published, which lead to a closer understanding of the problems involved.

Investigations were made in the dust research station of the British Steel Castings Research Association[7] upon stand grinder machines with 24-in. wheels running at a peripheral velocity of 9000 fpm, with the object of converting the orthodox wheel guard into an efficient unit with maximum simplicity and at minimum cost. Throughout these studies, the observations of dust flow have been made employing the Tyndall beam illumination, which enables the clouds of respirable dust to be seen and photographed.

The application of this method for making cinematograph film records of the movement of dust clouds in foundry processes has been initiated on an extensive scale by W. B. Lawrie. Experimental work has shown that for effective dust control it is essential that in the first instance all the dust generated by grinding enters the machine cowling. With the work rest in its normal position, that is close to the wheel face, it deflects some of the primary dust stream into the shop atmosphere and in order to minimize this stripping effect it is necessary to perforate the work rest.

Observations have shown further that the dust which is entrained in the slip stream of the grinding wheel and which normally emerges at the top of the cowling, as a secondary dust stream, can be reduced by preventing the ingress of extraneous air into the cowling, and concentrating the extraction system upon the dust stream itself. This requirement can be met by introducing simple modifications to the machine cowling such as (Fig. 1); (a) closing the gap beneath the work rest, (b) minimizing air gap between the sides of the cowling and the wheel and (c) fitting a top flap which is easily adjustable to wheel wear.

Regarding the design of the work rest various forms of perforations were tested.[8] These included slots cut parallel or at right angles to the wheel face and round holes of various size and pitch. If the per-

forations are too small they become clogged and if too large they interfere with the movement of the casting during grinding, and it was found that holes of ¼-in. diameter at about ½-in. pitch give a satisfactory performance providing they are kept open and the point of grinding is close to the work rest.

Tests under steel foundry production conditions have shown, however, that when grinding larger castings high up the wheel, which is a common practice, a considerable decrease in the effectiveness of the dust control occurs. This is due to the fact that the area of the remaining flat surface of the work rest is large enough to deflect a substantial volume of the primary dust stream and as the grinding proceeds the holes become clogged fairly quickly.

Following these observations a new design of work rest was developed, in which the perforated top was replaced by ⅝-in. diameter steel rollers spaced so as to leave a row of parallel slots ⅛ in. wide (Fig. 2). The rollers rotate as the work is moved across the wheel and so the spaces between the rollers are self-cleaning. The rolling contact of the work with the rest assists in moving the work across the wheel and minimizes wear, since contact is distributed over the peripheries of the rollers, the total area of which is considerably greater than the area of the upper side of the rest. A roller work rest was employed on production work in a steel foundry and has shown a satisfactory performance.

The work of the British Cast Iron Research Association[9] on the dust control on pedestal grinders resulted in developing an exhaust design. The main feature of this design is described as an external exhaust, which forms an addition to the orthodox exhaust system. This additional ventilation is provided by exhausting air through two vertical slots situated on the sides of the wheel and a horizontal slot above the grinding wheel. These intake-slots are made in a flange surrounding the wheel and they lead into an exhaust box at the back of the flange.

Swing Frame Grinding Machines. In its studies of the problem of dust control on swing frame grinding

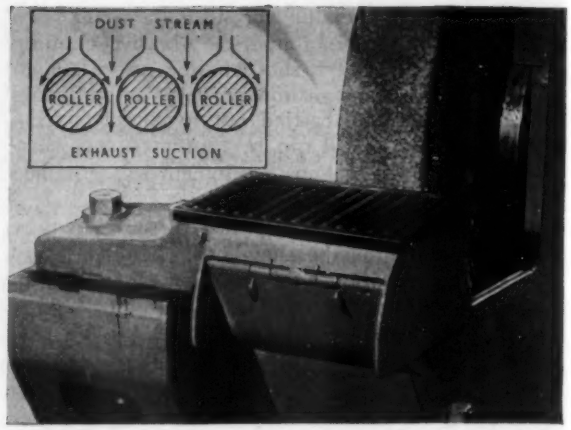


Fig. 2 — Roller type work rest on a stand grinder machine; inset shows the ideal streamline pattern of air flow.

machines, the British Steel Castings Research Association[8] carried out a series of observations both in the Dust Research Station and under industrial conditions, of the dust flow characteristics of various types and sizes of swing frame grinding machines as a preliminary to original experimental work in this field.

There are three main types of machines commonly used in the steelfounding industry: (1) "standard design" machine in which the grinding wheel is in-line with the main frame, the operator standing in front of the machine and controlling it by what are popularly known as "cow-horn" handles; (2) "astride design" machine, which is similar to the standard unit in so far as its grinding wheel is in-line with its main frame, but with the modification that the operator stands astride the wheel, the machine being controlled by a handle that is parallel to the wheel axis. The operator's back is toward the motor of the machine and his head is above the grinding wheel itself; (3) "transverse wheel design," in which the grinding wheel is mounted at right angles to the supporting beam, while the operator stands in front of

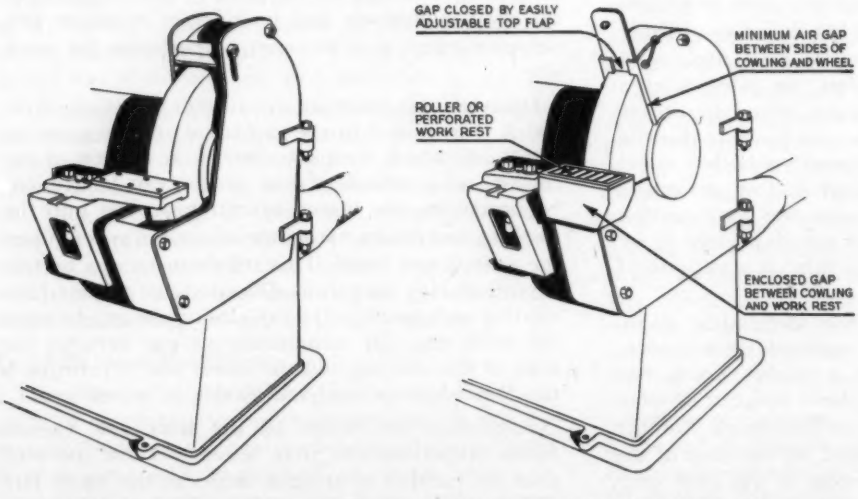


Fig. 1 — Conventional hood design for a stand grinder machine and an improved hood design showing the points at which the modifications are made.

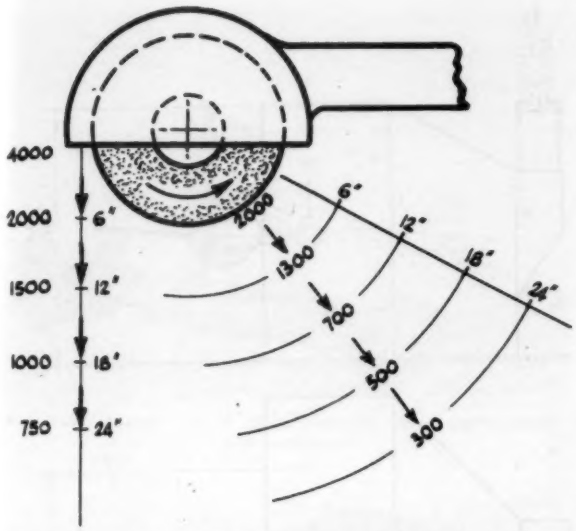


Fig. 3 — Velocities of air flow (fpm) produced by a rotating 16-in. diameter grinding wheel, with 9000 fpm peripheral velocity.

the unit in the position similar to that in the case of type (1).

These various types of swing frame grinders show the same characteristic pattern of dust flow, consisting of a primary stream of dust ejected tangentially from the point of its generation and a secondary stream of dust emerging from the wheel guard. The *primary dust stream* consists of particles of a very wide range of sizes. Large particles are of no importance from the point of view of the health hazard, but their influence on the flow of fine dust has to be considered. When grinding different materials, i.e., wood, steel castings, and rusty pig iron, it became evident that the nature of the material being ground has an influence upon the behavior of the primary dust stream.

These observations have shown that where a proportion of relatively large particles is produced by the grinding operation, these particles, due to the kinetic energy imparted to them on their formation and release, follow a straight path tangential to the wheel for a considerable distance, and that they in turn induce some of the fine dust to follow in the same path and to remain in the primary stream. The *secondary dust stream* is brought around with the slip-stream produced by the rotating grinding wheel, is discharged into the general atmosphere and may reach the operator's breathing zone. Whether it reaches the operator's breathing zone as a cloud of high concentration depends upon the position of the operator (in particular upon the position of his head), in relation to this stream of dust, and upon the contour of the casting that is being ground.

The effect of the shape or contour of the casting being ground is evident, as this stream, on leaving the hood, is directed downward, and, depending upon the position of the grinding machine relative to the casting will either continue unobstructed until it reaches the floor or will be deflected and spread either sideways and, in many cases, partially upward, depending upon the configuration of the obstruction, i.e., the casting or the bench, which it may meet. The velocities of air streams produced by a 16-in. diameter grinding wheel when rotating with a peripheral velocity of 9000 fpm are shown in Fig. 3.

The local exhaust systems on swing frame grinders can be classified as follows:—

I External exhaust systems: a) Down-draft, in which the grinding operation is conducted over or adjacent to a floor grid through which dust laden air is extracted. b) Booth (side-draft), in which the swing frame grinder operates in front of a booth that is coupled to a fan.

II Integral exhaust systems, in which the local exhaust is embodied in the swing frame grinder unit itself and a flexible pipe connects it with

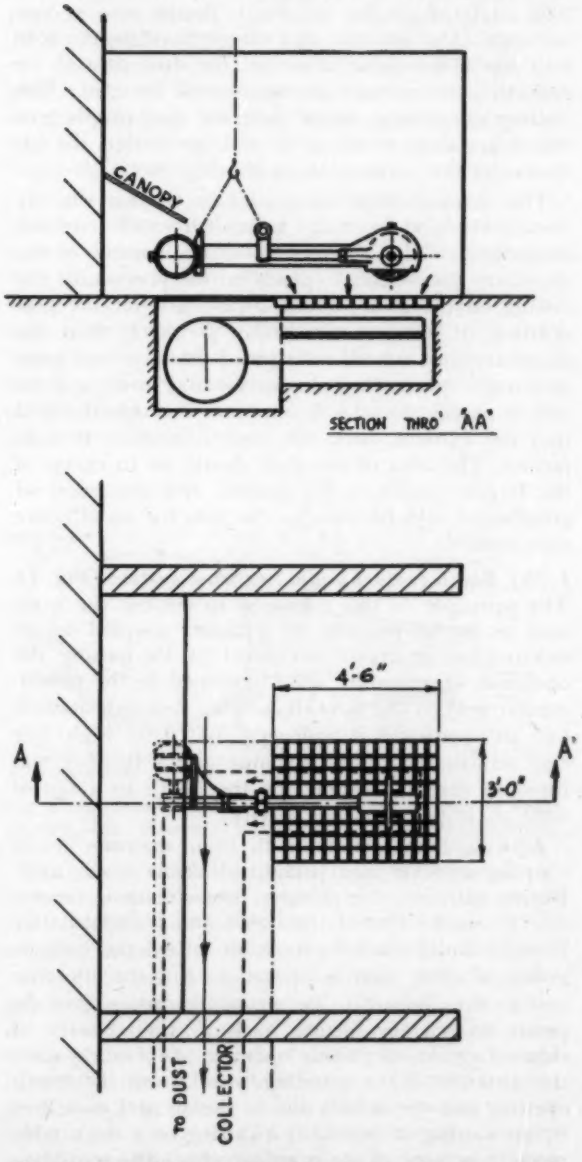


Fig. 4 — Downdraft exhaust system for a swing frame grinding machine.

an exhaust fan: a) the English Steel Corporation Ltd. exhaust system. b) The B.S.C.R.A. integral exhaust system.

III The B.S.C.R.A. combined booth and integral exhaust system.

The observations and trials conducted on each of the exhaust systems applied to swing frame grinder machines are now described.

I (a) Down-draught exhaust system (Fig. 4). Observations of the down-draught exhaust system were carried out in a steel foundry cleaning room, where this installation has been in operation for several years. The dust control is obtained by enclosing the machine and the working area on three sides and by providing local exhaust ventilation through a floor grid 4 ft 6 in. long and 3 ft wide. This grid is placed over a pit fitted with an exhaust duct through which 5000 cu ft of air per minute is drawn into a dust collector. The addition of a canopy-plate on the rear wall has a beneficial effect on the dust control by deflecting the primary stream toward the grid. Flat castings are ground mainly and the dust clouds generated are close to the grid, and are within the influence of the currents of air flowing into it.

The primary dust stream when thrown directly toward the grid openings is brought under control immediately. The effectiveness in the control of the secondary dust stream depends on the position of the casting on the grid. Should a casting cover the grid opening in front of the grinding wheel, then the secondary dust stream will spread sideways and some of it might escape the exhaust system. Providing there is free passage for the dust stream through the grid into the exhaust duct, the control of dust is satisfactory. The area of the grid should be in excess of the largest casting to be ground and the point of grinding should be close to the grid for an effective dust control.

I (b) Booth (side-draught) exhaust system (Fig. 5). The principle of this system is to enclose the work area as far as possible by a booth coupled to an exhaust fan to create a current of air passing the operator. The booth, which was used in the experimental work at the B.S.C.R.A. Dust Research Station, had an opening 5 ft wide and 5 ft 6 in. high. For easy adjustment of the exhaust air, a damper was fitted in the duct connecting the booth to a fan of 6000 cu ft per minute capacity.

A swing frame grinder with 16-in. diameter wheel running at 9000 fpm peripheral velocity was used. During grinding the primary dust stream is thrown directly to the rear of the booth and is immediately brought under control even with an exhaust volume as low as 1500 cfpm in operation. But the effectiveness in the control of the secondary dust stream depends on various factors, such as the velocity of exhaust air in the booth opening, the fan capacity, the distance of the grinding wheel from the booth opening and the obstruction to the air and dust flow by the casting or bench. If a casting or a work table projects in front of the grinding wheel, the secondary dust stream will be deflected sideways and upward.

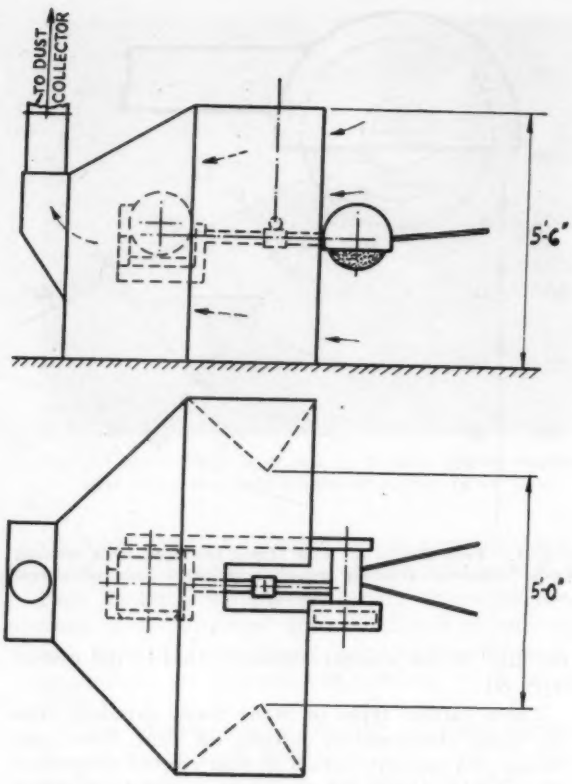


Fig. 5 — Booth (side-draft) exhaust system for a swing frame grinding machine.

It has been observed that even when grinding in the booth opening and with an exhaust rate of 6000 cfpm in operation, which gives a velocity of exhaust air in the vicinity of the grinding wheel of approximately 200 fpm, dust clouds of high concentration may reach the operator's breathing zone. If the conditions of grinding are such that the secondary dust stream travels unobstructed to the floor, its control will depend not only on the velocity of the exhaust air, but also on the distance of the grinding point above the floor and construction of work table.

Observations have shown that with the point of grinding 2 ft above the floor a considerable volume of dust was deflected by the floor and spread to a distance of several feet from the grinding position and dispersed in the general atmosphere of the shop. These conditions occurred even with an exhaust of 6000 cfpm and when the front of the work table was left open. In order to simulate the conditions when grinding a large casting standing on the floor, the front of the work table was covered. As can be expected, the dust clouds deflected away from the booth opening were more pronounced and more dust escaped into the shop atmosphere.

As a next step in experimental work the height of the work table was increased from 2 ft to 3 ft above the floor level, so that the secondary dust stream had a longer distance to travel, therefore its velocity was lower (approximately 500 fpm) when reaching the floor. An exhaust of 6000 cfpm was required to direct

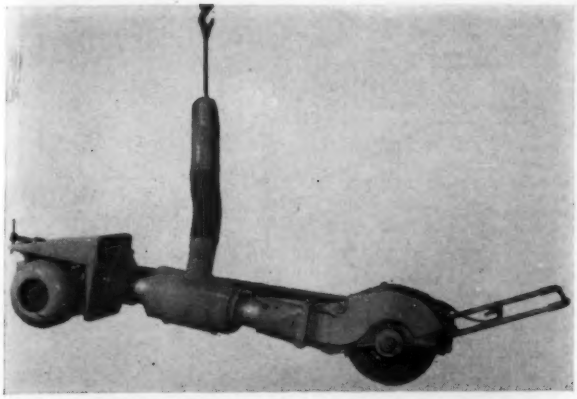


Fig. 6 — General view of 20-in. diameter swing frame grinding machine.

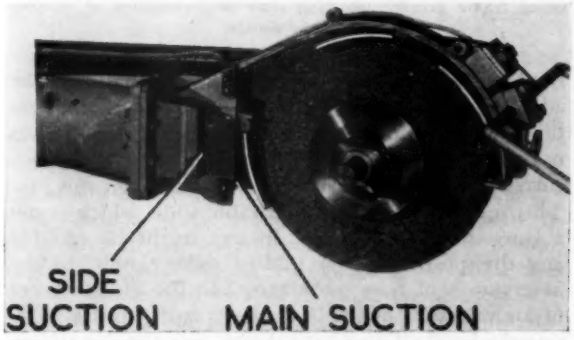


Fig. 7 — Grinding machine head with side of cowling removed to show position of baffles and exhaust intakes.

the dust stream into the booth opening before it reached the floor and to control it satisfactorily.

II (a) The English Steel Corporation Ltd. Integral exhaust system (Fig. 6 and Fig. 7). The general principle of the E.S.C. design is to control the dust generated by grinding, by exhausting it through the hollow frame or beam of the machine itself. In order to achieve this the primary dust stream is exhausted directly into an opening in the machine frame immediately to the rear of the grinding wheel. To control the secondary dust stream entrained in the peripheral slip-stream of the wheel, there are three adjustable stripping plates or baffles incorporated in the wheel guard. These baffles are coupled so as to permit simultaneous adjustment of their position in relation to the wheel periphery, by the operator.

Adjustment is facilitated by the fitting of a rattling device which comes into operation as the adjustable baffles contact the wheel periphery. The dust deflected or stripped by these adjustable baffles is then extracted in a direction parallel to the axis of the grinding wheel, into a chamber incorporated in the side of the wheel cowling. This chamber is in turn connected to the main exhaust duct formed by the hollow frame or beam of the machine. Incorporated in the machine's hollow beam at its mid-section is an expansion chamber for the purpose of allowing large dust and grinding particles to settle, while the main stream is carried upward into the flexible pipe to the dust collection unit. The expansion chamber is provided with a removable plate to permit cleaning.

Observations were carried out at the dust research station on a 20-in. diameter wheel machine with an exhaust of 900 cfpm at 7¼-in. w.g. static pressure. These trials have shown that the machine provides an effective means of controlling dust generated during grinding and of reducing to a considerable extent the contamination of the general atmosphere of the shop in which the machine is operated. The machine does not attempt to control the large particles (sparks) arising from the grinding operation.

II (b) The B.S.C.R.A. integral exhaust system.[10] The object of the B.S.C.R.A. work was to develop a device or modification that can be applied to new as well as to existing swing frame grinding machines in foundries. The experimental work was conducted employing a 'standard' machine, with grinding wheel 16-in. diameter, 2½ in. wide, driven by a 7½ hp electric motor giving a wheel velocity of 9000 peripheral ft/min. The various experimental attachments to the grinding machines were connected in each case through a 5½-in. bore flexible pipe to a paddle-blade fan, of 1500 cu ft/min capacity.

In the first stage of the experimental work an extension to the wheel guard was built in the shape of a box with two intake-openings in the underside (Fig. 8) which were provided with sliding covers to allow easy adjustment to their size. A portion of the back plate of the wheel hood was removed to produce a continuous internal cavity within the original hood and the "box" attachment. Gaps between the sides of the wheel and the hood and at the front of the hood were closed.

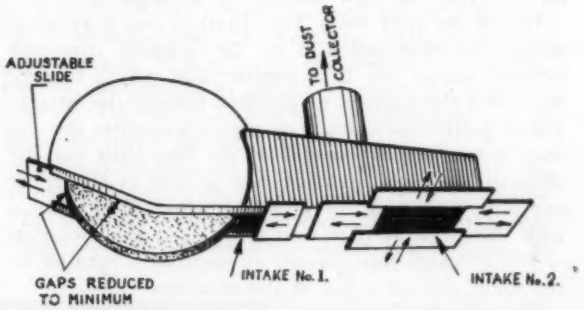


Fig. 8 — Experimental attachment for dust control on swing frame grinder. Note sliding covers on underside of attachment.

With both intakes adjusted to optimum sizes and positions, which were assessed by subsequent trials and observations, a satisfactory control of smoke while grinding wood and of fine dust while grinding steel castings or pig iron, was obtained, providing the grinding operation was carried out with little "sideways" or lateral swing of the wheel. The sideways swing of the grinding wheel during grinding resulted in moving the suction intakes from the primary dust stream and the speed of the sideways movement in relation to the velocity of the dust stream was seen to influence the decrease in the effectiveness of dust control.

It was concluded therefore that by enclosing the dust stream at the sides, this undesirable effect might be corrected and for this purpose two rubber sheet flaps were fitted, extending downward from the sides of the attachment, as shown in Fig. 9. Trials were disappointing; fine dust of the primary system was seen to escape underneath the bottom edge of flaps.

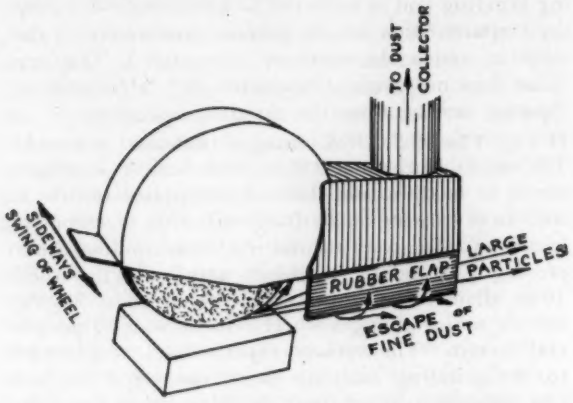


Fig. 9 — Showing position of experimental side flaps extending downward and indicating escape of fine dust which occurs while swinging the wheel sideways during grinding.

In a further attempt to correct this failure, horizontal flanges or extensions were fitted to the bottom edges of the attachment (Fig. 10) and an improvement while grinding with sideways swing was noticeable at once. By extending the horizontal flanges toward the "front" of the wheel hood the flow of air current produced by the fan effect of the flat faces of the exposed part of the grinding wheel was utilized for directing the dust toward the exhaust intakes.

It can be seen from Fig. 10 that the horizontal flange was extended also in the opposite direction beyond intake No. 2, a modification which further improved the pattern of air flow toward the intake. These preliminary stages of the experiments showed that a satisfactory control of the fine dust can be obtained with the exhaust rate as low as 600 cu ft per min, providing the grinding wheel is in contact with the work at its lowest point, i.e., the direction of the primary dust stream is essentially parallel to the main axis or beam of the machine.

If, however, the point of grinding be nearer the trailing edge of the hood, i.e. grinding is done with the 'front' part of the wheel, then the primary dust stream is directed downward and away from the exhaust intakes, the influence of the exhaust on the dust stream being thereby reduced. The position of the exhaust openings cannot be lowered without interfering with the operation of the machine and therefore, in order to extend the sphere of influence of the exhaust it is necessary to increase the volume of the exhausted air. It was assessed that with an *exhaust rate of 900 cfpm* a satisfactory degree of control of the fine dust was obtained when the primary stream was directed downward at angles of up to 45 degrees in relation to the main axis of the machine.

As shown in Fig. 8, the gap at the trailing end of

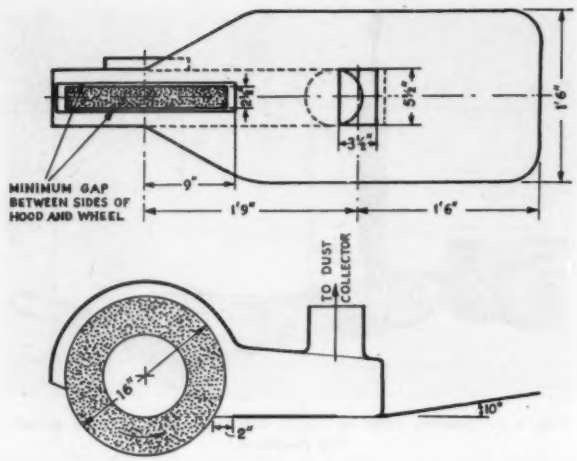


Fig. 10 — Design of exhaust hood, for a 16-in. diameter wheel swing frame grinder, showing sizes and positions of intakes and the flange.

the hood was sealed by means of a slide that could be adjusted as the wheel became worn. However, when the exhaust volume was increased to 900 cfpm, it was noticed that, if this gap was left open, there was a current of air flowing into the hood through this gap. This suggested that the adjustable slide, which is not a convenient or practical fitting, might be omitted and this possibility was studied more closely, by the observation of dust movement and the measurement of air velocities not only in the opening, but also inside the hood. These have shown that the slip stream of the wheel is overcome by the exhaust air over the whole area of the opening (Fig. 11) and a current of air flows into the trailing end of the hood with a velocity of 900 fpm close to the wheel periphery and 1700 fpm in the center of the opening.

Trials carried out with a *worn down wheel* fitted (10-in. diam – peripheral velocity 5625 fpm) revealed a satisfactory performance as demonstrated in Fig. 12, when grinding pig iron so that the dust stream was directed at an angle of about 45 degrees to the beam of the machine. With the exhaust volume of 900 cfpm the fine dust is under good control. A series of air velocity readings was taken outside the hood at various distances and on the basis of these results, the

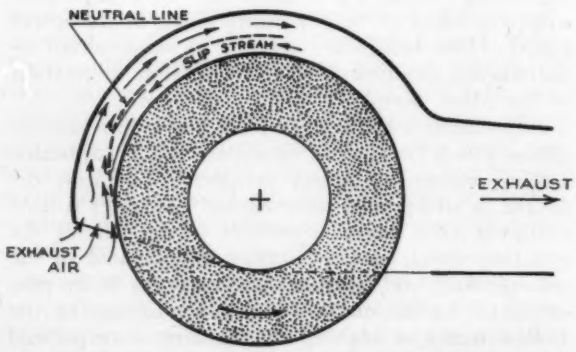


Fig. 11 — Diagram of air flow at opening and inside hood at trailing end of wheel. Exhaust 900 cfm. Peripheral velocity, 9000 fpm.

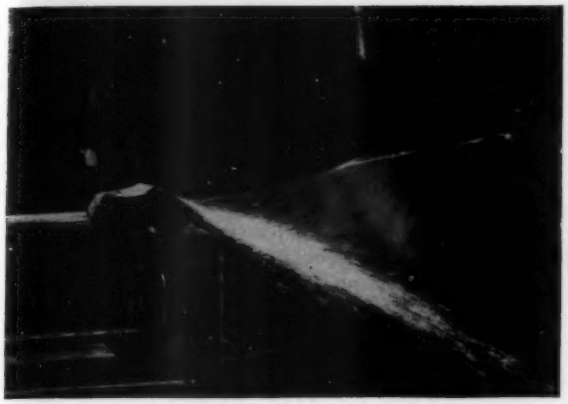


Fig. 12 — Control of dust when grinding pig iron with a worn-down wheel; exhaust, 900 cfpm. Although grinding position is such that sparks are projected at a 45-degree angle, the fine dust is brought under control.

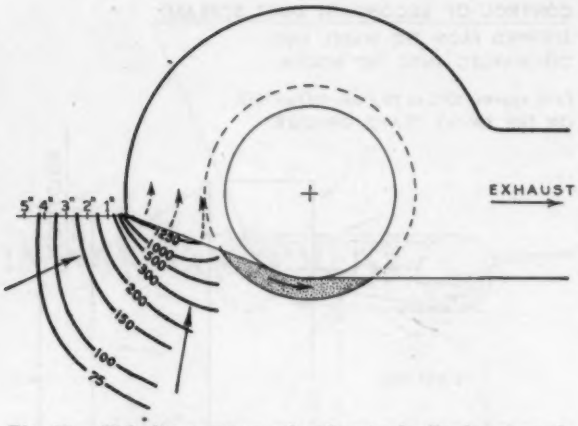


Fig. 13 — Velocity contours of exhaust air flowing toward the opening at the trailing end of hood. Wheel diameter, 10 in. Peripheral velocity, 5625 fpm.

velocity contours of the exhaust air flowing into the trailing end of the hood were drawn. From these it will be seen that air moves toward the hood opening at a velocity of 75 fpm at a distance of approximately 5 to 8 in. from the opening (Fig. 13).

The lower line of *the driving belt,* which runs toward the back of the machine, generates a current of air flowing in the same direction and it has been observed that at a distance of about 9 in. from the exhaust intakes this current is strong enough to overcome the air flow toward the exhaust system and carries the dust clouds toward the back of the machine. In order to eliminate this detrimental effect on the exhaust system, the belt should be encased at least on the side facing the beam of the machine. The effectiveness of the integral exhaust system fitted on a 16-in. diam wheel as shown in Fig. 10, is demonstrated in Fig. 14 which gives comparable illustrations when grinding with a conventional machine (a) and when grinding with a machine incorporating the integral exhaust system (b).

III B.S.C.R.A. combined booth and integral exhaust system. The integral exhaust system on the swing frame grinder developed by the B.S.C.R.A., although effective in the control of fine dust generated during grinding of steel castings, has some features which might be inconvenient under some conditions of operating the machine. These limitations are due mainly to the application of a flexible exhaust pipe and flanges extending at the rear of the hood, which may restrict the maneuverability of the machine.

It is obvious that the booth exhaust system is free from these particular restrictions, but observations have shown that, although the control of the primary dust stream does not present any difficulties, the secondary dust stream cannot, under various conditions of grinding, be controlled successfully even when very high exhaust volumes (6000 cfpm) are employed.

Experimental work was carried out on a *combined exhaust system,* the principle of which is shown in Fig. 15. The primary dust stream was controlled by the flow of air into a booth through which 2000 cfpm was exhausted. The velocity of the exhaust air was approximately 75 fpm in the booth opening. To control the secondary dust stream a fan of 100 cfpm capacity, driven by a ½-hp motor, was fitted on top of the swing frame grinder.

The suction side of this fan was connected with the grinding wheel cowling through the hollow beam of the machine (if a machine has a solid beam, a pipe connecting the fan with the wheel cowling can be

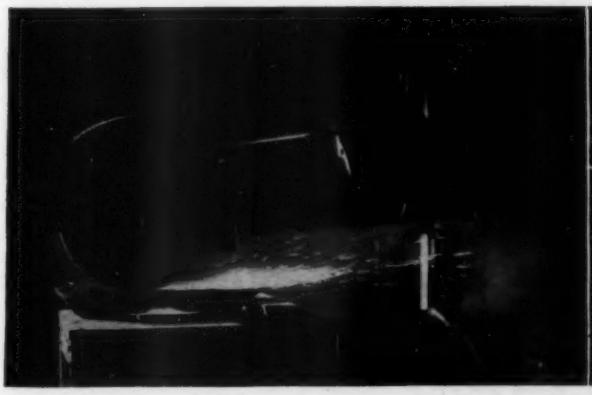


Fig. 14A — Dust clouds generated when grinding with a conventional swing frame grinder machine.



Fig. 14B — Dust controlled when grinding with a machine incorporating B.S.C.R.A. integral exhaust system.

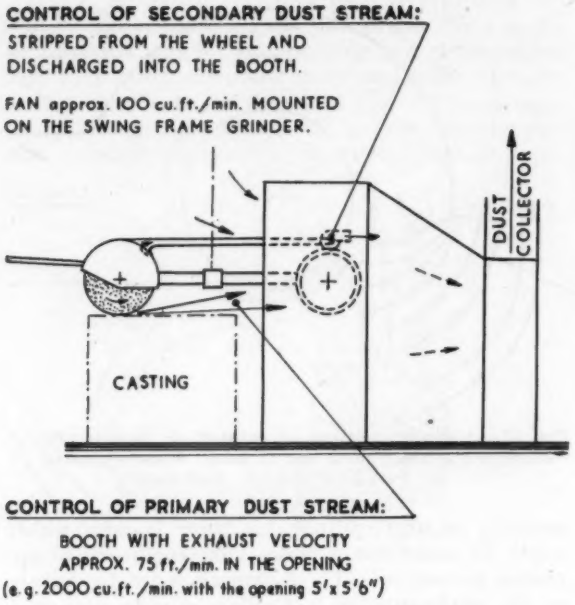


Fig. 15 — B.S.C.R.A. combined booth and integral exhaust system for a swing frame grinding machine.

fitted easily). The wheel cowling was modified according to the B.S.C.R.A. integral exhaust system. The outlet of this fan was directed toward the booth opening so that the fine dust stripped from the grinding wheel was discharged into the booth. No flexible pipe is required and no restrictions are imposed on the movements of the machine.

Observations have shown that a combined booth and integral exhaust system provides a satisfactory control of dust when grinding with a full size wheel (16-in. diam) or a worn-down wheel. Even when the grinding wheel is several feet in front of the booth opening, which might occur when grinding large castings, the effectiveness of this exhaust system was maintained providing the primary dust stream is directed into the booth opening. It would appear that the combined exhaust system not only secures a high degree of dust control but is also likely to show advantage in capital and operating cost.

The factories and foundries regulations[11] and recommendations indicate clearly the requirements regarding the cleanliness and housekeeping; regular washing and painting of walls; construction and maintenance of firm and level floors and regular cleaning of floors; provision of clearly defined gangways properly maintained and kept clear; orderly arrangements and storage in suitable areas of molding boxes, patterns, rammers, clamps, etc.; an adequate and clear space to be left around the melting furnaces, fire boxes of stoves; wetting of dust before disturbing it by sweeping; use of vacuum plant for removing dust.

These indications of how to achieve a desirable high standard of tidiness and cleanliness are widely known and applied in many foundries. To be fully successful all these operations have to be well organized and systematically carried out. No legislation can make foundries attractive places to work in without an enthusiastic approach by management.

Conclusions

While referring to various methods of suppressing dust no clear indication has been given regarding the degree of suppression achieved or required, as there are no generally accepted efficiency figures and no generally accepted method of assessment of efficiency. In assessing the degree of dust control by local exhaust systems the investigators very often concentrate on assuring that "the operator's breathing zone" has the same dust count as the general atmosphere or that no dust clouds reach it.

Although there is a wide field for improvements, the present knowledge of dust problems and the methods available, if applied intelligently, will give greatly improved working conditions in foundries.

By avoiding handling and disturbing sand and dust in a dry state the main source of atmospheric contamination in the molding bay will be eliminated.

If there were no sand adhering on castings when they reach the cleaning room, there would not be a silicosis hazard for cleaning room personnel. This does not mean of course that no dust suppression would be needed in the cleaning room. Any extra care or the use of improved methods in making molds and pouring, which result in a better strip and cleaner castings, will not only assist in combating dust but also may prove economically advantageous. Employing the most effective equipment and methods for cleaning castings before passing them to cleaning room operators is another step in the right direction.

Local exhaust systems can be very effective in controlling dust near the point of its generation, but it must be remembered that the air intakes and baffles will not control dust without the air flow required to carry the dust into the exhaust hood. Maintenance of the ventilation equipment and regular check-up of the exhaust volume are essential to ensure the efficient control of dust.

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THE HUMAN FACTOR IN NEW EQUIPMENT

By
John Taylor*

Great strides are continually made by equipment manufacturers in producing machines that perform work faster and better. By advertising, mailing literature, and personal contact, merits of the equipment are made known to the prospective purchaser.

When a new piece of equipment is sold to a company, it is sold with the sincere belief that it will produce the quantity and quality for which it was intended. The purchaser is also convinced that the equipment will produce the desired results, or he would not be the purchaser. At this point, the seller and the buyer are in accord.

These conditions must be met before a sale is made, and considerable time and expense may be involved before the purchaser decides to buy the equipment. Unfortunately, the purchaser is seldom the individual responsible for obtaining the results claimed by the equipment manufacturer.

Considering this last statement as a basis of fact, it is self-evident that the foreman or supervisor can not be enthusiastic about new equipment for which he is to be responsible if a sincere effort has not been made to sell him on its merits. In many instances his attitude is one of complacency, and he accepts the new equipment because the boss ordered it. Inwardly, he may resent the new equipment because he had little to say about the selection, or because he may have heard from miscellaneous sources that the equipment is over-rated.

Dissatisfaction With New Equipment

This kind of condition exists in many plants today, and is the underlying reason why many companies become dissatisfied with new equipment. They may be told by the foreman that the equipment is no better than the old equipment, or it does not produce the expected results. In many cases, it then becomes necessary for the actual purchaser to go into the shop and try to prove the value of the machine in order to justify his own convictions. Failing this, expert help is usually requested of the equipment maker.

Assuming the representative of the equipment company is successful in demonstrating the merits of the machine, and the plant is now satisfied with its purchase, there is still the item of wasted time and expense involved in proving the value of the machine.

Thus far, this article may seem to be discussing the routine procedure of the purchase of new equipment. However, careful consideration of the basic facts will reveal an influence usually overlooked. This influence is the "Human Factor" and unfortunately gets little recognition in most purchasing.

Much has been written, and also done, about the Human Factor element in the supervisory group toward labor relations, management policies, company costs, etc. So why not in the purchase of important equipment? The supervisor can save the company considerable money, if he is "sold" on the new equipment before it arrives, or it is the kind and size of equipment which he believes will do the kind of job required. It is only reasonable to assume that maximum efficiency of a machine can be obtained only when the operation of the equipment is fully understood, and the desire to "make the machine talk" exists in the mind of the person responsible for its operation. This basic point, then, is the key to the problem of getting the results anticipated.

Operator-Purchaser Coordination Required

It is true that the operating personnel at plant level are normally not in a position to interview the sales representatives of equipment manufacturers. By the same token, unless there is extremely close coordination between operating personnel and the person who does the purchasing, the latter may not be familiar with such important detail as depth of foundation, head room, sand supply, etc.

It is also true that operating personnel are not in a position to decide on the economics of the purchase of new equipment. The purchaser, who has the authority to make the decision on the purchase, may or may not be as thoroughly informed on the merits of the equipment as he thinks he is.

This definite gap is usually bridged by the industrial engineering or plant engineering departments in large companies. However, in most medium- and

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small-sized foundries, this gap does exist and is the reason for considerable lost efficiency.

Some individuals may insist that a good loyal supervisor will accept any piece of equipment purchased for his department. Unquestionably, this is true, but it is not the whole truth. He will accept the machine, but the question is — how? Again, it is generally agreed that most loyal supervisors will try to use the machine effectively, because the boss purchased it. The remaining loyal supervisors may not try too hard to use the equipment effectively because the boss purchased it and they were not consulted. So, going from the loyal supervisors to the ones whose loyalty is just average, or to the prima donna type, this condition becomes even more aggravated.

It is a known fact that such conditions do exist to a greater or lesser degree in every organization, and they will always exist where there are people. Assuming recognition of a problem, the objective then becomes its solution.

The condition should be viewed in much the same manner as other problems dealing with human relations, and that is to discover the basic reason behind the poor attitude toward this new equipment.

Responsibility of Supervision

Even ruling out the abnormal personalities of the prima donna, or the few cases of below average loyalty, it is necessary that recognition be given to the extent of the responsibilities of the supervisor. Many of these are defined by management, and many are assumed without definition, but the fact remains that the supervisor is held fully responsible for the results attained by his department.

He is held responsible for the conduct and output of his labor force, and he is also held responsible for the production, maintenance, and operating efficiency of his equipment.

It is interesting to note that the average supervisor is being constantly reminded of his responsibilities, either because management believes he needs to be reminded or because management is trying to obtain greater efficiency or more output. Therefore, it is a very obvious fact that the supervisor is held responsible for the results obtained from, and the maintenance of, the equipment within his department.

Accepting the proven scientific theory of cause and effect, it is found that the cause in this case is produced by the supervisor recognizing his responsibility to produce with the equipment provided for him by management. Under these conditions, authorized and stimulated by management, it is only human nature for a supervisor to take the most pride in, and to give his special attention to, that equipment in which he is most interested.

Therefore, the effect portion of the cause and effect theory is the mental reaction of the supervisor, and this is governed to a large extent by management's actions when a piece of new equipment is purchased. If management recognizes its own policy of holding the supervisor responsible for getting the optimum results from the equipment in his department, then it will recognize the importance of making sure that the supervisor is thoroughly sold on the new equipment and believes it will do what it is supposed to do.

Attitude of Supervision

If management does not recognize the value of holding a man responsible for things in which he believes, and purchases equipment which is not liked, for one reason or another, or it is the wrong capacity, then the effect will be opposite and the results unsatisfactory. The mental reaction of the supervisor is then one of resentment, to a degree, because he is being held responsible for a piece of equipment which, in his mind, is not right and, therefore, he should not be held responsible for the results. He reasons, and rightly so, that if he is to be responsible for something, he should have some opportunity to express his opinions before the equipment arrives, and preferably before it is purchased. In a sense, it is much the same as the purchaser of a car who wants a certain style and color, and who is not satisfied with anything different.

Reviewing the side of management, or the owner of the business, it is their money and, therefore, they should have the privilege to spend their money as they see fit. Actually, this is correct, but it is not being done wisely if it is not fully appreciated and properly accepted.

To some people, these comments may seem to be without any facts, but to the industrial engineer, and particularly to the consulting engineer who gets into scores of plants, the foregoing conditions are found quite often. When they are, it usually permits the engineer to produce results more quickly simply by proving to the immediate personnel that better results can be obtained.

Some executives are aware of this problem and make every effort to minimize or eliminate it from their organization. Those who do not, or those who feel that they should spend their own money as they choose, will benefit greatly from a change in procedure.

APPLICATION OF RADIOGRAPHY IN THE MANUFACTURE OF BRONZE CASTINGS

By

N. A. Kahn,* Solomon Goldspiel,** and R. R. Waltien**

ABSTRACT

The authors present a series of radiographs and allied material to illustrate the application and value of radiography in the bronze foundry. Radiographs and discussion are divided into five categories according to the phase of foundry activity represented. The five phases are: (a) Gating and Risering, (b) Evaluation of Soundness, (c) Inspection, (d) Salvage, (e) Procurement. Several examples are presented in each category and the discussion accompanying each example clarifies the defects and discontinuities shown and relates the appearance of the radiograph to the specific problem involved.

The paper includes a brief discussion of radiography and the importance of standards for its optimum utilization in the bronze foundry.

Introduction

Radiography is probably the oldest and most commonly used nondestructive method as applied to industrial material testing and inspection and yet it is relatively a recent engineering development. Radium radiography, for example, had its inception only about 28 years ago when Mehl et al[1]† did the pioneering work at the U.S. Naval Research Laboratory. As a matter of fact, many of the general procedures and practices followed to this very day in industrial gamma-ray radiography are based on developments in the "early thirties" by Mehl and his associates.

X-ray radiography, while actually older than radium radiography, began to assume industrial significance only as late as World War II. It emerged as a natural development of a strong impetus, about that time, to produce x-ray equipment primarily designed for industrial work and to replace the makeshift medical equipment used until then.

The potential of radiography as an important foundry tool was recognized early in its industrial development.[2,3,4,5,6,7] Its acceptance grew however as the development and validation of reference or standard radiographs progressed.

Military agencies recognized the importance of standardization and sponsored development and preparation of standards in the early stages of growth of radiography. Thus the U.S. Navy Bureau of Engineering issued "Gamma-Ray Radiographic Standards for Steel Castings for Steam Pressure Service"[8] as early as 1938. These were subsequently reissued as "Radiographic Standards for Steel Castings" by the U.S. Navy Bureau of Ships in 1942 and ultimately adopted as "Industrial Radiographic Standards for Steel Castings" by the American Society for Testing Materials in 1952.[9] These standards were originally established primarily because of their importance in the evaluation of steel castings used in power generation and in certain strategic structural parts of motive machinery in naval vessels. The Bureau of Aeronautics followed with the development and preparation of reference radiographs for light metals which culminated as its "Reference Radiographs for the Inspection of Aluminum and Magnesium Castings" and which were ultimately adopted by the American Society for Testing Materials.[10]

Paper Objective

While radiography has been applied through the years to copper-base alloy castings, the amount of standardization work reported in this field is relatively sparse. In effect radiography in the brass and bronze industry is currently at the stage where the radiography of steel stood in about 1938. Fortunately the experiences gained in the other two important base metal types will tend to accelerate the growth of radiography in the brass and bronze industry. The object of this paper is to illustrate the applicability of radiography in the brass casting industry and to pinpoint the advantages of its use to the foundryman, inspectors and casting consumers.

Related Work by Other Investigations

Literature specifically concerned with radiography of non-ferrous castings is not extensive. This phase of radiography is usually treated as one section of the

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†Numbers in superscript refer to References appended.

general subject of nondestructive testing. The Marburg Lecture of 1952 by R. C. McMaster[11] includes a very extensive bibliography on nondestructive testing from which much information on radiography may be gleaned. This paper also includes an unusually clear and concise discussion of radiography as related to nondestructive testing in general.

Romanoff,[12] in a 1949 paper on "Quality Control of Copper-Base Alloys" also includes a short discussion of radiography as an inspection tool. He says in part that one of the disadvantages of radiographic inspection is "– the lack of agreement in interpreting the radiographs of brass and bronze alloys. Radiographic standards have been made for steel and aluminum alloys, but to date none are available for the copper-base sand casting alloys." This statement sums up the current status of radiography in relation to its use in the examination of non-ferrous castings.

The first important efforts to study radiography as specifically applied to bronze castings are contained in two papers by W. H. Baer[13,14] which contribute to the knowledge of the defects and discontinuities to be found in tin bronzes as revealed by x-ray examination. The latter of the two papers also includes a discussion of hydrostatic tests performed on some of the defective plate castings.

Radiographic Technique

Most foundrymen have come into contact with radiography at one time or another. Chest x-rays are routine procedure in modern physical examination and no dentist will proceed without first viewing x-rays of the teeth involved.

Similarly, radiography is of great importance to the foundryman in connection with the manufacture of castings. By its ability to reveal the interior continuity and homogeneity of metals radiography can assist the foundryman in many ways, such as:

(a) *Determination of the adequacy of gates and risers.* By the type, location, and amount of shrinkage shown on a radiograph the foundryman is able to evaluate the efficiency of a gating and risering system and to plan modifications calculated to overcome the defects noted. This subject was treated by B. N. Ames in greater detail.[15]

(b) *The evaluation of the soundness of castings.* An experienced technician can evaluate a casting on the basis of the comparison of its radiograph to suitable references or standards to determine the fitness of the casting to withstand specific service requirements.

(c) *Inspection.* In the inspection of foundry production radiography enables the inspector to remove from the production line castings which appear sound on the surface but which might show defects after machining, or in some cases to sort them into categories such as sound, repairable and rejectable.

(d) *Salvage of castings.* Some types of bronzes are not amenable to repair – that is, their metallographic structure may be adversely affected, or perhaps some constituent of the alloy may melt or vaporize at welding temperatures. In the case of alloys which may be repaired by welding, radiography is indispensable in the planning and follow up of the defect removal and

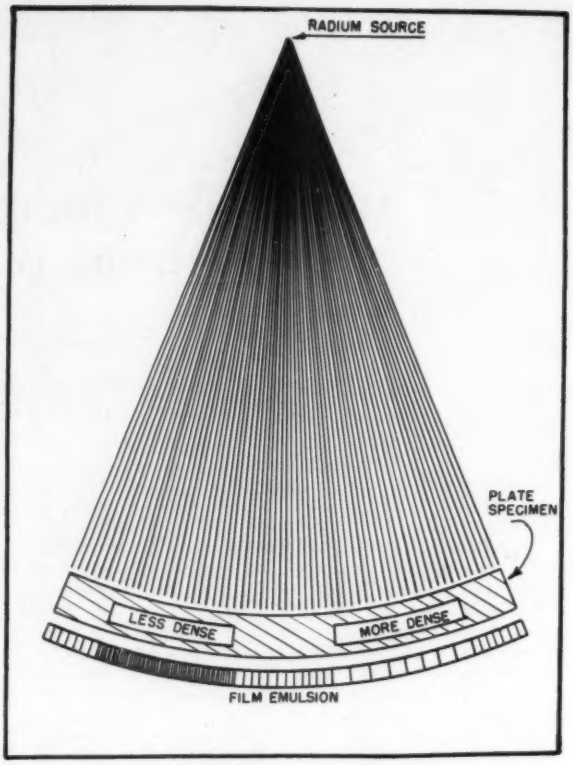


Fig. 1 — Diagrammatic representation of the basic principle of gamma-ray radiography.

in the verification of the soundness of the repair weld. The original inspection radiograph is used to locate the defective area of the casting for the removal of the unsound material by chipping, or grinding. After the defect has been removed, radiography may be used to check the completeness of removal of the defect. When the area has been repaired a radiograph is made to prove the final soundness of the weld.

(e) *Procurement.* In the procurement of castings radiography may be specified in the purchase contract to be performed by the manufacturing foundry, the buyer or by an independent laboratory. In any case such examination assures delivery of sound castings.

Industrial radiography utilizes other radioactive media in addition to x-rays to study the internal soundness of metals. Of these radium is the most popular because its gamma-rays, having shorter wave lengths than the x-rays produced by medium voltage industrial x-ray machines, possesses greater penetrating power and because the radium capsule is completely portable as compared with the complex apparatus required for the generation of x-rays. Since the advent of atomic fission and the creation of radioisotopes other gamma-ray sources are coming to the fore in special applications.[16] A few of these applicable radioisotopes are cobalt 60, thulium 170, iridium 192.

The chief advantages offered by the artificial radioactive fission products, in addition to portability, is that they afford a choice of penetrating power commensurate with good sensitivity for various alloys at comparatively low cost.

The disadvantages of these materials compared to x-rays are that

(1) for existing source activities the exposure times are rather long;
(2) over a period of time the activity changes at rates which depend on the isotope half-life, necessitating periodic recalculation of the source strength.

Properties of the isotopes which may at times act as advantages and at other times as disadvantages when compared to x-rays are that

(1) the contrast of gammagraphs is rather lower than that of exographs and
(2) the energies of many of these sources are limited to one or two wavelengths, hence wavelength and penetration, which is fixed by it, cannot be varied as with an x-ray tube.

All radiographic examination, whether medical or industrial is based on the principle of absorption of the rays as they pass through the material under examination. The sketch, Fig. 1 illustrates the changes in strength of the radioactive rays as they pass through a material under examination and affect the photographic emulsion of the film on the other side of the material. It should be noted that the intensity of the rays striking the photographic emulsion varies with thickness and density of the material through which it passes before reaching the emulsion.

The making of a setup for radiography is not a complicated procedure and may be performed by comparatively unskilled personnel once they have been oriented and familiarized with the tools and aids available to them for this purpose.

Factors which must be considered in preparing the radiographic setup are:

(a) geometry of casting

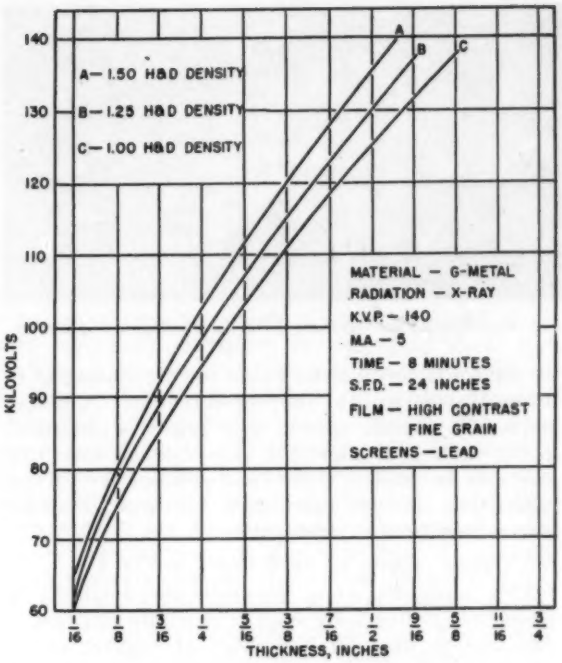
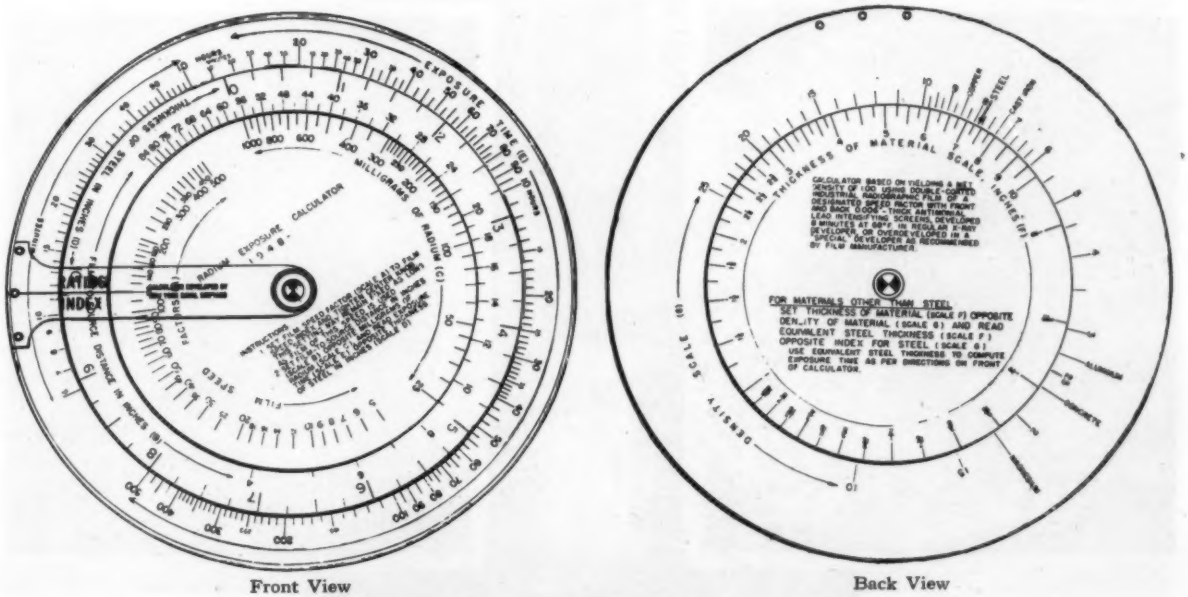


Fig. 2 — Typical exposure chart for 140 kv x-ray apparatus.

(b) section thickness to be penetrated
(c) intensity of source at hand
(d) type, speed and size of film
(e) source to film distance.

These factors can all be simplified and integrated to the point where it is possible to prepare graphic formulas for x-ray use which require only a simple reading to set the apparatus. A typical chart for use with an x-ray machine is shown in Fig. 2. For gamma-ray sources slide rules are available, on which



Front View Back View

Fig. 3 — Universal exposure calculated for radium radiography.



Fig. 4 — Typical radiographic setup for castings using radium.

the known factors may be set for the determination of the unknown, such as the exposure time. A photograph of a circular radium slide rule[17] is illustrated in Fig. 3. It is of interest to note that this slide rule may also be used to make computations for various radioactive isotopes once their radium-equivalency factors have been determined.

A typical radium set up is shown in Fig. 4.

This particular set-up illustrates the flexibility of gamma-ray radiography since it demonstrates how it may also be used for a number of castings at the same time. By placing the source at the center of a circle and suitably arranging the castings about it, different size castings with varying section thicknesses can be exposed simultaneously.

Film for industrial radiography is specially coated with emulsion on both sides of the base in order to double the effect of the radioactive x-rays and therefore permit shorter exposure. Standard stiff or flexible cassettes are used, according to the application and lead screens are usually employed to intensify the effects of the rays. After the exposure, the film is developed in special radiographic developer by conventional darkroom techniques.

Examination and interpretation of processed radiographs is one of the more difficult aspects of the examination of castings by the radiographic method. The radiography inspector must be able to visualize the portion of the casting covered by the radiographs; he must be familiar with foundry techniques employed; he must recognize and distinguish between all of the defects or discontinuities shown in the radiograph; he must be familiar with the types of sources and films and their effects on the appearance of defects and discontinuities; he must have sufficient background and experience to know the relative effects of the various defects and discontinuities on the serviceability of the casting under examination, and finally he must be able to locate the defective areas of the castings to guide the chipper in the removal of the defects.

Some guides and helps are available, however, to make the task of the evaluator less complicated. A sketch or plan of the casting is of value in assisting the technician to visualize the casting during evaluation and in laying out the defects on the casting itself for repair. A sketch of the gating and risering arrangement is useful in diagnosing the defects, determining

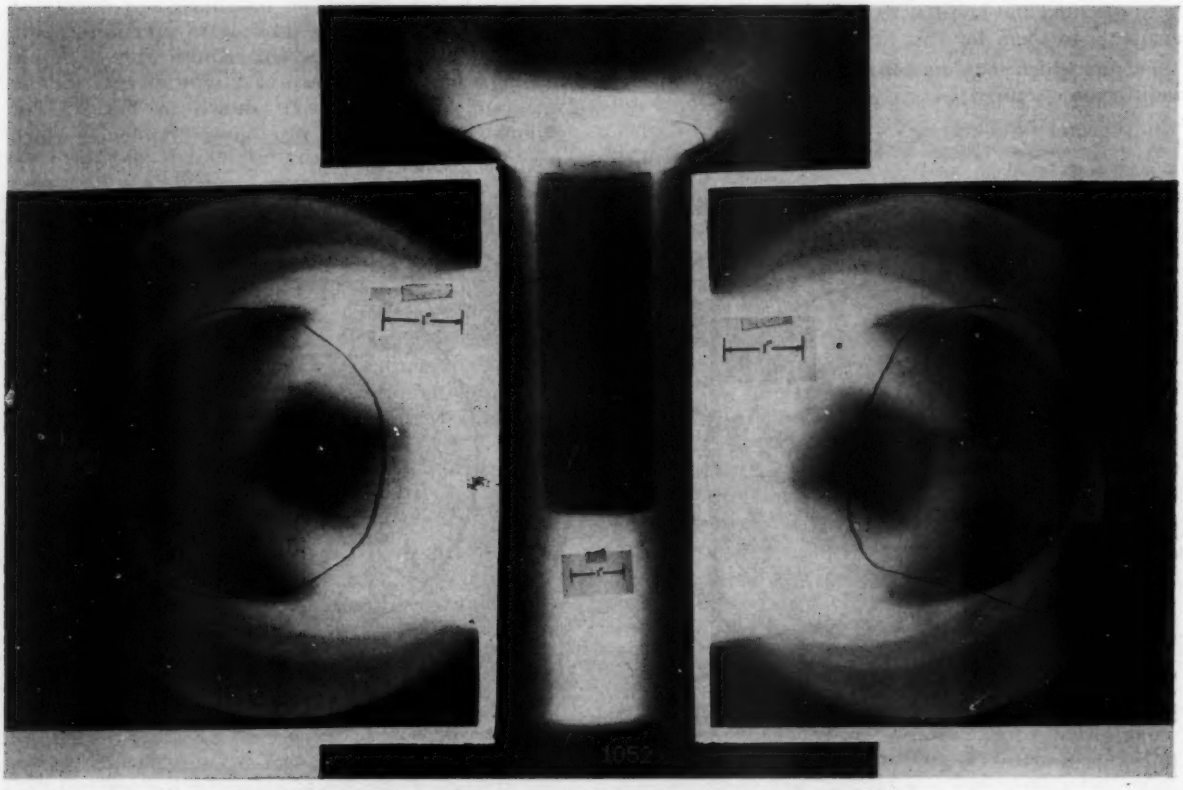


Fig. 5 — Foundry practice series 57-3, overall and flange gammagraphs.

causes and prescribing remedies. Reference radiographs or radiographic standards for the alloy under examination are of inestimable value in grading the various degrees of defect according to service requirements.

In the foregoing only the highlights of radiography are presented. For more detailed information on the theory, mechanics and techniques reference may be made to available texts on the subject.

To illustrate the applications of radiography and its value in the brass and bronze foundry the authors have selected several examples for each of the five major categories of applicability enumerated above. Photographs and radiographs for these illustrations, except for the reference radiographs, have been selected from the files of the laboratory with which the authors are associated and represent actual castings produced in the New York Naval Shipyard foundry and intended for shipboard use or for use in the research and development program of the laboratory.

Included in these series are radiographs representing just about every major defect or discontinuity known to occur in bronze castings. In a few of the cases particularly those in which heavy shrinkage is involved, some indication of the defect was evident on the surface of the casting. However, in most of the examples the surface appearance was of a good, sound casting.

The authors wish to point out that the examples shown here represent the worst castings accumulated over a 15-year period, including the war years, and should not be construed as representative of the New York Naval Shipyard production.

Gating and Risering

Foundry Practice Series

In developing gating arrangements for new and intricate castings the shipyard foundry and the material laboratory cooperate in recording data and evaluating the effects of changes in foundry technique. Figures 5 through 8 illustrate two gating arrangements for a valve bronze piston casting with data sheets and gammagraphs for each. The photograph of the casting forms a permanent record of the pertinent details of the gating arrangements. The data sheet records the foundry techniques used and the results of the subsequent examinations of the casting. The gammagraph shows the relative efficiency of the risers used and the effectiveness of the gating system.

Data and graphic material as shown in Figs. 6 and 8 are assembled for each pilot casting. After evalua-



8/25/48

FOUNDRY PRACTICE SERIES NO. 57 TEST NO. 3

Pattern No. 8-397748 J.O. No. 45026-1-153

Description Piston

Heat No. 9131 Alloy M Date Cast

Supervisor Moeller Surface Inspection

SAND CONDITIONS

Mold Mixture Moisture %

Green Comp. Perm. Fineness Clay Content %

Core Mix Radiography

MOLDING

Down Gate 2" dia. x 23" long In Gate (1) 1" x 4" (2) 2½" x 1½" Risers 3½" taper to 2½" dia. x 5" long

Remarks Dummy risers feed ingates to casting 2" dia. x 4" high. Cast inverted (Flange Down) (Top Solid)

MELTING

Furnace No. 9-E1 Melter Bergendale Started 13:04 Arc off 13:45 Tapped 13:55 13:45

Tapping Temp. °F Pouring Temp. °F of bars 2040 Casting 2100

Charge	Cu	Sn	Zn	Pb				Ingot	Gates	P-C
900			2½					300	600	

PHYSICAL

	Yield Pt. P.S.I.	Tensile P.S.I.	Elong. % in 2"
Spec. 4688		36,000	22%
9131	22,980	44,440	40.0

CHEMISTRY

	Cu %	Sn %	Zn %	Pb %	Fe %	P %	Ni %
Spec. 4688	86-89	5.5-6.5	3-5	1-2	.25max.	.05max.	1.0max.
9131	87-96	6.25	3.68	1.47	.03	.02	.59

INSPECTION

Radiography #1052 Gross shrinkage in fillet area

Oil Penetrant

Recommendations Rejection recommended

Hydrostatic Pressure Test

Other Inspection Tests

General Remarks

Fig. 6 — Foundry practice series 57-3; gating arrangement and data sheet.

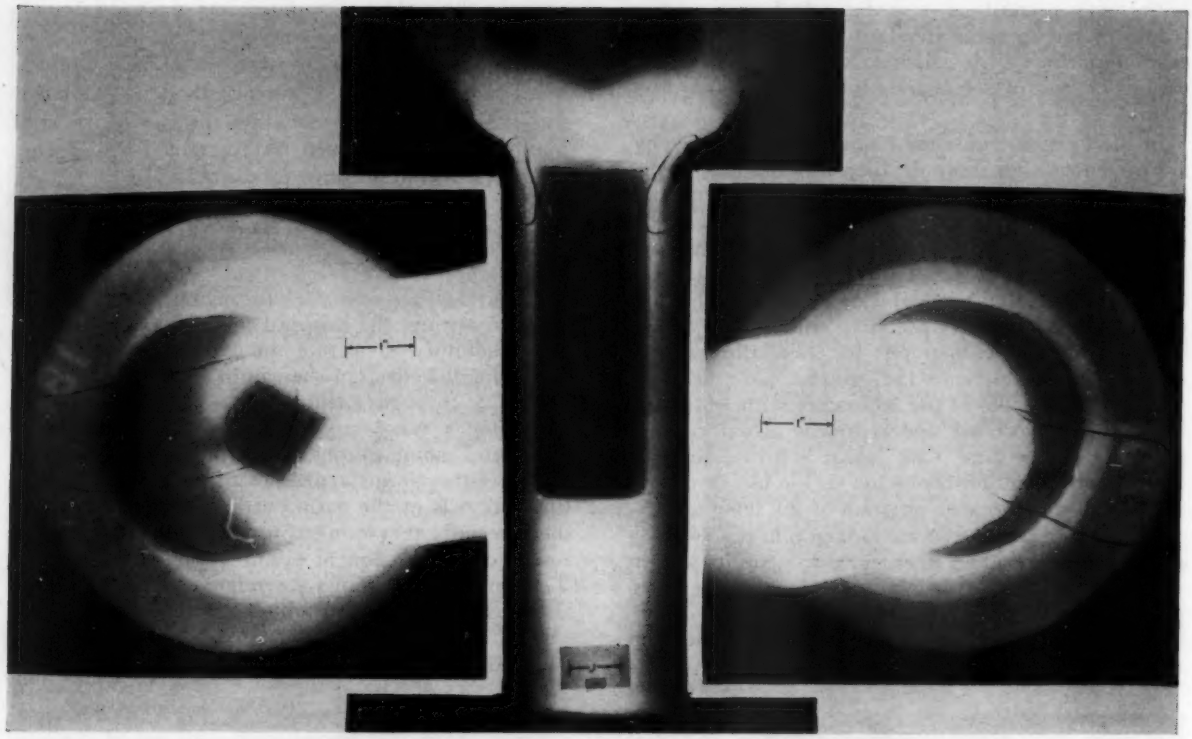


Fig. 7 — Foundry practice series 57-4; overall and flange gammagraphs.



FOUNDRY PRACTICE SERIES NO. 57 TEST NO. 4

Pattern No. 8-397768 J.O. No. 45086-1-001

Description Piston

Heat No. 9198 Alloy M Date Cast

Supervisor Surface Inspection

SAND CONDITIONS

Mold Mixture Moisture %

Green Comp. Perm. Fineness Clay Content %

Core Mix Radiography

MOLDING

Down Gate 2½"dia.x20"high In Gate (2) 3" x 2" Risers 4"dia.x7"high

Remarks: Ingate to riser 1"x½". Runner to ingates at base 1½"x2". Dummy riser on ingate used in previous practice, eliminated for this casting.

MELTING

Furnace No. 3-M2 Melter Sternisky Lighted Arc on 13:25 Arc off 14:12 Tapped 14:17

Tapping Temp. °F 2300 Pouring Temp. °F of bars 2040 Casting

Charge	Cu	Sn	Zn	Pb			Ingot	Gates	F-C
850	88	2	6½	6 ¾			400	350	

PHYSICALS

	Yield Pt. P.S.I.	Tensile P.S.I.	Elong. % in 2"
Spec.		36,000	22
9198	20,000	42,750	50.0

CHEMISTRY

	Cu %	Sn %	Zn %	Pb %	Fe %	P %	Ni %
Spec.	86-89	5.5-6.5	3-5	1-2	.25max	.05max	1.0max
9198	86-18	6.61	3.03	1.62	.01	.01	.5%

INSPECTION

Radiography #1096 Shrinkage is well above fillet & at ingates in fillet area.

Oil Penetrant

Recommendations

Hydrostatic Pressure Test

Other Inspection Tests

General Remarks

Fig. 8 — Foundry practice series 57-4; gating arrangement and data sheet.

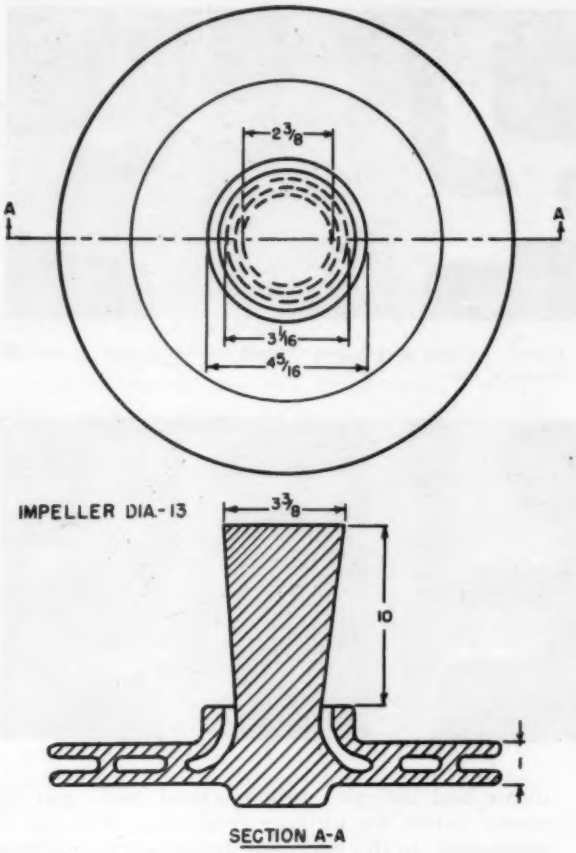
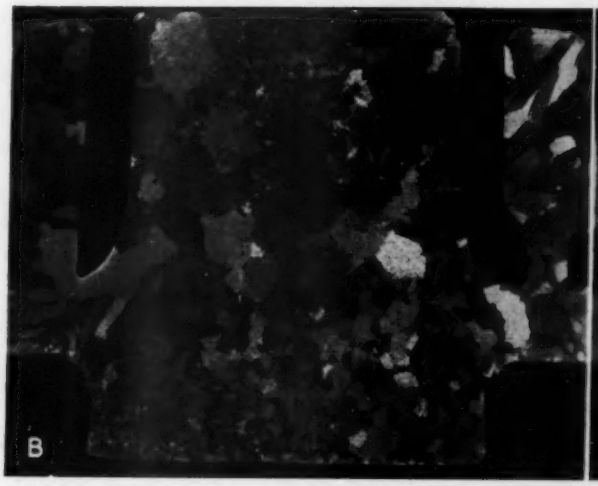


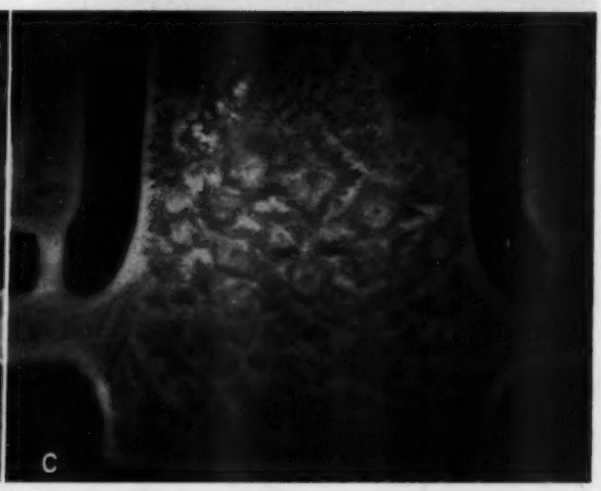
Fig. 9 — Original gating arrangement of gunmetal impeller.

tion, recommendations are made for changes in the foundry techniques which are intended to eliminate the defects found in the casting. Then a second pilot casting is founded and the procedure repeated. Usually one or two pilot castings are necessary to achieve an acceptable casting, but in the case of in-



Fig. 10 — Exographs of ¼-in. cross sections through riser and hub areas of tin bronze impellers.





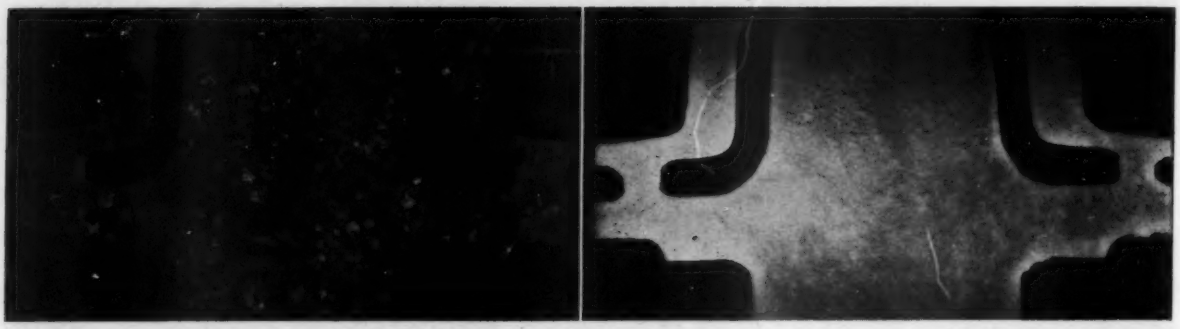


Fig. 11 — Nitric acid macro-etched section (left) and exograph (right) of thin slice taken through center of hub of typical impeller cast with insulated riser.

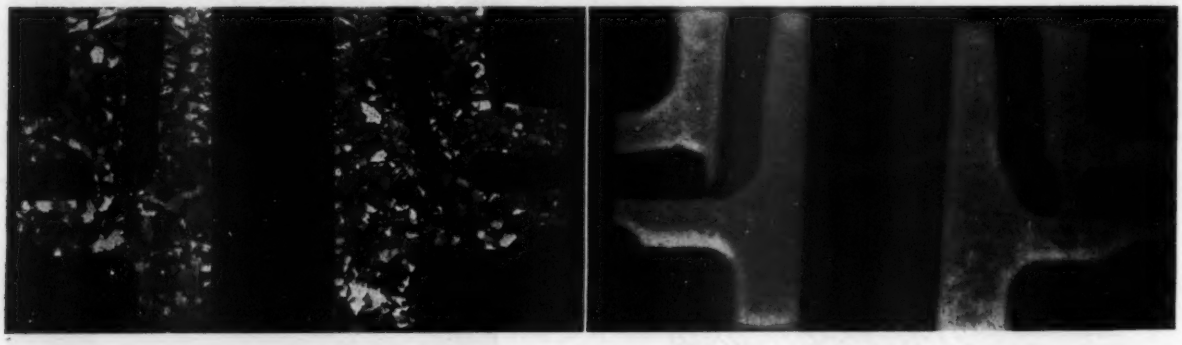


Fig. 12 — Nitric acid macro-etched section (left) and exograph (right) of thin slice taken through center of hub of typical impeller cast with cored hub and no riser.



Fig. 13 — Gammagraph of boss in valve bronze cylinder.

tricate and difficult castings several more may be needed before the ultimate production technique is established. In this particular application the gammagraphs of the two castings show considerable differences in the amount of gross shrinkage present. Casting No. 57-4 is obviously greatly improved over No. 57-3. Of course, casting No. 57-4 is sub-ultimate in that shrinkage is still evident at the ingate areas. However, after additional modification based on the appearance of these radiographs a sound casting was produced, proving that the evolved technique was sound and that it could be used with confidence by the foundry in production runs.

Study of Gun Metal Bronze Impellers

In a study of gating and risering problems in connection with a 13-in. diameter water pump impeller radiography was utilized in combination with acid etching of ¼-in. slices taken through the center of the hub and riser sections as shown in Fig. 9. Figure 10 illustrates the appearance of the riser and hub section (A) and (B) as etched, and as exographed (C). The gross shrinkage at the base of the riser immediately indicates that it is inadequate to properly feed the casting. The thin skin due to the chilling effect of the sand is evident in the radiographs but does not show to great advantage in the etched section. A close study of (B) and (C) will reveal that both methods of examination detail the internal structure of the metal although it must be kept in mind that etching permits examination of only a single plane, while the exograph integrates the full cross-section thickness.

Figures 11 and 12 show the results of modifications

of the foundry practice on the soundness. Obvious improvment is shown in each case over the first casting. However, even in the cored casting some shrinkage remains at the junction of the shroud and ring turn. Notwithstanding this shrinkage however, castings in this series proved satisfactory for service and a recommendation was accordingly made for a change in the foundry technique. Machining and hydrostatic testing of subsequent castings made by this method verified the validity of this decision.

Evaluation of Soundness

Analyzing a Leaky Cylinder

The gammagraph shown in Fig. 13 illustrates the flexibility of radiography and is interesting because it shows how occasional difficulties involved in the examination may be overcome. The boss in question is 3½ in. thick and is located in a confined space inside of an operating cylinder. In order to obtain the exposure a 250-mg radium source was placed 4½ in. from the film (1 in. from the source side of the boss). The exposure time was 55 minutes and the film employed was a fine grain, high contrast type using two 0.006-in. lead screens.

Figure 14 shows the location of the boss in the cylinder casting. Hydrostatic leakage was observed in the cylinder wall at the base of the boss after the ingate had been removed by chipping (area circled). In order to check for shrinkage due to the change in section thickness it was necessary to make the exposure shown in Fig. 13. The film cassette was placed against

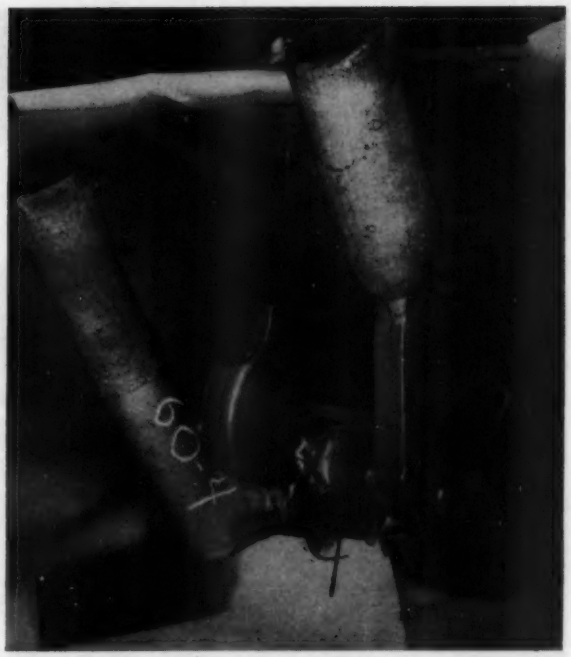


Fig. 14 — Operating cylinder casting showing gating arrangement and location of boss.

the flat surface of the boss to the right of the arrow and the radium was placed in position inside of the cylinder body. Notwithstanding these difficulties it

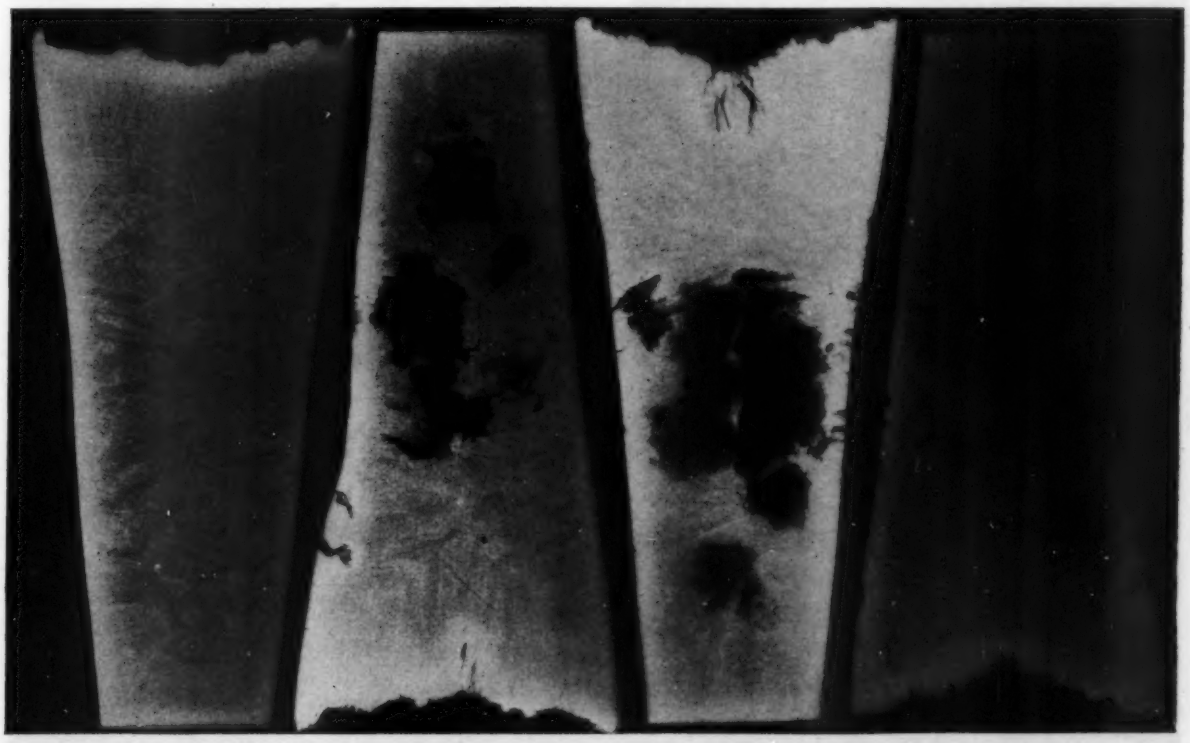


Fig. 15 — Positive print of x-ray of ⅛-in. longitudinal slices of wedge casting.

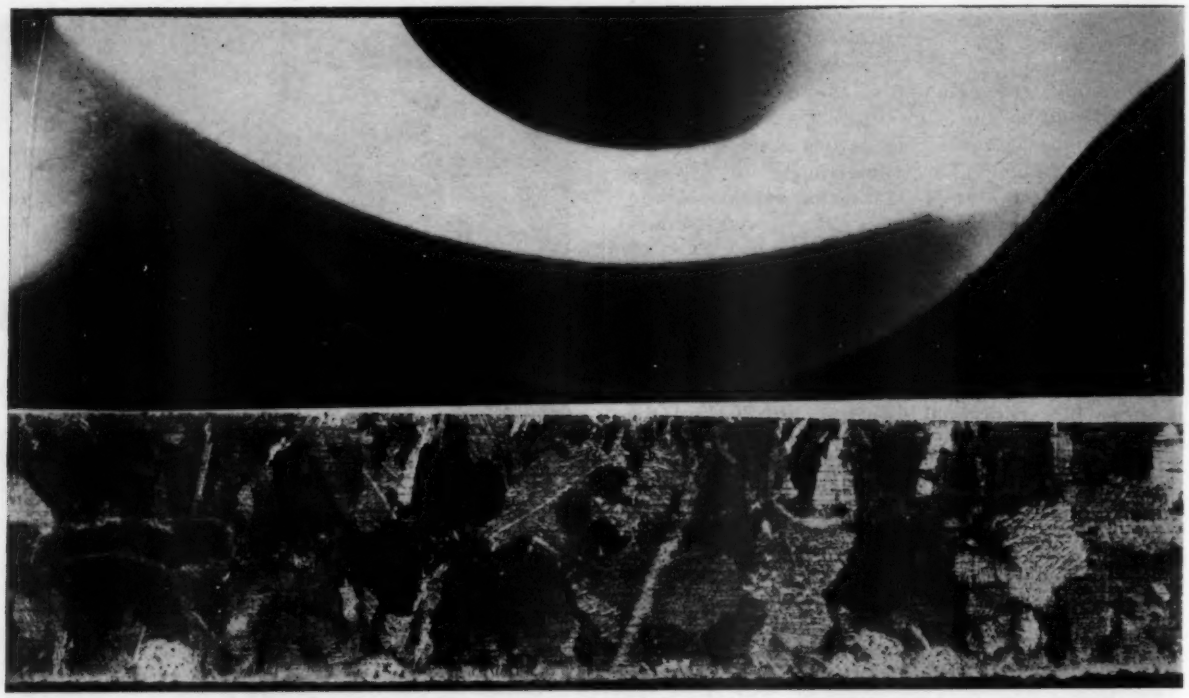


Fig. 16 — Gammagraph of bonnet flange showing slight microshrinkage and macro-etched section through same shrinkage.

will be noted that the resulting radiograph was of acceptable quality and revealed the shrinkage with considerable clarity.

Analysis of Open Sand Wedge Castings

In an investigation of the solidification characteristics of gun metal a number of wedge-shaped castings

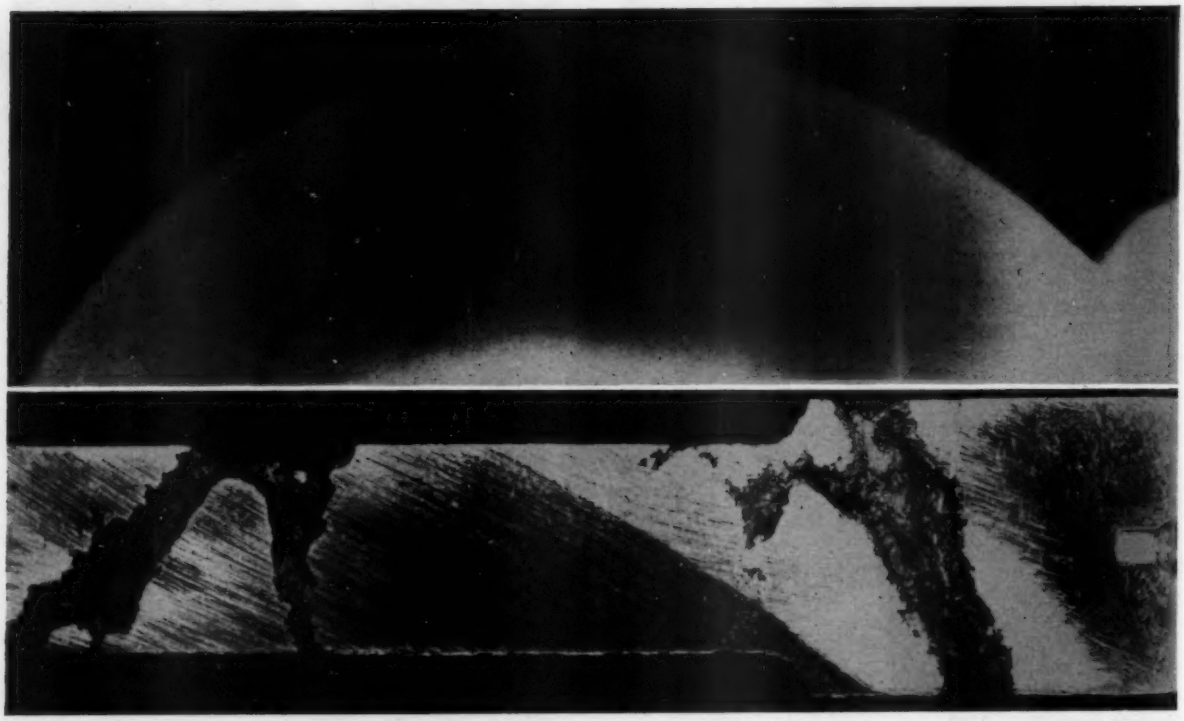


Fig. 17 — Gammagraphs of gunmetal valve body flange showing gross shrinkage, and photographs of sectioned flange (Mag. 2X) showing same shrinkage.

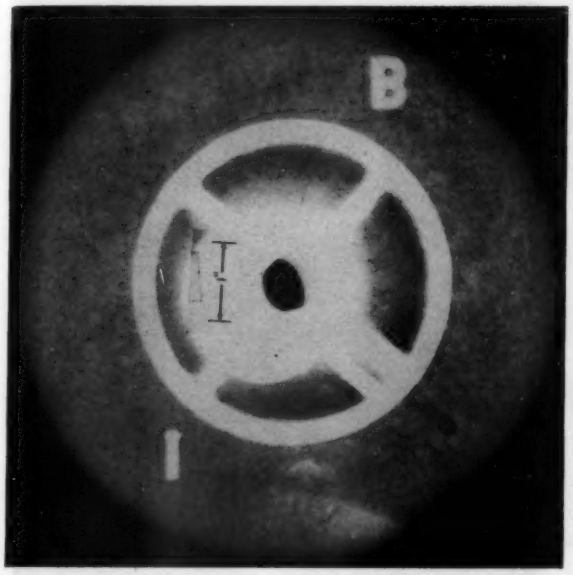


Fig. 18 — Exograph of valve disk showing microshrinkage in tin bronzes.

were poured at different temperatures. In order to evaluate the effect of changes in pouring temperature ⅛-in. thick slices were taken through the centers of the wedges. Exographs of four of these slices are illustrated in Fig. 15. The changes in structure and shrinkage porosity are obviously controlled by pouring temperature since no gating was used. It will be noted that the wedge poured at the highest temperature (2250 F) is completely sound and generally shows columnar crystallization and satisfactory soundness. On the other hand, the wedge poured at the lowest temperature (1950 F) shows equiaxed crystallization with an intimate admixture of microporosity and a considerable amount of segregated gross porosity. In-between wedges show combinations of these characteristics which cannot be tolerated in castings.

An Analysis of Shrinkage Defects

Figure 16 shows a section of the bonnet flange of a high pressure gun metal valve in which slight shrinkage is visible. This type of shrinkage is normally encountered in gun metal castings and may be said to be inherent since it is due to the dendritic mode of solidification of this alloy. The etched macro sections were cut through the defect and have been enlarged to double size. The shrinkage is shown to be scattered throughout the area of the section with some concentration at the centerline. The small section of ingate shown indicates that the gating system used in founding this casting was not adequate since gross shrinkage is present dangerously close to the cut-off line.

Figure 17 shows a similar bonnet flange in which the riser, placed over the flange, was totally inadequate as evidenced by the gross shrinkage extending several inches down into the flange. A longitudinal section of the flanges reveals the characteristics of this type of shrinkage. The photographed sections are enlarged to show more clearly the details of the shrinkage cavities.

Inspection

Valve Disk Shrinkage

The exograph of an "M" metal valve disk shown in Fig. 18 illustrates typical microporosity. Stringers evident at the top (to the left of letter "B") and on the left side (1 in. above number "1") are typical of the channels formed by an agglomeration of minute shrinkage cavities, which result in leakage on hydrostatic test. Concentrations of shrinkage are also evident around the inner fillet of the reinforcement particularly in the lower right quadrant close to the double bar. Where a casting intended for pressure service shows as much general and concentrated microshrinkage as this one it should be viewed with suspicion since leakage after machining, even though only a small area such as the seat is machined, is al-

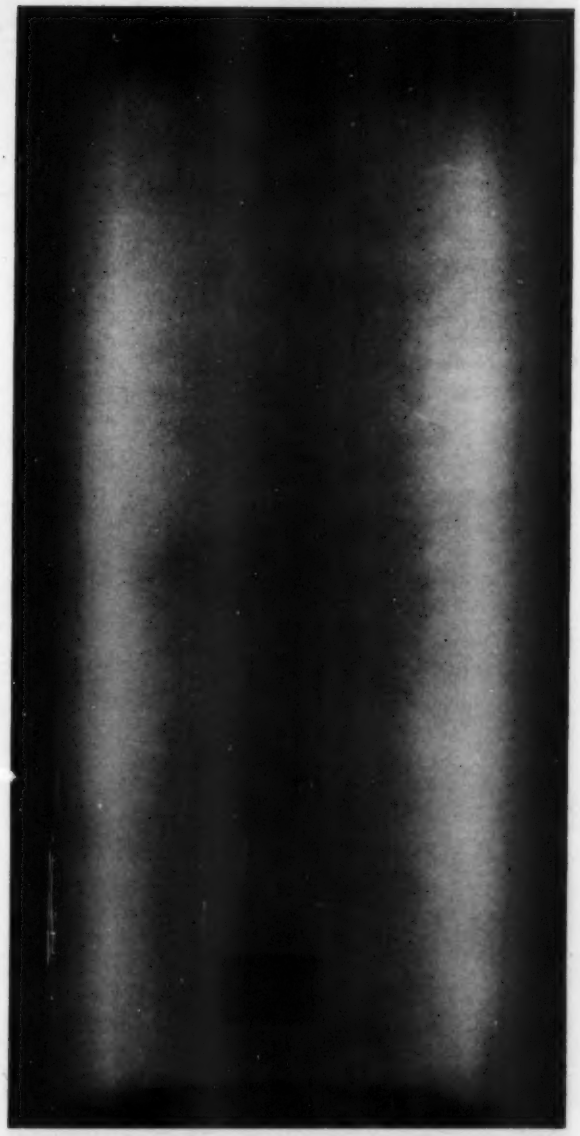


Fig. 19 — Gammagraph of nickel-copper cylinder showing overall microshrinkage and gross shrinkage under riser.

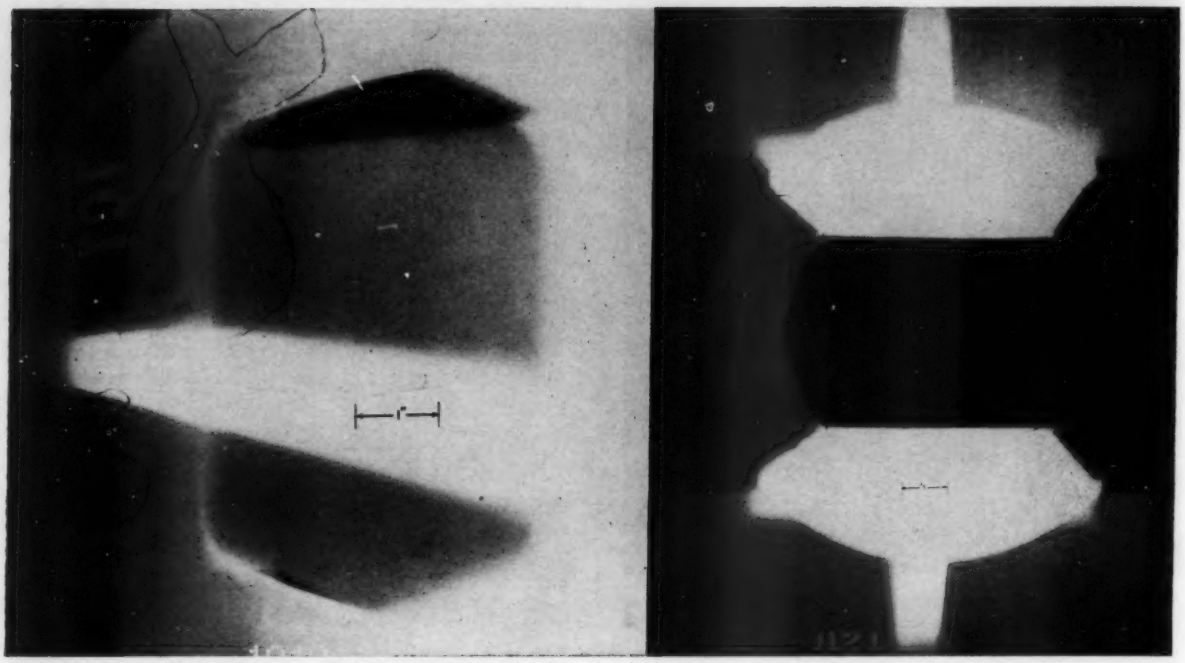


Fig. 20 — Gammagraphs of valve yoke showing gross shrinkage in flanges.

most certain. Radiography of such castings before machining can eliminate many scheduling headaches and of course precludes the waste of money on useless machining.

Copper-Nickel Shell

Figure 19 is a double wall gammagraph of a copper-nickel cylinder using a 500-mg source at 58 in. Wall thickness of the cylinder is 1 in. The cylinder was bottom gated with two peripheral risers. Gross shrinkage is evident under one of the risers (top center) extending as much as 4 in. down into the casting. The overall mottled effect is characteristic of the alloy and does not necessarily indicate unsoundness unless the contrast is somewhat more pronounced than is shown here. However, the gross shrinkage shown in this casting is sufficient to cause rejection.

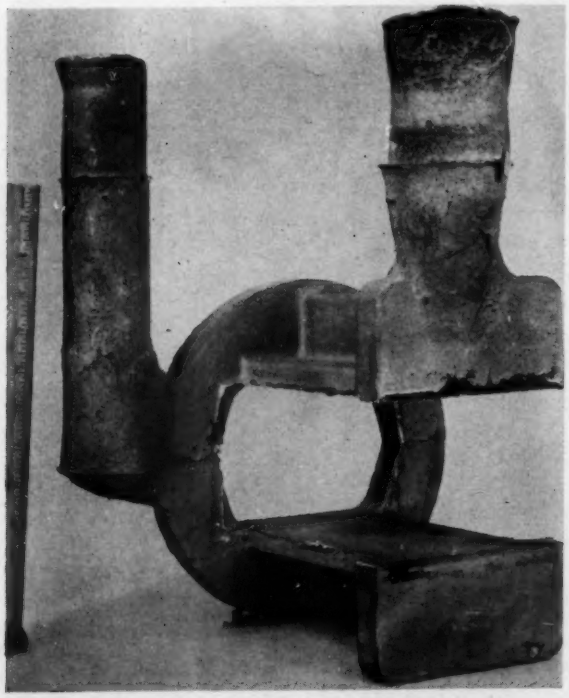


Fig. 21 — Gating arrangement of bronze valve yoke casting.

Gun Metal Valve Yoke

Gammagraphs shown in Fig. 20 illustrate typical gross shrinkage due to faulty feeding in a gun metal valve yoke. The gating arrangement utilized in the production of this casting is shown in the photograph, Fig. 21. It will be noted that the two areas of shrinkage are (a) adjacent to the ingate and (b) on the cope side of the casting continuous with the upper rib. The corresponding rib-flange area on the drag half of the casting was sound. In both of these cases elimination of gross shrinkage was simply and easily accomplished by judicious application of padding-around the rib-flange fillet and at the ingate. Re-radiography then revealed a sound casting throughout.

Salvage

Unfused Chaplets

Figure 22 presents gammagraphs of high pressure gun metal elbows showing cut copper nails used as chaplets and cut etched sections through the same nails. The radiographs show little difference in density between the body of the nail and the background metal, but the unfused periphery of the nail is clearly delineated, particularly the area under the head where gases collected.

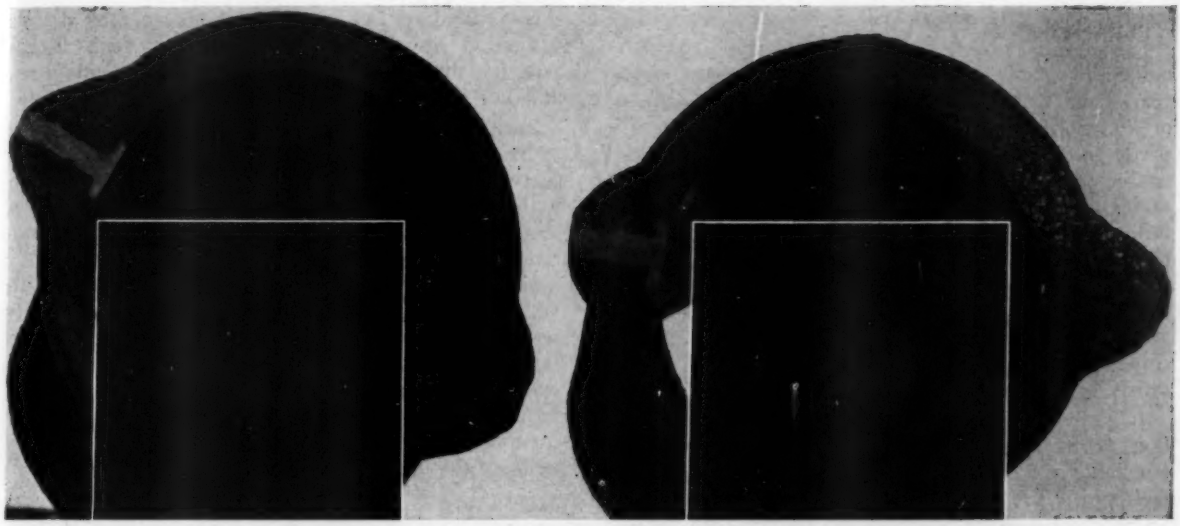


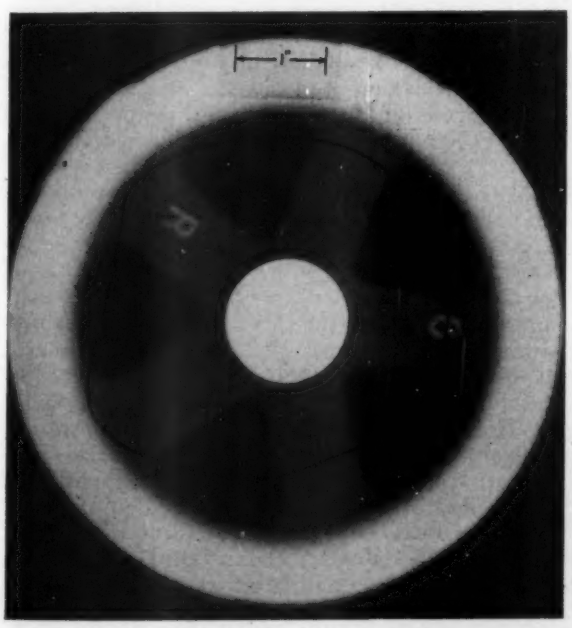
Fig. 22 — Gammagraphs of unfused copper nail chaplets and etched sections of same chaplets.

Procurement

Manganese Bronze Gear Blanks

Figures 23 and 24 illustrate gammagraphs of two manganese bronze gear blanks showing, respectively, dross in the spoke area and shrinkage in the tooth area. Dross, such as this, is one of the major problems in the casting of manganese bronze. An oxide skin is normally formed over the surface of the ladle and even the stream during pouring. If some of this is broken away it is carried into the mold with the result shown. Turbulence of metal in the mold due to improper gating also causes the same defect by enhancing surface oxidation.

The second gear blank shows shrinkage in the rim which would be exposed on machining and particularly when teeth are cut in the periphery. The radiographic technique necessary in the examination of this type of casting must necessarily include separate exposures to show details in the hub and rim and in the web or spoke area. Radiographic examination of manganese bronze castings is a rapid, nondestructive means of evaluating the gating system according to the appearance of the dross and shrinkage in the radiograph.



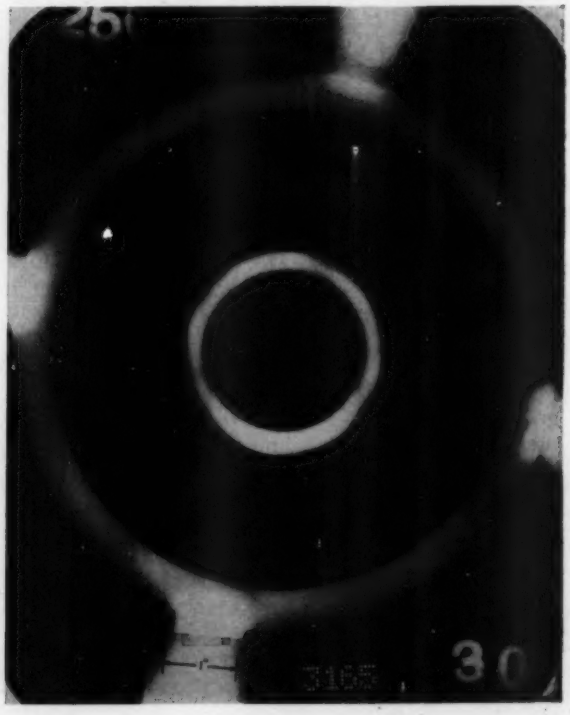


Fig. 23 (Above) — Gammagraph of manganese bronze gear blank showing shrinkage and dross in web; Fig. 24 (Right) — shrinkage in tooth area.

Impregnation of Castings

Foundrymen are occasionally plagued with an epidemic of "weeping" castings. Throughout the years it has been the practice to resort to some form of sealing to achieve salvage. Recently improved impregnation processes involving the use of thermosetting plastics which are introduced into the casting under pressure, or pressure preceded by vacuum and followed by a cure, have found considerable application in the bronze casting industry. While approved processes frequently save castings which show slight leakage or sweating, for their intended service, it is generally accepted that the process requires careful control and that the impregnation medium must satisfy service conditions.

It is also recognized that the impregnation process does not contribute to the structural strength of the affected area. Cognizant of both the usefulness as well as the limitations of the process, the Bureau of Ships has established a definite procedure[19] which must be followed by suppliers of Navy castings prior to impregnation, namely that radiographic inspection of the casting must precede the impregnation and that it show that the leakage and need for impregnation are due to microporosity or minor shrinkage and that the casting is otherwise structurally sound.

Radiographic Standards

The foregoing radiographs are of limited value to the foundryman per se; unless accompanied by a suitable and acceptable "yardstick" for comparison. For optimum usefulness a criterion must be set by which each radiograph may be evaluated in order to establish or predict the effect of the discontinuities noted on the serviceability of the casting.

Such a "yardstick" has been established for steel castings and is in wide acceptance by the industry for over 15 years. Similarly a "yardstick" for the radiographic inspection of aluminum and magnesium castings established by the Bureau of Aeronautics in August 1951 and is in general use by the aircraft industry. The code requirements for fired and unfired steel pressure vessels contain reference radiographs which represent minimum acceptance standards depending on service requirements. A.S.T.M. designation E99-55T[18] represents a set of reference radiographs for welds in steel and encompasses 35 radiographic illustrations of common and uncommon discontinuities in metal arc welds which upon agreement between supplier and consumer become acceptance standards.

It is believed that the availability of such radiographic references or standards has enhanced the usefulness and has contributed to the acceptance of the radiographic process in each particular industry.

The material laboratory of New York Naval Shipyard is now in the process of preparing a set of radiographic standards for tin bronze. These will be utilized by inspectors of naval material and other personnel interested in the quality of castings, to determine the relative fitness of a particular casting to meet specified service requirements and by foundry supervision to control foundry production quality. These standards will comprise the types of defects ordinarily associated with tin bronzes, such as gas porosity, shrinkage, sand inclusions, unfused chaplets, and hot tears.

Each type of discontinuity will be presented as a series of radiographs of graded severity and will be correlated with mechanical properties by comparison with test bars containing similar indications. Figures 25 through 29 illustrate the appearance of representative examples of these several defects as revealed by x-ray examination of ¾-in. thick plate castings in which the defects had been synthesized. As finally set-up the standards will be issued as transparencies for both x-rays and gamma-rays and will include a casting classification table by the use of which it will be a comparatively simple matter for even relatively inexperienced personnel to determine the acceptability of most tin bronze castings for a specific service.

Summary Discussion

The applications of radiography to brass and bronze castings discussed above indicate that this method of non-destructive testing has earned a permanent place in the industry. They indicate that radiography is a relatively easily applied foundry tool which is invaluable in many ways. In design and in production of pilot castings it results in ultimate savings which exceed many times the cost of testing. In quality control its introduction at the earliest stage in fabrication of components results in savings since it can weed out defective parts which more than pay for the cost of testing.

In production salvage it may save castings with defects which would have caused outright rejection by specifications which were in effect prior to the use of radiography and definitely eliminates castings which have too widespread defects to justify attempts at repair. In the salvage of service failures it has frequently saved costly replacements by providing reliable means of following-up defect removal and subsequent check-up of repairs. Its use increases confidence in the judgment of technical people involved in the endorsement of continued use of the salvaged parts. Experience in the other fields where radiography has made its mark has shown that for optimum utilization availability of and agreement on references, or standards, is the controlling factor in its value. This view, held by Juppenlatz[5] for non-destructive tests in general, is especially true of radiography which converts in effect indications contained within a volume of material into a plane projection on the ultimate radiograph. The work currently under way on the development of radiographic standards for bronze casting alloys and the increased availability of fully portable radioactive sources will markedly contribute to the use of radiography as a tool in the brass and bronze casting industry. The old adage that "One picture is worth a thousand words" may be replaced, in this instance, into "one radiograph is worth many cross sections."

Acknowledgment

The authors wish to express their appreciation to the Navy Department Bureau of Ships for sponsoring



Fig. 26 — X-ray radiographic standard for the sand inclusion defect in gunmetal bronze.



Fig. 25 — X-ray radiographic examples of the chaplet defect in gunmetal bronze.

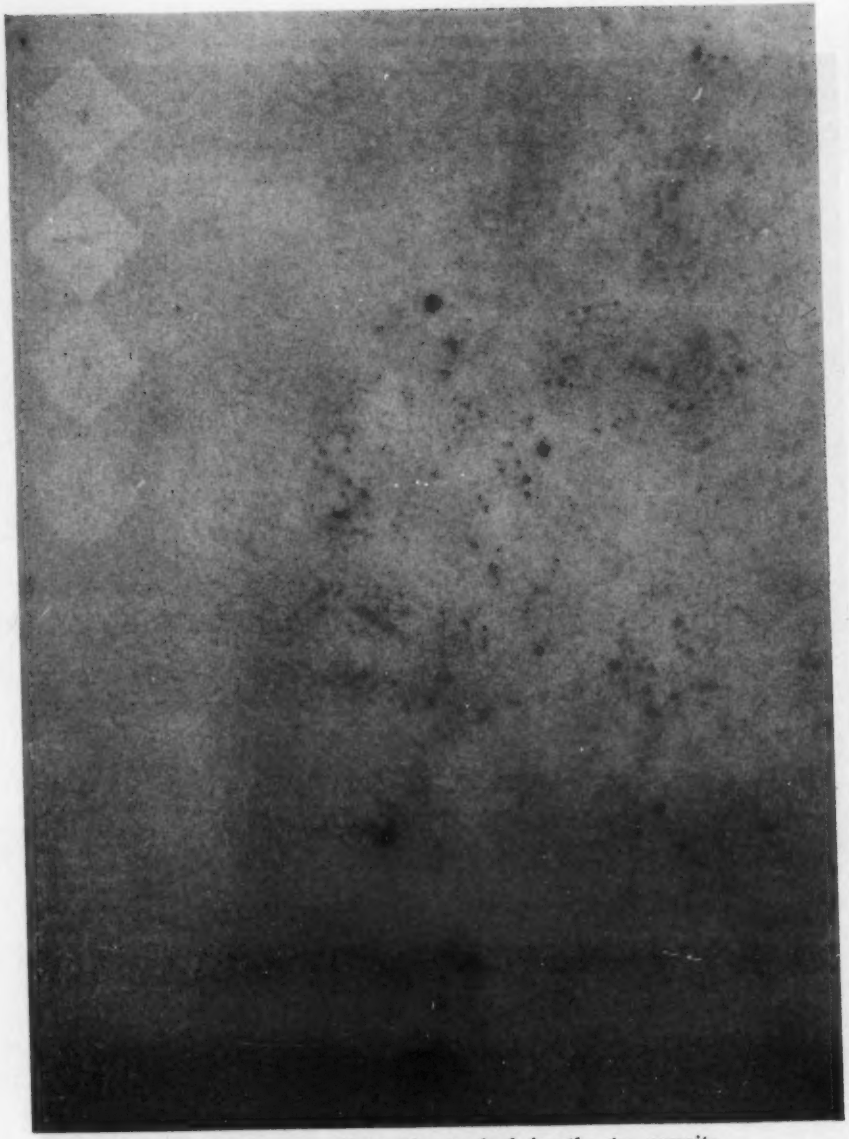


Fig. 27 — X-ray radiographic standard for the gas porosity defect in gunmetal bronze.

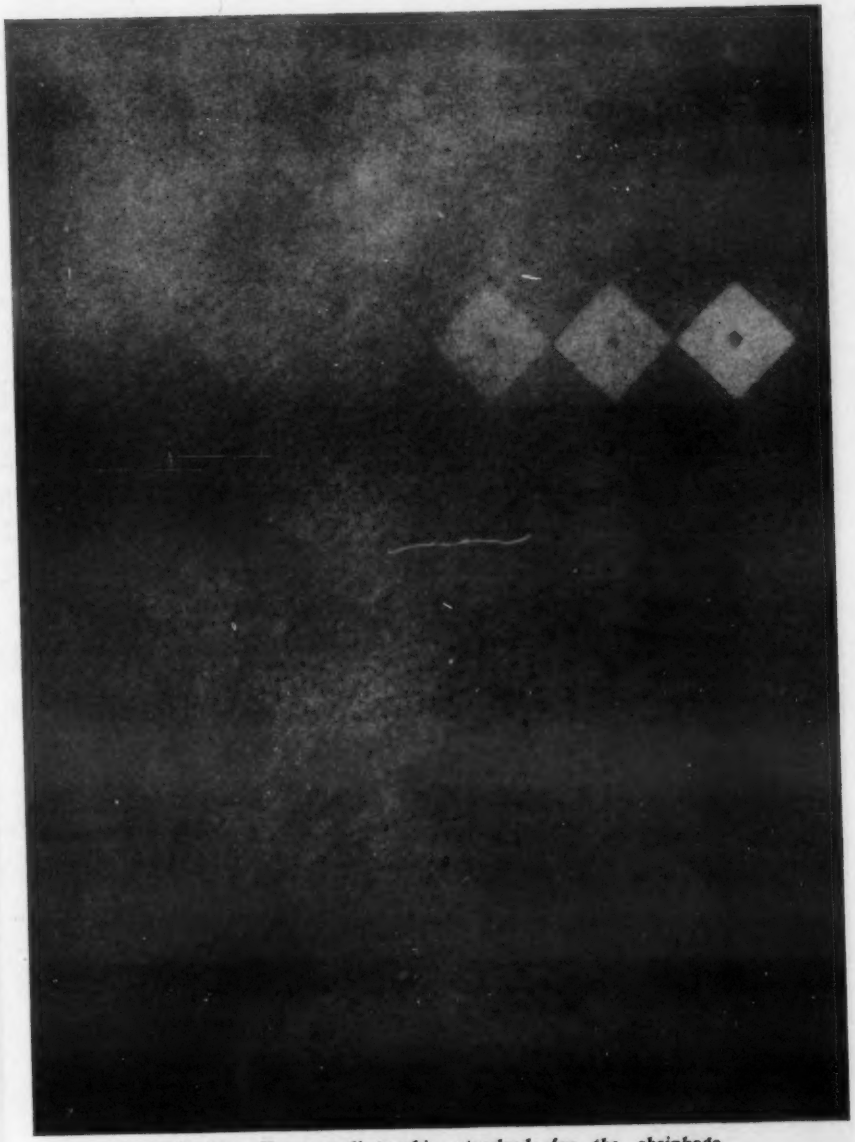


Fig. 28 — X-ray radiographic standard for the shrinkage defect in gunmetal bronze.

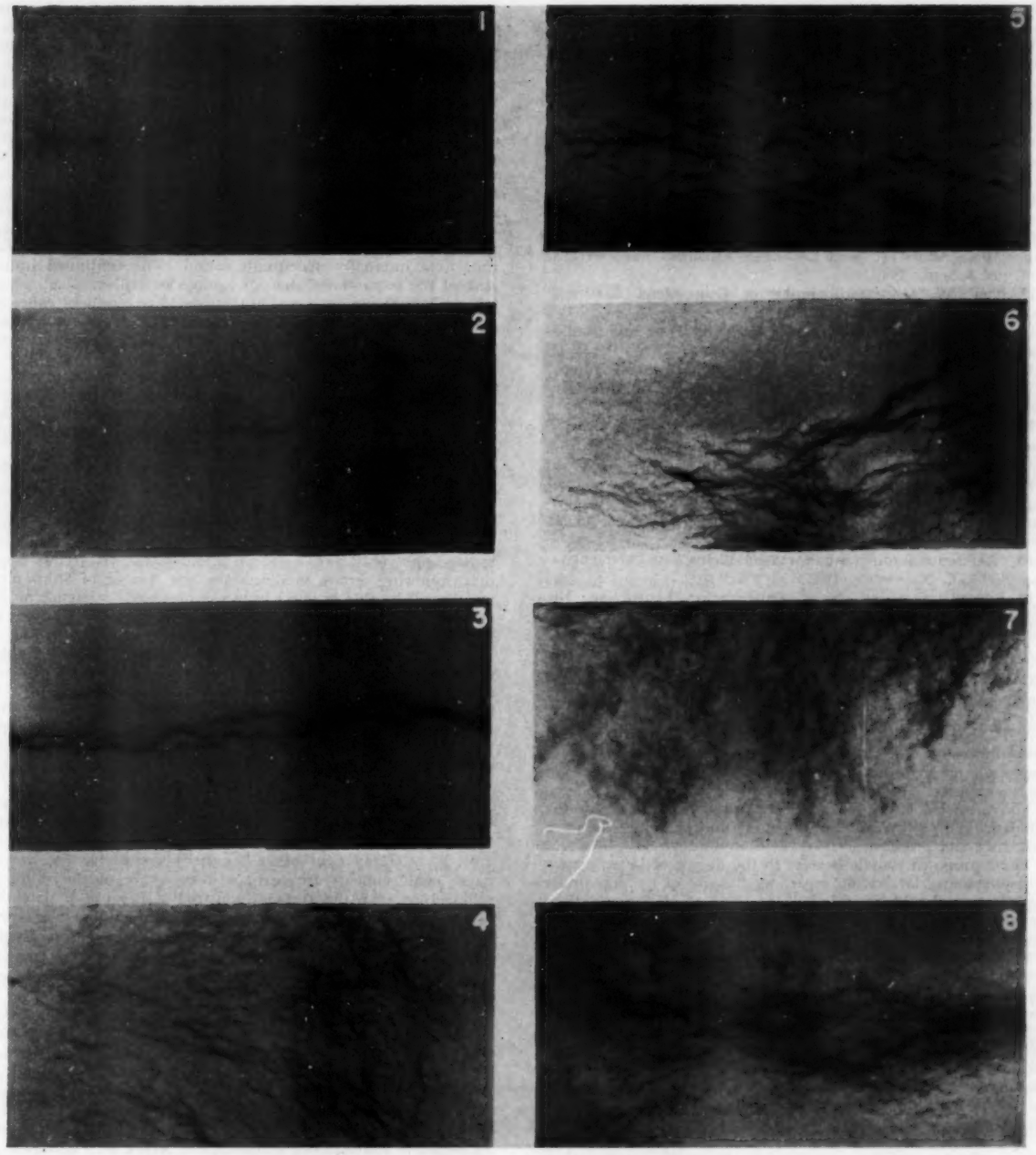


Fig. 29 — Photos of exographs of typical hot tears in tin bronze.

and encouraging the development program which served as a basis for this paper and for permission to publish the results thereof. They are grateful to A. B. Jones, Captain U.S.N., Director of the Material Laboratory and his staff for encouragement in the conduct of the program. Indebtedness is expressed to Shipyard foundry personnel who cooperated with the authors in the studies involving applicability of radiography to bronze casting problems and in the preparation of the castings. The helpful suggestions derived from the earlier work of W. H. Baer and B. N. Ames are also gratefully acknowledged.

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DISCUSSION

Chairman: W. H. BAER, Bureau of Ships, Navy Dept., Washington, D.C.

Co-Chairman: G. H. BRADSHAW, Navy Dept., Philadelphia.

A. P. SPOONER[1] *(Written Discussion):* We wish to compliment the authors on a fine contribution to the development of greater knowledge of conditions encountered in the manufacture of bronze castings.

The proposed standards are a constructive approach to the problem and will lead to a better understanding between purchaser and producer.

These proposed standards refer to the degree of imperfection. Probably some clarification regarding the size of castings under investigation should be included. In the case of larger castings we assume that the degree of imperfection rather than the extent is the governing factor in classifying the castings.

Probably some short reference to rules for safe handling and precaution necessary for handling the energy source should be included as a warning to those not familiar with this method of testing.

J. F. KLEMENT[2] *(Written Discussion):* The authors have done well in illustrating the practicability of the radiographic tool in the bronze foundry. The increasing costs of operating a bronze foundry, coupled with the desire to remain competitive, should be reason enough to use methods such as these to aid the foundrymen.

The reference to radiographic standards for bronze casting alloys should be of particular interest inasmuch as these will help bring the producer and user of cast bronze products closer together on the subject of what is desirable and necessary for a particular application.

WM. H. BAER[3]* *(Written Discussion):* The authors are to be congratulated on the manner in which they have presented the data contained in this paper.

The Navy Department has advocated the use of radiography on castings for many years. When the "Radiographic Standards for Steel Castings" were issued by the Bureau of Ships in 1942, there was considerable confusion concerning the application and interpretation of the use of the "Radiographic Standards".

After continued use of the "Steel Standards" for 14 years, manufacturers have become accustomed to the Navy's radiographic requirements for steel castings and currently are producing castings that meet these requirements.

In 1953, the Bureau of Ships recognized that some control was necessary with regard to the salvage of bronze pressure castings by impregnation. In May of 1953 a Bureau of Ships Instruction on "Impregnation of Porosity in Copper-Base Alloy Castings" was issued. In this instruction, the use of radiography combined with the "Pilot Casting" technique is recommended. It was considered that the only logical means of determining whether a pressure casting that leaked when subjected to a hydrostatic test, was satisfactory for salvage by impregnation, was to make it mandatory that the castings be radiographed to determine that they were internally structurally sound. The continued insistence of the requirement that all castings be radiographed, when salvage by impregnation was anticipated, has produced concrete evidence that the bronze pressure castings currently produced for Naval use, have improved considerably in quality in a short period of three years.

The Bureau of Ships has sponsored many research programs in Government laboratories in attempting to develop satisfactory and realistic radiographic standards for bronze castings. The authors of this paper have cited publications concerned with radiography of bronze castings covering the past ten years.

The continued research at the Material Laboratory of the New York Naval Shipyard, has recently culminated in a report illustrating gradations of five types of casting defects, with respect to their effect on the structural soundness of the casting. At the present time, these radiographs are classed as "reference radiographs" only and have no official status. It is our hope that, following certain modifications, the Bureau of Ships ultimately will be able to select from the proposed "reference radiographs", the illustrations that will ultimately be issued as "Radiographic Standards for Bronze Castings."

It is my impression, based on past experience, that radiography of bronze castings has a definite part in the production of satisfactory castings. The continued use of radiography will provide knowledge of inestimable value to the foundrymen. Also, the use of radiography has saved many hours and considerable expense with respect to machining a defective casting that would ultimately be rejected as unsatisfactory. Radiography is also very helpful in final inspection and salvage if required.

The continued use of radiography of castings will increase our knowledge in many ways, that will be reflected in consistent improvement in the quality of bronze castings.

R. R. WALTIEN *(Authors' Closure):* Those of us who are in naval establishments are perhaps more aware of the value of radiography than the average industrialist. Because of potential consequences of failure we must take every precaution to guarantee serviceability of castings used on Naval vessels. Accordingly, all castings for critical application and most other castings are subjected to radiographic examination at some stage in their manufacture to insure soundness.

The primary value of radiography to the foundryman is not in the inspection of finished castings, but rather in its utilization as a tool to economically establish foundry techniques which will produce sound castings. The "pilot casting" method as practiced by the foundry at the New York Naval Shipyard saves considerable time and money in production of complex castings. The first casting made from a new pattern is subjected to complete radiographic examination. Based on evaluation of the radiographs, changes are made in the foundry technique to eliminate areas of shrinkage etc. A second pilot casting is then made and radiographed. This is repeated until a satisfactory casting is achieved, after which the full requirement of castings is produced. While this method seems at first glance to be rather time-consuming, in the long run it is economical since experience has shown that two or three pilot castings are usually adequate to establish the foundry technique. Thus the machining and return of unsound castings is avoided. Also, inspection of production castings can be minimized, usually only critical areas are examined and sometimes only a percentage of the castings are inspected, according to the size of the lot.

The advent of nuclear energy for industrial application, which is imminent, will require closer foundry quality control than heretofore. The high pressures and temperatures involved and the potential radiation hazard will call for heavy castings of almost absolute soundness. The foundryman will do well to consider carefully the potential value of radiography to insure the quality of his castings for this application.

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*NOTE: The opinions or assertions contained herein are the private ones of the author and are not to be construed as official or reflecting the views of the Navy Department or the naval service at large (Article 1252, Navy Regulations).

BASIC COST CONCEPTS FOR THE SMALL-MEDIUM FOUNDRY

By

Ashley C. Sinnett*

During 1955 forty foundries went out of business due to bankruptcy and liabilities amounted to $2,466,000.00. To state it briefly, three causes are responsible for this high mortality; inexperience in technical operations, short working capital and lack of knowledge of operational data. Dun and Bradstreet in their report, "Why Business Fail," for 1955, listed 91.4 per cent as due to inexperience.

Due to the fact that there is such a broad diversity in the size of foundries, the abstract analysis of "Basic Cost Concepts" is of considerable importance to our industry. Like governmental bureaus, it is possible to develop reports, charts, schedules, budgets and an almost unlimited analysis of operating activity to such an extent that confusion is the result.

Need for Operational Reports by Management

For large single or multiple operations where executives are remote from contact with production activities, it is necessary to keep them informed of all phases of operations so that decisions can be made and company policies developed. In order to do this, they must have a clear picture of operations and progress to substantiate their activities. Therefore, budgets, productivity reports, cost analyses and marketing data are required; also, personnel with special training and ability which means overhead to be absorbed that most small producers cannot afford.

Our industry is primarily made up of small operations, foundries employing less than 100 persons. Eighty-one per cent of approximately 5,400 foundries in the United States employ less than 100 persons. In most instances, the executives are also the individuals directly responsible for production operations and sales. It is to this group that this paper is directly addressed, although a review of cost accounting principles might also benefit the representatives from the larger companies. In particular, this paper will endeavor to guide those who need a cost system or believe they need improvement in the one presently being used. However, this paper does not present a system but an analysis of what should be considered in developing one for your individual requirements. Every cost procedure *must* be tailor made.

Results of Cost System

Let us consider the question "What should my cost system accomplish?" The end point of a cost system should be a formula that will develop a cost, per unit of weight or per piece, of a casting from an individual pattern that reflects a sound basis for determination of its selling price. We must abandon the obsolete thinking that we are selling tons of castings and replace it with the knowledge of how many dollars of labor, conversion materials, taxes, insurance, depreciation and other overhead costs we are selling. In order to do this, it is necessary to know what cost activities are required to produce the casting and the cost of each activity. Let us consider the following definitions:

Cost Centers are the points of concentrated effort necessary to produce castings;
Cost Elements are the costs of the activity of each Cost Center, and the
Cost Formula reflects the unit results.

It is recognized that it is not practical to determine the exact cost of a given casting. All castings being produced from different patterns at the same time have an influence on the production cost of any given casting in the group or pattern mix. Any change by addition or deletion of patterns in the pattern mix will change the actual cost of the individual castings. However, the Cost Formula, properly developed and applied, will reflect a reasonably accurate cost if there is not a drastic change in pattern mix between the period of basic calculations and time of production. Any major change that affects the operations will require a new analysis of costs. This would include pattern mix, engineering, labor rate changes, overhead cost and increased cost of purchased materials and supplies.

In the development of a cost formula, an analysis should first be made of the plant production methods. This will reflect that all castings require certain facil-

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ities and in some instances only particular facilities. Note in particular your Core Room. You may have floor blowers, bench blowers and hand operations as direct labor centers. If so, there would be three cost centers. Sand preparation and baking could be an additional cost center, but if desirable from cost control analysis, it could be two cost centers. The centers mentioned are not necessarily required. A close analysis of your type of production and relative costs within the department must be considered in deciding how many Cost Centers are necessary for accurate cost findings.

This same procedure should be followed throughout your plant. Remember: Different processes; separate cost centers; different process methods; additional cost centers. The final result places you in a position to set up a chart of accounts that will channel costs of elements into groups whereby they can be summarized into cost centers. In accomplishing this, due consideration must be given service departments and expenses that are general in character. General expenses should be separated between expenses that are fixed and variable; also, those caused by payroll or other activities. Each of the general classifications would be considered a cost center.

Account Numbering System Illustrated

The following illustrates an account numbering system that standardizes classes of cost elements:

CLASSIFYING ACCOUNTS FOR FOUNDRIES

Chart Scheme		Example	
Department	1000	Department two—	
	2000	Molding	2000
	3000	Cost center one—	
	etc.	Sand preparation and supply	100
Cost Centers	100	Overtime premium	32
	200	Account number	2132
	300	Department three—Core	3000
	etc.	Cost center three—	
Direct labor	10	Blower	300
Indirect labor	20	Direct labor—	
Delay time	30	Painting cores	11
Allowances	31	Account number	3311
Overtime premium	32	Department three—Core	3000
Shift differential	33	Cost center four—	
	etc.	Baking	400
Supervision and clerical	40	Maintenance supplies	70
Supplies	50	Account number	3470
Maintenance labor	60		
Maintenance supplies	70		

In addition to an accounting procedure to funnel expenses to the proper cost center, statistical information must also be accumulated to reflect activity of production. An analysis of the cause of cost is necessary to determine what the basic cost units should be. The molding department probably would reflect as a minimum the following: mold sand weight, mold metal weight and molding direct labor dollars. The method of calculating payroll must be considered when making this analysis. Also, the location of the work accomplished.

Some foundries remove sprue from castings in the Molding Department whereas others have a separate department for this operation. Other controlling factors which must be considered and are of particular importance, are the weight of cores per casting, weight of rough or prefinished castings and weight of finished castings in saleable state. The objective of this determination is to have a vehicle by which costs can be applied to specific jobs for only those expenses incurred in its production.

There has been considerable discussion regarding allocation of Metal Costs to specific jobs because of variations in Yield. Many hold to the group using the so-called Yield Table. The author does not disagree with them insofar as their claim to accuracy is concerned. However, he believes that it is an unnecessary calculation and the same results can be obtained by a single calculation whereby the variation is absorbed in the Cost Unit based on Mold Metal Weight.

The author believes the theory is sound that metal costs the same per weight unit in every casting made from the same metal composition regardless of yield. The cost variation due to Yield is costed to the particular job based on metal in the mold. This is accomplished by calculating the cost of Metals Used and arriving at a unit price of given dollars per unit. Using this price, a value is placed on Saleable castings produced. The difference between the metal value of saleable castings produced and cost of metals used is the variable and becomes part of the cost distributed on the basis of Mold Metal Weight. This variable which includes the cost of melting is definitely a part of the conversion cost, and inasmuch as we are deeply interested in the cost of converting metals to castings, this portion must be considered in this manner. This principle is the one difference between the author's recommended method and that of developing a yield table.

Distribution of General Expenses

Distribution of General Expenses is of considerable importance. The total dollar costs of this group is generally large and care should be taken in properly directing the values to the cost centers used for formula purposes. The author will describe some of the costs to be considered and a recommended method for disposition.

Payroll costs such as labor taxes and labor insurance require an allocation to cost centers based on labor charged direct. This distribution must include all cost centers that have accumulated charges for labor expense.

Maintenance costs are sometimes difficult to determine in small foundries, especially if Direct Labor operators are permitted to make minor repairs. The Chart of Accounts should provide for maintenance costs, both for labor and supplies, for each cost center. Proper distribution will then account for these costs insofar as direct charges are concerned. However, some maintenance costs are general in nature such as the Maintenance Supervisor, Building Repairs, etc. These should be allocated or distributed to Cost Centers on an applicable basis. This could be maintenance charged direct. Floor area or whatever basis that seems justifiable.

Depreciation, if at all possible, must be charged to

each cost center based on Buildings and Equipment located in or required by the cost center. In relatively recent years, consideration has been given to using depreciation rates applied to replacement values for cost accounting purposes. This is due to the deflation of the purchasing dollar and the realization that depreciation reserves accumulated in the past are insufficient to cover replacements in the present or future. It is true, that unless this theory is followed, and the dollar continues to be deflated, replacement of your equipment will be partially paid out of profits. Also, the older your equipment is, the greater the demand will be on your profits for this purpose.

If an appraisal of buildings and equipment is maintained currently, the replacement value is readily available. If an additional source is necessary, the author found that most insurance companies can supply a chart or schedule reflecting periodic changes in values so that the relationship between the origination and the present can be determined. This, of course, can be obtained from various appraisal companies also, and possibly your trade association is providing its members with a schedule as is being done by the Malleable Founders' Society.

General Plant Overhead costs that cannot be allocated to specific cost centers by a controlling factor, should be distributed on the basis of total costs accumulated by cost centers.

Administrative and Selling Expenses, should be applied as a Cost Formula rate on the cost of converting metals to castings, that is, the total cost of production minus Metal Cost. This follows the theory that foundries today are selling a service; the ability to convert metals to castings according to customers specifications.

The actual cost to produce a given casting does not vary due to variations in volume of total production. Calculations will reflect that the cost *does* vary with production volume unless overhead expenses are applied in the development of the Cost Formula on a normalized basis.

Concept of Normal Overhead

The basic concept of normal overhead is that the actual cost is not necessarily the best criterion of the true cost, but that the proper amount of overhead expense to be charged in your formula is the amount incurred at normal capacity. Two elements are thus involved:

1. Normal production, involving a determination of normal capacity.
2. Normal expense, involving a computation of normal burden rates. By normal expense, it is meant that if accruals or prepaid accounts are not a part of your general accounting procedure, then when examining the expenditures for the period that the Cost Formula is to be based on, any expenses that are abnormal should be adjusted, such as vacation expenses, prepaid insurance, real estate and personnel property taxes or any other that are occasional or periodic in nature, and not particularly applicable to the period involved.

Presuming a six month period is to be used in the development of a Cost Formula, the following analysis is presented as a theoretical normalizing factor:

NORMAL OPERATING HOURS

Days per year		365
Less: Saturday and Sunday	104	
Holidays	6	
Vacation	5	115
Maximum working days		250
Standard Work-day (Hours)		8
Maximum work-hours		2000
Operational down-time	10%	
Major equipment maintenance	2%	240
Net annual Scheduled Hours		1760

Example

50 Molding Stations @ 1760 hours	88,000	hours
6 Months Basis	44,000	
Actual Hours worked	40,000	
Normalizing Factor	91%	

The results indicate that only 91 per cent of fixed Overhead Costs should be absorbed in the Formula in order that calculated unit costs will be normal. Contra to the above, if the calculation reflects 105 per cent, fixed Overhead Costs would be increased and your cost savings would be retained instead of passing them to your customer by undercosting your production.

It should be understood before accepting this theory, that the calculation can give whatever result management desires. The illustration indicates that 50 molding stations were used as a normal operation. It does not necessarily mean that only 50 stations were available. Management must determine how many stations would be an average or normal operation from which they would expect to make a satisfactory profit.

Other important operational activities that must be considered in arriving at an accurate basis for costing are:

1. Castings produced for own use.
2. Skim cores and furnace cores.
3. Green sand cores.
4. Returned castings, scrapped and salvaged.
5. Order charge (short orders, production runs, repetitive production runs from single orders).

Departmental Cost Factors Established

When we work up costs applicable to the centers of operation within departments, we have prepared the way for sound costs. This step is most needed in highly mechanized production shops but is also necessary in the small conventional jobbing shop. For both, it will provide tailored castings costs which include only elements required for production in the case of each pattern and will exclude costs for facilities not used.

The procedure suggestions made are applicable to the entire foundry industry. Certainly a knowledge of true costs, determined realistically, would be beneficial to all foundries, regardless of the metal used. Competition would be stimulated by confidence in costs and loss price-setting would be minimized.

The procedure opens up a field for cost control of which few foundries take advantage. Knowledge of cost of operation of production centers developed on a comparative monthly (or more frequent) basis will give management tools to work with. If a chart of accounts is developed to properly record classifications of labor and supplies, the suggestions which have been made are relatively simple to embody in a procedure. It also provides the groundwork for flexible budgets covering all indirect expenses, both labor and supplies.

In the smaller foundries perhaps the easiest to develop and maintain are graph charts on which would be plotted man-hours per ton by departments and scrap percentages for departmental controls. These figures would be a guide to the supervision as to their progress and when a second year is available, a comparison with prior years on the same chart proves invaluable. Along with this, of course, it is required to properly interpret them and in particular, man-hours per ton is affected by the average weight of castings. So, if this is available, it should also be plotted in order to be a guide for the proper reading of the man-hours reports.

Here again, the author emphasizes that the foregoing reflects basic principles of Cost Accounting that should be considered when developing a cost procedure and cost formula. It is intended to high light certain necessary data for accurate cost findings.

DEVELOPMENT OF HOT-BLAST CUPOLA MELTING TECHNIQUE IN EUROPE

By

Ernst Loebbecke*

Historical and Fundamentals

When in 1832 Faber du Faure operated his cupola for the first time with hot blast, he would hardly have suspected that he would become the father of a new melting technique, the suitability of which would still be strongly argued a hundred years later among outstanding experts. The long path of development of the cupola with preheated blast, accompanied by many setbacks, is well known from international literature so that it will not be necessary here to refer to the many promising and also sometimes fantastic attempts of this first stage.

The real history of the modern hot-blast cupola begins after the first world war when foundrymen like F. K. Vial[1] with remarkable foresight and for the first time clearly recognized the metallurgical possibilities of melting with preheated blast, while the majority of experts still considered coke savings as the most important economic possibility.

In spite of this it required a second period of distressed conditions in Europe (World War 2) to establish with certainty the concept of the hot-blast cupola. Based, among others, on the valuable pioneering work of E. Piwowarsky,[2] the installation and start of operation of the first cupola plant with hot blast over 752 F (400 C) at Georg Fischer A. G., Schaffhausen, Switzerland[3] in 1941, was the beginning of that turbulent evolution from which the modern European hot-blast cupolas in their various forms have derived.

It would be far beyond the scope of one report to make reference to the whole of the literature of this last period of development or to subject it to a critical consideration. It can only be summarized here that this development as a whole was a model of international co-operation, although the individual experts concerned may not have realized this at all times.

The interesting results of experiments made by W. W. Levi,[4] S. F. Carter[5] – just to recall a few – were read and discussed in Europe with the same attention and interest as were the valuable leads and operating experiences in the area of refractory materials for cupolas published by J. P. Holt,[6] E. Bremer,[7] T. J. Wilde, M. R. Hatfield and V. J. Nolan,[8] P. Provias,[9] and others.

In Great Britain E. S. Renshaw[10] and also the sub-committee of the Institute of British Foundrymen[11] under his guidance have done remarkable pioneer work in the field of the hot-blast cupola and its metallurgy. D. Fleming[12] began to employ thermodynamics to clarify the metallurgical possibilities of melting with hot blast and has thereby opened new avenues of research and understanding. France, with its pioneer of the smelting gas producer M. H. Philipon,[13] with Marcel Guédras[14] and others, Belgium with R. Doat,[15] De Bock[16] and A. De Sy,[17] Sweden with B. Thyberg,[18] J. Drachmann,[19] Italy with C. Bortolani, W. Demicheli and C. Longaretti,[20] Spain,[21], Japan,[22] Hungary,[23] and Russia[24] – they all contributed to unfold the picture of the present day hot-blast cupola.

The number of publications in Germany, the birthplace of the European hot-blast cupola, exceeds those of all other non-American countries, so that we shall mention here only the work of E. Piwowarsky[25] and H. Jungbluth[26] who formed the foundation of modern European cupola technique. There will be opportunities later to refer to other valuable German publications.

For a better understanding of the subsequent material it may be advisable first to mention the differing *opinions regarding the hot-blast cupola* which played a decisive role in the current development and helped to form its external aspect. It was mentioned previously that the first European cupola with hot blast over 750 F (400 C) was erected during the war, when it became more and more difficult for Switzerland to obtain the high-quality foundry coke from the Ruhr.

This hot-blast cupola was, therefore, only intended at that time to save coke by the use of blast preheating, and it fully met this requirement. Experience showed that with the use of hot blast the

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melting process in the cupola had altered and that there were unexpected deviations in the chemical composition and in the quality of the metal. However, all these things were gladly tolerated in exchange for the great advantages derived from the coke savings. Only a very few at that time began to realize the metallurgical possibilities which have today come to make the hot-blast cupola so valuable and attractive.

Naturally, the further evolution of designs and melting techniques during the immediate following years was primarily influenced by thermotechnical considerations. The still frequently used "Esslingen" cupola of the Vereinigte Economizer Werke is a typical example of this trend. This particular system takes the major portion of the stack gases off the cupola just above the combustion zone at a point where it is believed the carbon dioxide in the gases has not yet been partly reduced to carbon monoxide. The very hot gases are then led to a cast iron recuperator for preheating the blast, utilizing the sensible heat.

In this system, conducted without preheating the charged materials in the cupola stack, the melting process will inevitably take place in an oxidizing cupola atmosphere and is different from the process in the cold blast cupola only with regard to the higher temperatures in the melting zone due to a narrower concentration of this zone by the use of hot blast.[27]

The expected savings in coke realizable by this process by virtue of eliminating the reducing reaction $CO_2 + C \rightarrow 2\,CO$ can be readily negated by other losses, such as loss in silicon.[28,29] Despite these disadvantages, this type of furnace has considerably contributed to the introduction and spreading of the hot-blast cupola.

The same applies to plants which, after the model of the first unit at Georg Fischer, Schaffhausen, were equipped with *steel recuperators* "System Schack" and from which springs the type of hot-blast cupola now developed by Gesellschaft fur Hüttenwerksanlagen. In this system the cupola gases are taken just above or—and this is more normal —just below the upper level of the charged materials in the stack and are utilized for blast preheating with the use of both the sensible and the latent heat.

This system has proved to be very efficient for smaller and medium-size plants of melting rates of up to 12 tons per hour as long as fluctuations in the hot-blast temperature could be tolerated and the melting periods were not too long. Both types of hot-blast cupolas described so far constituted—at least in the beginning—the thermotechnical group, i.e. to the group which tried to make the cupola more economical from the thermotechnical angle. Originally, this development was essentially carried on by thermotechnical experts and designers.

Another line was followed by R. Doat and his colleagues[15,16] with their *Metallurgical Blast Cupola* (M.B.C.) which originated in France[13] and was further developed in Belgium. R. Doat made no attempt to reduce the fuel consumption in his type of cupola but actually increased it over that of the cold-blast cupola in order to produce metallurgical effects which would compensate for the additional consumption of fuel. Details of his melting process are well known from various publications so that it will not be necessary to give further descriptions.

One might point out in this connection, however, that the adjective "metallurgical" for his type of cupola can only be understood to depict his adverse views to the merely "thermotechnical" trend which was described earlier. *Any cupola is, of course, a metallurgical cupola, even the cold-blast cupola!* Each and any alteration in the chemical composition in the charged materials, for instance, of silicon losses in the cold-blast cupola or of pick-up in the hot-blast cupola of any system, is a metallurgical process irrespective of whether the course it takes is in the desired or undesired direction.

A more exact description of the processes and designs mentioned so far will probably not be necessary in view of the excellent reports of these installations in American and European publications. One might point out in particular the reports by G. P. Phillips,[30] B. Thyberg,[18] F. Schulte,[31] F. C. Evans,[32], G. J. Shaw,[33] and others.[34]

The fourth type of hot-blast cupola, which also belongs to the "metallurgical" group and which is frequently used in Europe particularly in large plants of up to 30 tons per hour melting rates, has not often been mentioned[35] as yet in general surveys.

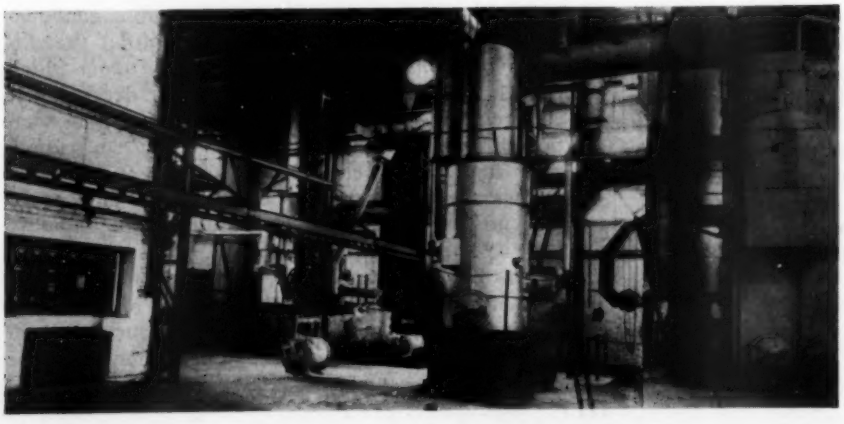


Fig. 1 — View of hot-blast cupola research plant (works photograph Buderus'sche Eisenwerke, Wetzlar).

Without appearing to propagandize this system the writer might therefore be permitted to go into somewhat closer detail of this type, the more so because he took an active part in the development of this system and because the investigations referred to later in this paper were essentially made with hot-blast cupolas.

During long years of conscientious work this melting unit was developed into a *compact melting aggregate* in which the cupola in its various designs holds the center and where the mechanical-thermotechnical part of producing hot blast is intended solely to influence the *metallurgical* processes readily and reliably and to *guide* them into a desired direction which can, to a certain extent, be calculated in advance.

To this end it was necessary to place certain demands both on the cupola and on the production of hot blast, which can only be fully appreciated from the knowledge of the metallurgical processes in the hot-blast cupola. In the beginning of this development, however, there was nothing but the metallurgical idea and, resulting from it, the general trend. The actual knowledge of what was happening and what could happen metallurgically in the hot-blast cupola was very minute.

The insight had to be gained by extended research work. A suggestion made by Strico was taken up by several foundries and finally led to the formation of an international Research Group for the Hot-Blast Cupola which under the guidance of Buderus'sche Eisenwerke in Germany was joined by various British, German and Swiss firms. A scientific working committee under Professor Dr. H. Jungbluth was entrusted with fundamental metallurgical research work to which Buderus gave the necessary melting unit, a hot-blast cupola system with an I.D. of 24 in.

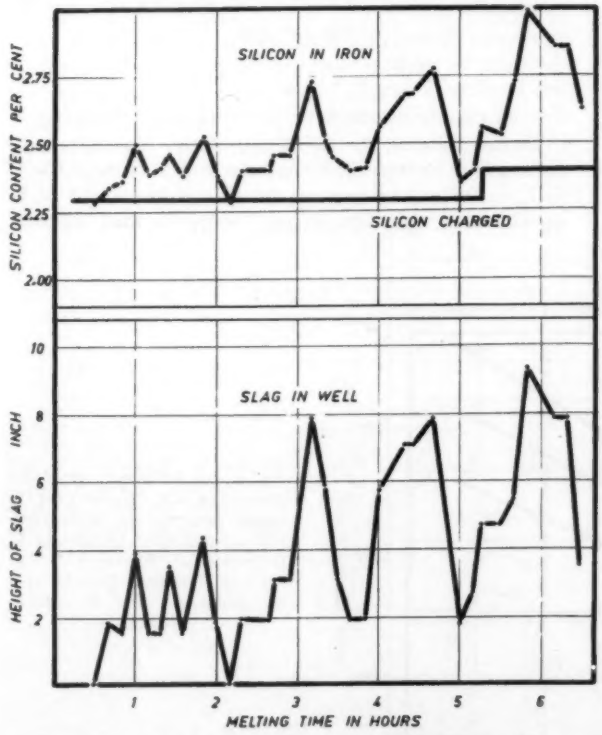


Fig. 2 — Influence of height of slag in the well on silicon content of cast iron (according to H. Jungbluth and K. Stockkamp).

By cordial consent of the member firms, part of the fundamental metallurgical findings to be reported here were taken from this work of the research group for the hot-blast cupola, to the working committee of which the author himself belongs, and part also from a publication by H. Jungbluth and K. Stockkamp[36] which was based on this work. Meanwhile this research group has at its disposal a modern research installation at Buderus'sche Eisenwerke Wetzlar (Fig. 1) in which it is possible to conduct every desired metallurgical or process investigation. This not only serves the further development of hot-blast plants, but also advances the hot-blast cupola melting technique.

Now what requirements have to be asked of a modern hot-blast cupola melting unit and what are the metallurgical findings and fundamentals on which such requirements are based?

Figure 2 shows the result of a pilot test in an acid-lined hot-blast cupola with intermittent tapping, from which the influence of the height of slag in the well on the silicon content of the molten metal is particularly apparent. The silicon content of the iron goes up when the height of the slag level rises while, for the time being, it is basically unimportant whether as an end result there is a reduction in silicon loss or, as in this particular case, even a silicon pick-up.

On the basis of this observation the first important conclusion may be drawn. To guide and control the metallurgical process in the cupola it is necessary to have a closely defined height of slag in the well at all times, and it must also be possible if necessary, to alter this height. This requirement necessarily leads to the *use of siphons at the tapping hole* with continuous removal of slag and iron.[37] Figure 3 shows one of the designs normally used in Europe. All further experiments were, therefore, made with a cupola provided with such siphon.

It was already known from the cold-blast cupola and was at various occasions illustrated by H. Jungbluth[26] that the molten metal temperature can be raised by increasing the coke charge. It is also well known that the use of preheated blast increases the theoretical combustion temperatures of coke as is shown in Table 1. This elevation in the combustion temperature and the concentrated delivery of heat, which, among others, O. Mattern[35] has demonstrated to be a result of the concentrated combustion zone due to the hot blast, was bound to result in an increase of metal temperature.

TABLE 1

Blast Temperature F	Blast Temperature C	Combustion Temperature F	Combustion Temperature C
68	20	3488	1922
600	316	3943	2173
700	371	4023	2217
800	427	4104	2262
900	482	4183	2306

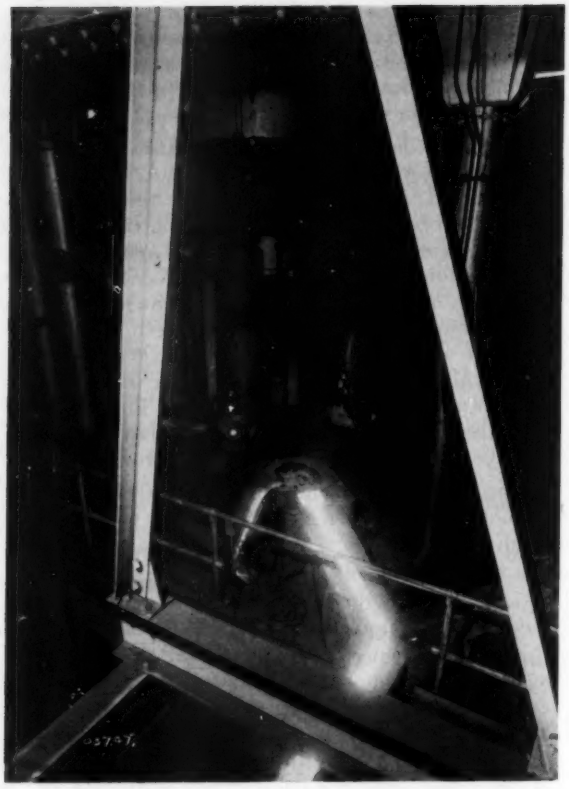


Fig. 3 — Hot-blast cupola with continuous iron and slag tapping (works photograph Platts (Barton) Ltd., Great Britain).

Results of experiments shown in Fig. 4 which are the average values of many single experiments show graphically the interrelations between metal temperature and blast temperature which apply to the hot-blast cupola. It follows that the influence of the blast temperature considerably exceeds the influence of the amount of heat brought in by the blast. If, for instance, an increase in the coke charge with cold blast from 1:11 to 1:7.7 results in an increase of the metal temperature of only approximately 20 F (11 C), the increase in the blast temperature to 932 F with a coke charge of 1:11 brings about a temperature increase in the metal of approximately 158 F (88 C), and with a coke charge of 1:7.7 as much as 229 F (127 C).

This increase in the metal temperature and the corresponding increase in the slag temperature facilitate metallurgical reactions in the hot-blast cupola which cannot be expected in any cold-blast cupola. One of these reactions most important to the ironfounder is the *alteration of the silicon content* which was shown to be possible in Fig. 2. It became the task of further experimentation, therefore to determine those factors which are responsible for the alteration of the silicon contents in the cupola and to measure these quantitatively. This part of the research work is going to take a considerable additional time because reliable data can only be obtained from a great number of single experimental melts.

According to the theoretical calculations by Fr. Körber and W. Oelsen,[38] R. W. Heine,[39] and also by H. Jungbluth and K. Stockkamp[36] it is possible to reduce silica in the slag provided the temperature is high enough.[40] A low Si concentration or respectively a high SiO_2 concentration favor this process so that the reduction to silicon will be higher the more acid the slag is. High carbon monoxide contents in the cupola gases, i.e. a reducing cupola atmosphere also favor the reduction of silicon from the slag as was already shown by D. Fleming.[12] All these prerequisites are available in the hot-blast cupola. It is generally acknowledged that the use of hot blast, exactly like an increase in the fuel charge, decreases the combustion ratio

$$\eta_v = \frac{100 \times \% \, CO_2}{\% \, CO + \% \, CO_2} \, \% \, ,$$

i.e. the cupola atmosphere becomes more reducing.[26]

Figure 5 shows the influence of the blast temperature, using various coke contents and a ratio of CaO: SiO_2 = 0.7 in the slag, on the silicon losses and pick-up in the above-mentioned experimental cupola,

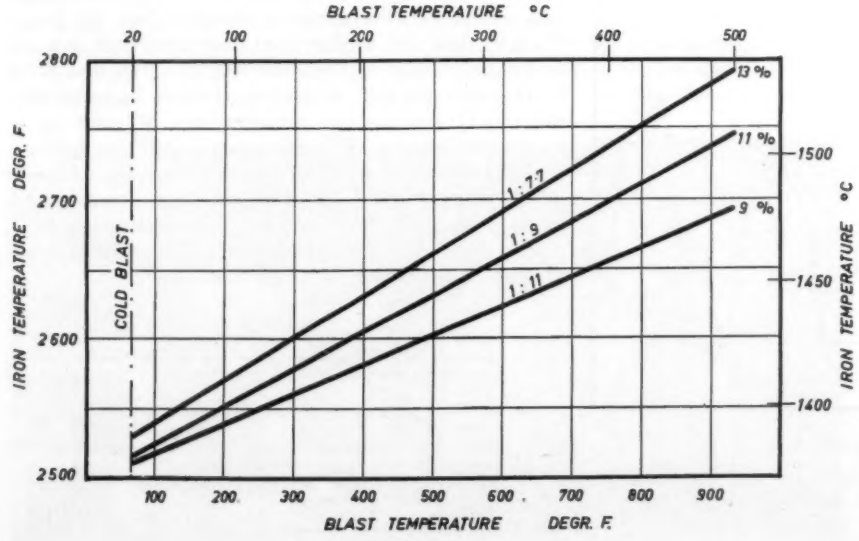


Fig. 4 — Influence of blast preheating on the increase in metal temperature with various coke charges (according to International Research Group for the Hot-Blast Cupola, and H. Jungbluth and K. Stockkamp).

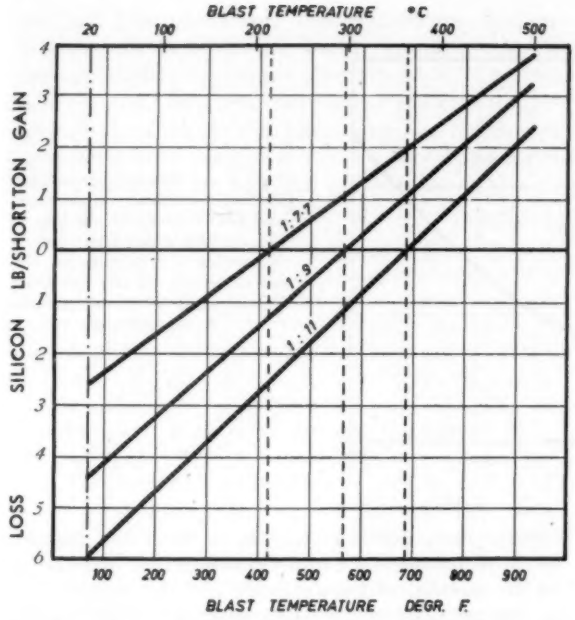


Fig. 5 — Influence of blast temperature on silicon loss and pick-up in relation to coke charge (according to International Research Group for the Hot-Blast Cupola, and H. Jungbluth and K. Stockkamp).

insofar as this can be shown from the experiments made to date. It follows from these results that, with a coke charge of 1:11, a blast temperature of over 680 F (360 C) is required to avoid silicon losses while with a coke charge of 1:7.7 it is necessary to preheat the blast merely to 410 F (210 C).

Another question which is of considerable interest to the ironfounder is the *sulphur content* of the molten metal. It was to be expected that as a result of the smaller coke charge in the hot-blast cupola the sulphur pick-up would also be lower than in the cold-blast cupola. This was consistently proved in practical experience. Careful observations of hot-blast cupolas soon showed, however, that the sulphur content of the cast iron did not come down proportionately to the extent expected from the reduction in the coke charge but decreased many times greater.

E. Piwowarsky and F. Hage[29] in an investigation of a hot-blast cupola system V.E.W.-Esslingen found the relations of sulphur contents in the metal, in the slag and in the cupola gases which are shown in Fig. 6. The illustration demonstrates that even the acid slag — $CaO:SiO_2$ = appr. 0.7 — that was used here contributed to desulphurizing the iron to a considerable extent and that in this process high metal temperatures and obviously high cupola temperatures, are a prerequisite because both the metal and the gas have already reached a state of equilibrium 80 in. above the tuyere level and hence slag reactions cannot be expected as yet.

The experiments of the Research Group for the Hot-Blast Cupola made it possible to establish the quantitative relations which are shown in Fig. 7. The proportion of sulphur in the slag to sulphur in the metal rises with increasing metal temperature and with decreasing FeO contents of the slag. The results agree with the conclusions arrived at by R. Rocca, N. J. Grant and J. Chipman.[47] Further details as well as the theoretical basis for these results are given in the paper by H. Jungbluth and K. Stockkamp[36] which has already been mentioned and which recapitulates the latest stage of European cupola research.

After this short survey of some of the fundamental experiments in the field of hot-blast cupola melting technique, which cannot possibly be elaborated further in the scope of this paper, it should now be possible to set forth certain requirements of a hot-blast cupola plant which have to be met in order to reproduce and utilize the metallurgical possibilities. An important role is played, of course, by the blast temperature, and this was where development work on the author's hot-blast system plants set in.

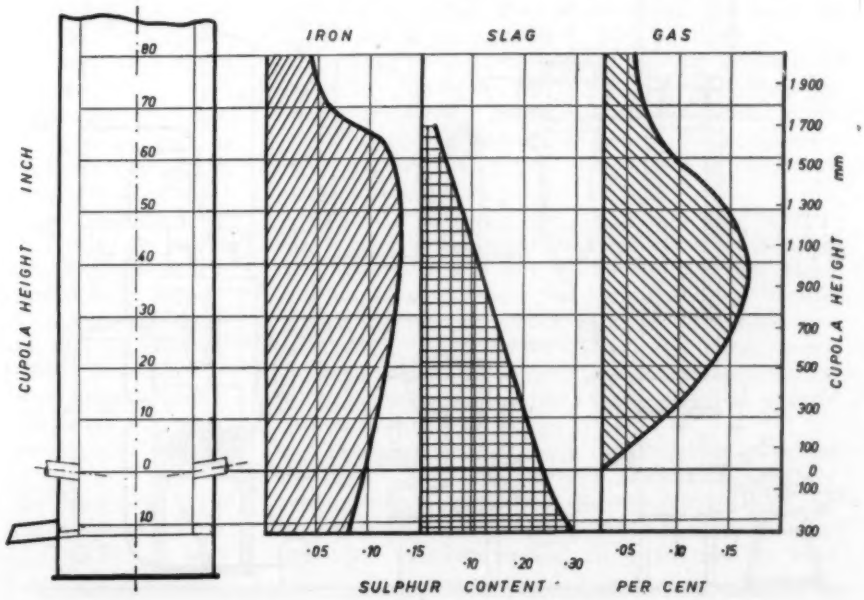


Fig. 6 — Behavior of sulphur in the acid-lined hot-blast cupola of System Esslingen-V.E.W. (according to E. Piwowarsky and F. Hage).

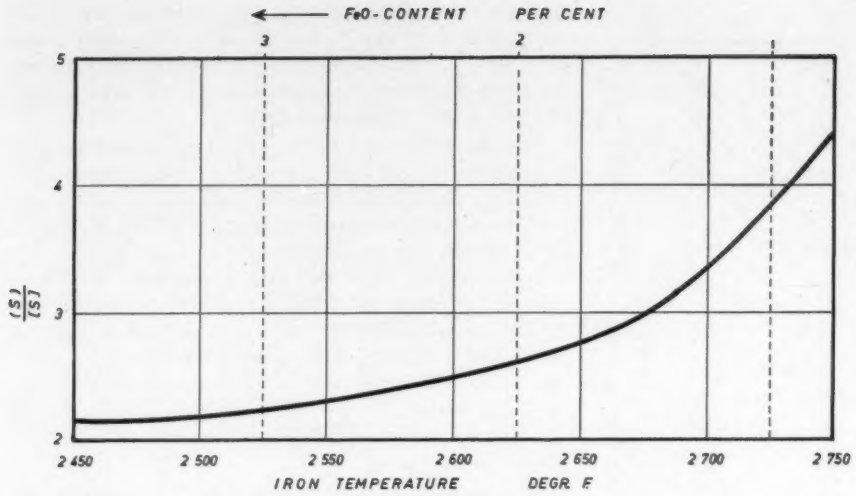


Fig. 7 — Effect on sulphur content by metal temperature and FeO content of the slag in the acid-lined hot-blast cupola (according to International Research Group for the Hot-Blast Cupola, and H. Jungbluth and K. Stockkamp).

The object was to produce a *recuperator unit* that would be capable, even with the use of heavily dust-laden cupola gases, to produce a blast temperature which is not only constant over any desired length of melting time, but is also easily controllable. Furthermore, the blast temperature should be comparatively high, above 750 F (400 C) in any case.

Metallurgically the need exists for a cupola that can be operated for any desired length of time and in which metallurgical processes will not be disturbed by uncontrollable factors as, for instance, lining consumption.

Figure 8 shows schematically the design of such a cupola plant which endeavors to meet the requirements mentioned to a considerable extent. Depending on the amount of gas available and the melting rate of the installation, the stack gases are taken off the cupola at a temperature of approximately 750 to 1100 F through one or more take-offs distributed around the circumference of the cupola, just below the upper level of the charged material. The charged materials above the take-offs and up to the level of the charging door serve to prevent the intrusion of infiltrated air into the system.

Through a dust separator (4) and a stack gas damper (5) the stack gases flow into the stack gas

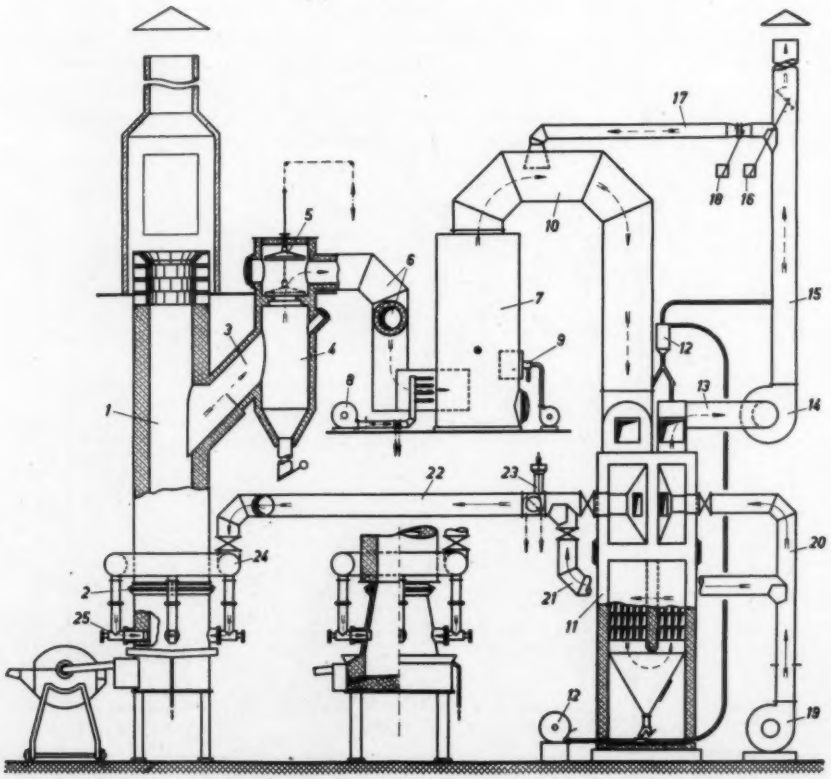


Fig. 8 — Diagram of a hot-blast cupola melting unit of author's system.

main and from there into the combustion chamber (7) where the sensible and latent heat of the stack gases are utilized by means of combustion. The hot gas main (10) then leads the burnt gases to the recuperator into which the gases enter, in the design shown here, from the top and which they again leave at the top to go through the exhaust fan (14) and the stack (15) out above the roof into atmosphere.

The cold blast air from the blower also enters the recuperator at the top, goes in at the same end where the waste gases leave the recuperator, passes through the recuperator in a cross-counter flow manner and leaves it also at the top at the gas inlet side, now as hot blast. This hot blast is then led through a carefully insulated main (22) into the wind belt (24) of the cupola and from there through the tuyeres (25) into the cupola. A by-pass main (21) at the recuperator makes it possible to operate the cupola with cold blast if for some reason or other the hot blast plant goes off. So far, the general arrangement corresponds to the generally known design principle of the hot-blast plant of any system; however, it differs from other designs greatly when it comes to design details of the various components.

The need to keep the hot-blast temperature constant over a long melting campaign, which from the metallurgical point of view is probably the most important requirement that has to be met by a hot-blast unit, is closely connected with the question of dust separation from the cupola stack gases. Therefore, a series of tests was initiated on a number of well-known dust separating devices, such as cyclones of various design. The success was surprisingly low!

It was possible, of course, to separate the major part of coarse and medium size dust, but it was just impossible to get out the very fine dust which is so detrimental to the efficient recuperator operation. These very fine flour-like dusts still deposited on the heat exchange elements and soon reduced the heat transfer and thereby the preheating of the blast. Irrespective of the detail recuperator design it was, therefore, frequently necessary to clean the recuperator with compressed air and at least once in a while more thoroughly with sand blast or brushes.

To explain all the many experiments made to solve this problem would be far beyond the scope here; it should be sufficient to merely touch upon some of the operating details that account for the present satisfactory results. Actually, dust separation commences in the stack gas take-off. This was intentionally made rather large in order to keep the velocity of stack gases comparatively low at this point. A large proportion of the coarser dust carried over is separated here, and these dust particles drop back into the cupola; the steep rise of the take off pass favors this dust separation. A further pronounced dust removal takes place in the dust separator (4). The dust separated here is collected in the conical bottom and can be drawn off whenever necessary.

The combustion chamber, as the successive large component part, has to fulfill a dual purpose, i.e. to burn the gases completely and also to serve as a further dust separator. To this end it is made cylindrical as a cyclone combustion chamber. The stack gases enter tangentially through the cylindrical shell after they have been thoroughly mixed at the inlet with a stream of combustion air supplied by the combustion air fan (8).

The result is a long spiral flame rotating around the center line of the combustion chamber. This increases the path length for combustion to take place and thereby assures complete combustion of the stack gases. At the same time there is a further dust separation based on a cyclone principle, by which practically all of the remaining medium size dust is removed. At the side of the gas inlet there is a small pilot burner (9) which serves for the ignition of stack gases in the combustion chamber. This pilot burner is usually operated on oil or gas.

The gases now leaving the combustion chamber and from which the dust has been separated, as has been described, actually still contain some amount of the very fine dust. All attempts to separate this latter dust mechanically or electrically have proved to be in vain. The only solution seemed to be to design the recuperator in such a way that a detrimental effect on its efficiency by the inevitable amount of very fine dust would be avoided. For this reason the recuperator was purposely equipped with *cast heat exchange elements*, which we shall call "tubes" for short. These tubes, which have an elliptical shape and walls between 0.16 and 0.28 in. with total lengths of 59 up to 78.7 in. (1.50 - 2.00 meters), are placed horizontally in the gas stream which passes by the outside of the tubes while the blast air goes through the inside.

The various tubes are grouped into banks dependent upon their particular function and consequent chemical composition, for instance highly heat-resisting alloys in the hottest part of the recuperator. In these banks the tubes are tight against one another but can still move against one another on ground surfaces. This prevents tension cracks to a considerable extent and makes possible the continuous operation of the recuperator. The various banks, and for blast temperatures up to 1020 F (550 C) there are normally four of such banks with varying numbers of tubes, are joined to the recuperator steel frame by means of specially designed expansion joints which prevent stresses being carried over to the recuperator structure, but still permit the natural expansion of tubes and banks of tubes.

With this particular design any tube can, furthermore, be taken out and replaced without difficulty. Unfortunately, it is not possible to comment about the frequency of this replacement operation because a great number of hot-blast plants of this system have now been in operation for several years without needing replacement. However, facilities for replacing the tubes are present and have been tested on experimental recuperators.

Although all these newer developments increased the life of the recuperator and made it mechanically safe in operation, they still did not entirely solve the dust problem, because it was still necessary to blow the tubes with compressed air from time to time and to clean them thoroughly at least once a month.

Particularly inconvenient were sinterings which attached to the hottest tubes at the gas inlet side of the recuperator and were difficult to remove. These developed where the burnt gases from the combustion chamber entered the recuperator at too high a temperature.

Such high temperatures did not present any danger to the heat exchange elements themselves, because these could readily stand temperatures up to 2100 F (1150 C) due to their chemical composition, but they were often above the sintering temperature of the dust getting into the recuperator. It was first attempted to cool the burnt gases by means of introducing cold air, but these attempts were not successful. Apart from the fact that cooling by cold air was always very abrupt, there were occasions, when for instance the cupola was operated on high coke ratios and correspondingly high carbon-monoxide percentages in the stack gases, that a further combustion took place, causing additional temperature increases at this most endangered spot of the recuperator.

The problem was solved only through the use of a *waste gas recirculation* (17) with which all plants of this system are now equipped. By operating butterfly valves between the waste gas recirculation main and the exhaust stack an accurately controllable volume of waste gases from the far side of the recuperator can be recirculated to the hot gas main (10) ahead of the recuperator and thus be mixed thoroughly with the burnt gases from the combustion chamber. These waste gases still have a temperature of approximately 480 to 570 F (250 to 300 C) so that they carry a considerable amount of heat, which makes the cooling effect very mild.

Simultaneously the waste gas recirculation presents this important advantage, namely, that the corresponding quantity of heat which is recirculated need not be taken off the cupola in the form of stack gas, which means that less stack gas is taken from the cupola. The stack gas, of course, is responsible for bringing in the dust, while the recirculated gases have already lost their last dust contents, so that the use of this kind of cooling reduces the amount of dust in the recuperator considerably. A control of the gas inlet temperature to the recuperator by means of this waste gas recirculation to a maximum level which is below the sintering temperature of the dust furthermore avoids any possibility of the dust sintering on to the tubes and possibly blocking the recuperator.

The latest step to settle the dust question for the recuperator and, therefore, to facilitate continuous operation of the hot-blast unit was the introduction of mechanical shot cleaning of the recuperator tubes. With this shot cleaning according to a Swedish process which was known from boiler practice and which was suitably developed further for use in this hot blast system, a spray of small steel shot is dropped on to the recuperator tubes from a container on top of the recuperator (12) over suitably shaped dispersers.

The dust is carried along by the shot and is col-

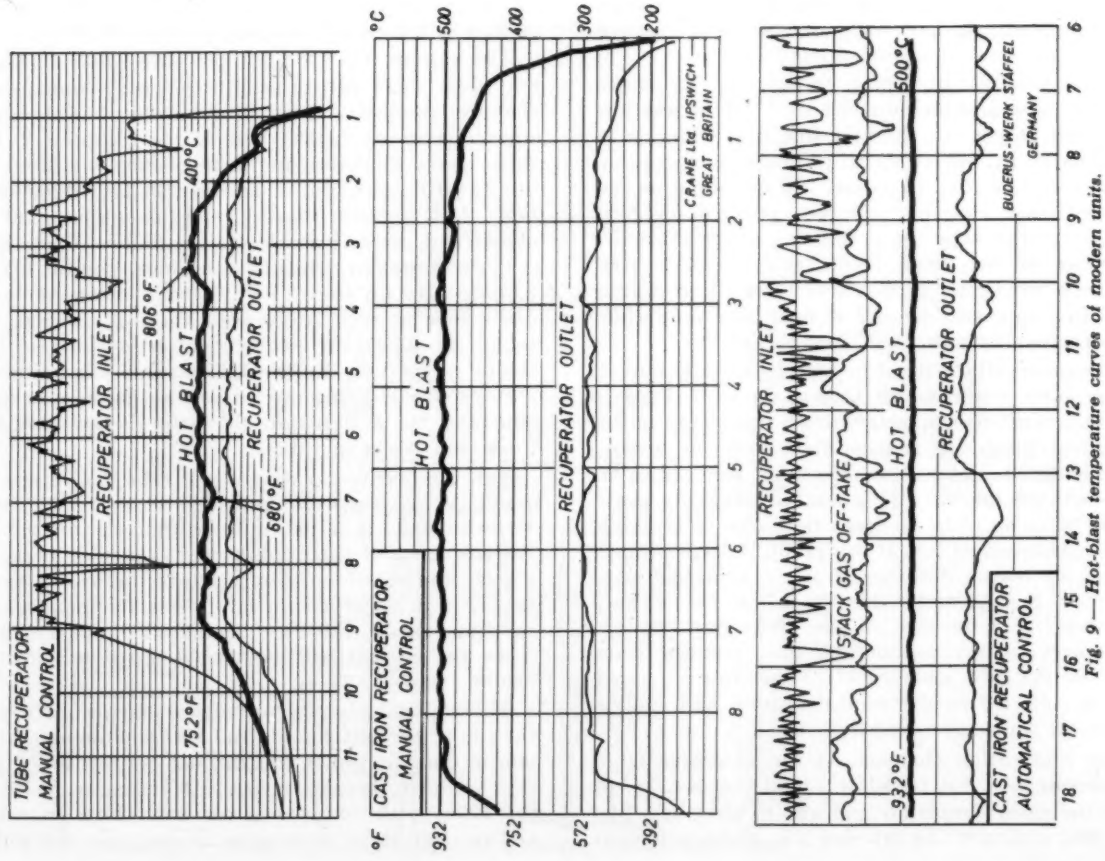


Fig. 9 — Hot-blast temperature curves of modern units.

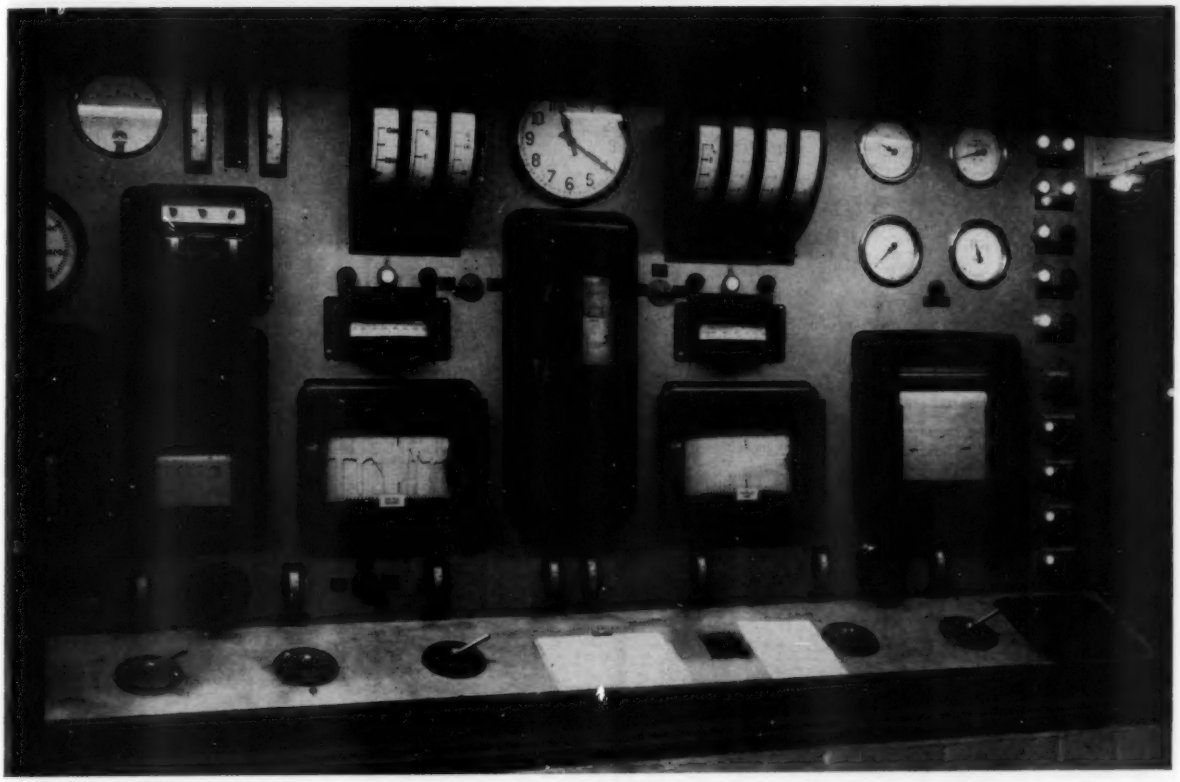


Fig. 10 — Control panel of a fully automatic-controlled hot-blast cupola melting unit (works photograph Platts (Barton) Ltd., Great Britain).

lected together with the shot in a hopper in the lower part of the recuperator. The steel shot is then separated from the dust and carried back into the container by means of compressed air supplied by a small blower, while the dust can be removed easily whenever required. This method of cleaning the recuperator, which is done during *actual operation,* made it possible together with the other steps described to operate the hot-blast unit *continuously* with a *constant* and easily *controllable* blast temperature.

All these control and regulating devices can, of course, be operated manually. However, it appeared desirable to develop automatic controls for the operation of the hot-blast unit which, after all, is only an auxiliary part of the cupola. This development has also been satisfactorily concluded, so that it is sufficient today merely to set this system at a desired blast temperature and a safe gas inlet temperature to the recuperator, and all further measures and control operations described here are done automatically.

Figure 9 shows the hot-blast temperature curves of a hot-blast unit incorporating a steel tube recuperator and operated by hand, below this a manually-controlled plant with cast recuperator which was equipped only with waste gas recirculation but had no shot-cleaning equipment, and below this a section of the recordings of a fully automatic-controlled large plant. A comparison of the different curves shows to what extent the target that was set, namely to keep constant blast temperatures over long melting periods, has been reached. A view of a control panel of such a modern hot-blast cupola plant is given in Fig. 10.

So far we have discussed only the hot-blast unit. But the good old cupola itself had to alter its face to meet the higher requirements imposed on it. Freely movable windbelts (24) and new tuyeres (25) were developed, and new cupola linings were investigated and experimented with. Much has been written about this so that it will not be necessary here to refer in detail to the various designs of a conventional cupola with lining. It may be mentioned, however, that with the utilization of American experience a new lining material came up in Europe for the operation of hot-blast cupolas with basic slag, namely carbon.[8,9,42]

Extensive investigations were made during the last few years on the development of a cupola in which it would be possible to melt for long periods with either acid, neutral or basic operation. Experiments led to a cupola with cooling device where in the melting zone there is no lining but a thin, undercooled and, therefore, neutrally reacting slag-iron layer. Water-cooled cupolas, also for cold-blast operation, have existed for a long time. They were primarily intended to delay lining erosion in the melting zone and thereby permit longer melting periods. The reduced lining erosion also made it easier to operate such a cupola with neutral or basic slag.

Apart from the known materials of acid, basic or neutral composition, a lining material of carbon or a

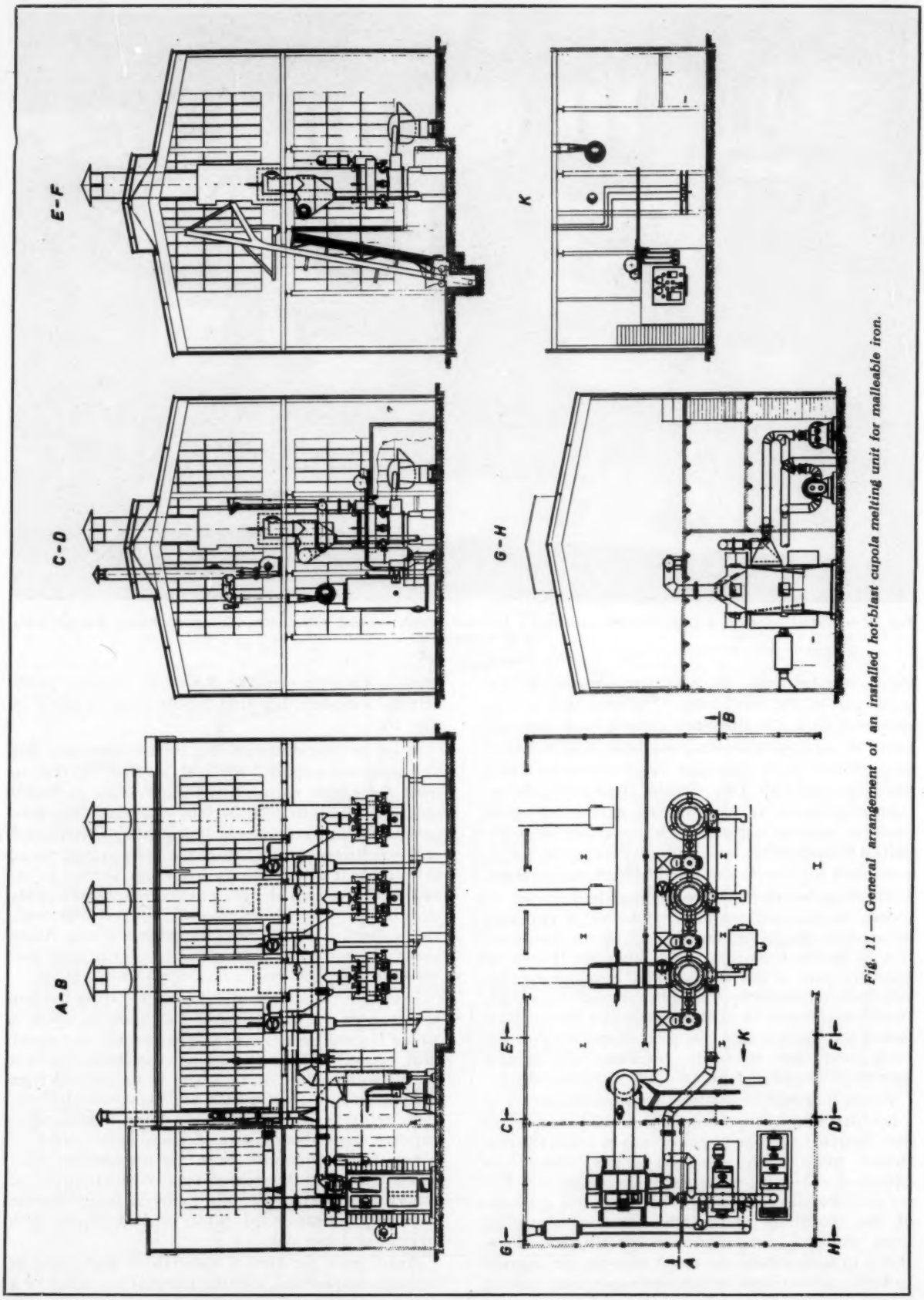


Fig. 11 — General arrangement of an installed hot-blast cupola melting unit for malleable iron.

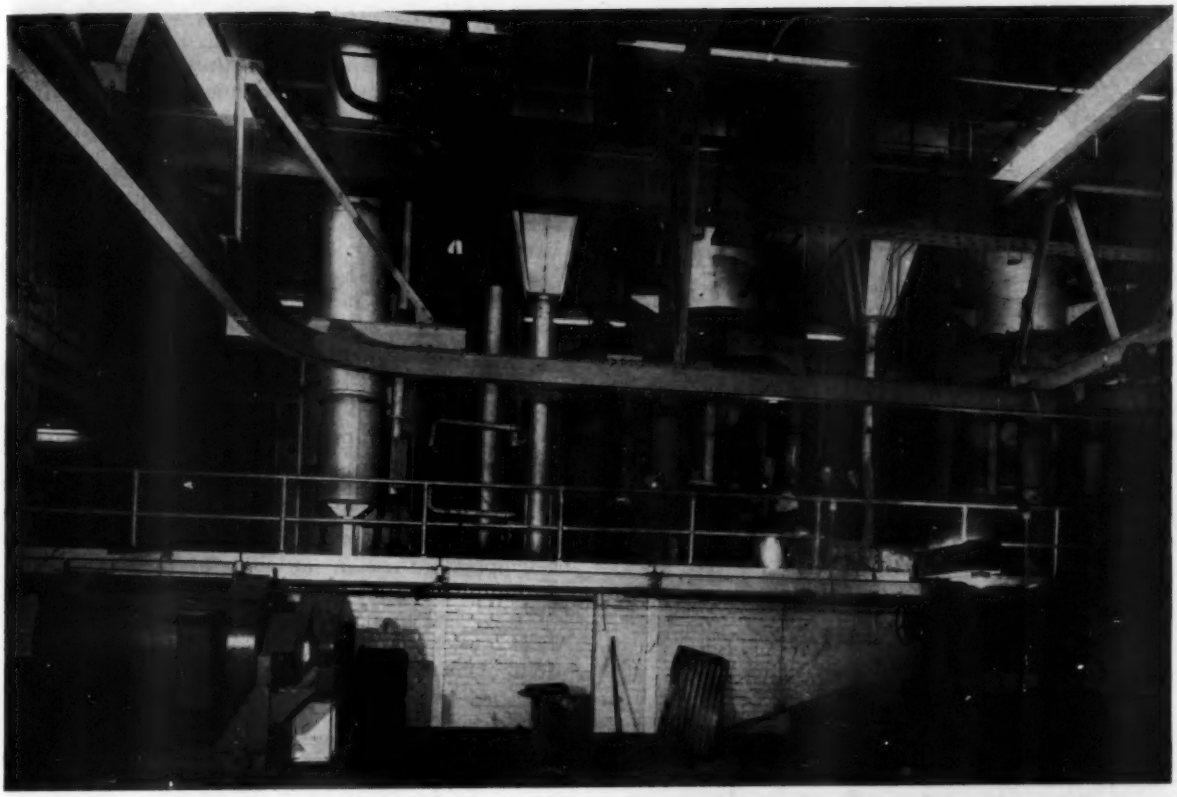


Fig. 12 — View of a hot-blast cupola melting unit operating without melting zone lining (works photograph Platts (Barton) Ltd., Great Britain).

mixture of ganister and carbon gave satisfactory results. Such cupolas started out with the melting zone covered by a layer of lining which in the course of the melt burnt down and left only a thin protective layer of slag on the cupola shell. The cupola was then taken out of operation and relined. The cooled cupola required a certain volume of water to cool the cupola shell and also, if required, the tuyeres, as well as a larger amount of fuel to compensate for the heat losses caused by the water cooling.

The data in the literature on the amount of the additional required fuel and on the required amount of water fluctuate exceedingly, which is no doubt attributable to the widely varying melting practices. Deviations of the same extent were found in the figures for the specific melting rate per hour, per unit of cross section of the cupola. Apart from a few older publications[43] there was very little experience on which to base the development of the new cupola without lining.

However, in more recent times there were some notable publications which confirmed the results mentioned earlier. Particular reference should be made to the reports by Th. Kootz and H. Rellermeyer,[44] J. R. Njedosjek,[45] P. Hubert,[46] F. Danis,[47] and A. De Sy, R. Doat, R. Balon and L. Winandy.[48]

Fundamental metallurgical advantages were shown in pilot experiments, using a modified design of a water jacket cupola; these were made some years ago

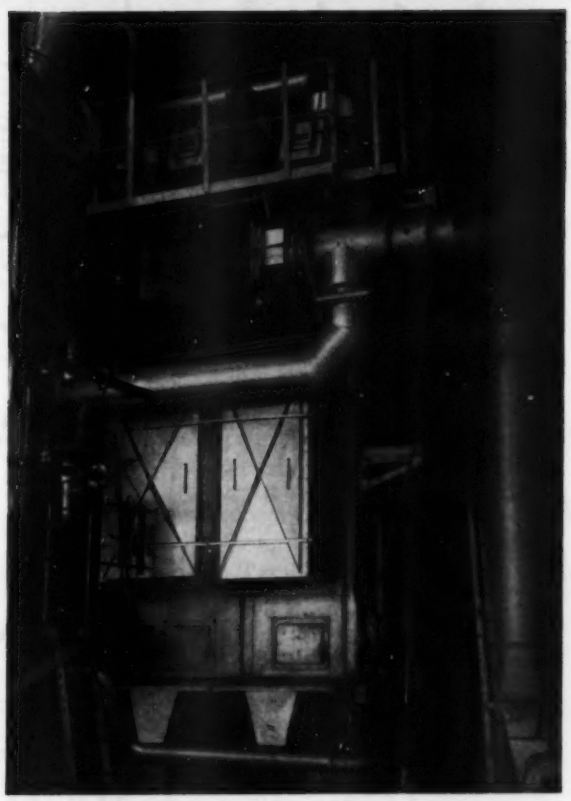


Fig. 13 — Cast iron recuperator of author's hot-blast cupola melting unit (works photograph Dobson and Barlow, Lt., Great Britain).

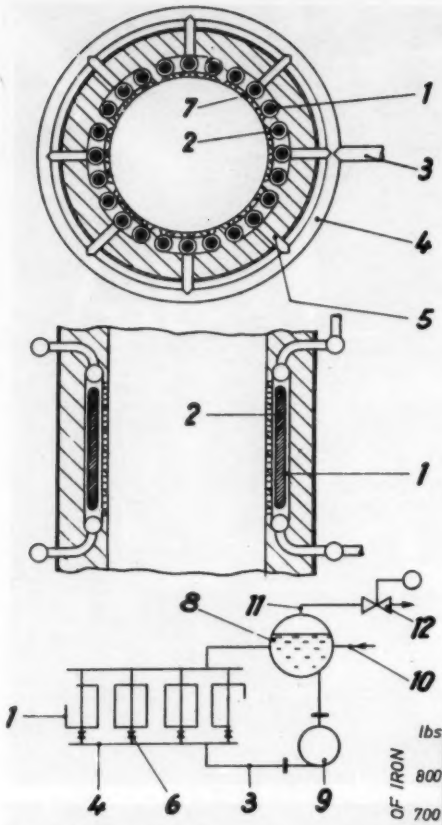


Fig. 14 (Above) — Diagram of pressure cooling for cupolas.

and had been reported.[49] The long course of development brought about two different designs of water-cooled hot-blast cupolas, without lining, which differed from one another not so much in their metallurgical behavior but in the method of cooling. The simplest design with conical, water-sprayed shell in the melting zone and projecting water-cooled tuyeres is shown in Fig. 8. Plants of this system are now in satisfactory operation in Great Britain at Dobson & Barlow, Ltd., and at Platts (Barton) Ltd. of Textile Machinery Makers Ltd. One such plant is shown in Figs. 11, 12 and 13. The metallurgical experience gathered here will be mentioned in somewhat more detail at a later point.

We shall next describe another form of water-cooling that was proven to be satisfactory in the experimental plant mentioned earlier. To cool the cupola with water requires a considerable amount of water, which according to R. Doat and DeBock[16] and also J. R. Njedosjek[45] can amount to as much as 2,200 gallons per ton of molten metal, when operating with a temperature rise in the water of 18 F. The water consumption figures obtained in the meantime with a cupola without lining were found to be around 1,900 gallons per ton of molten metal (8.5m³/t). With the shortage of water such as prevails in many

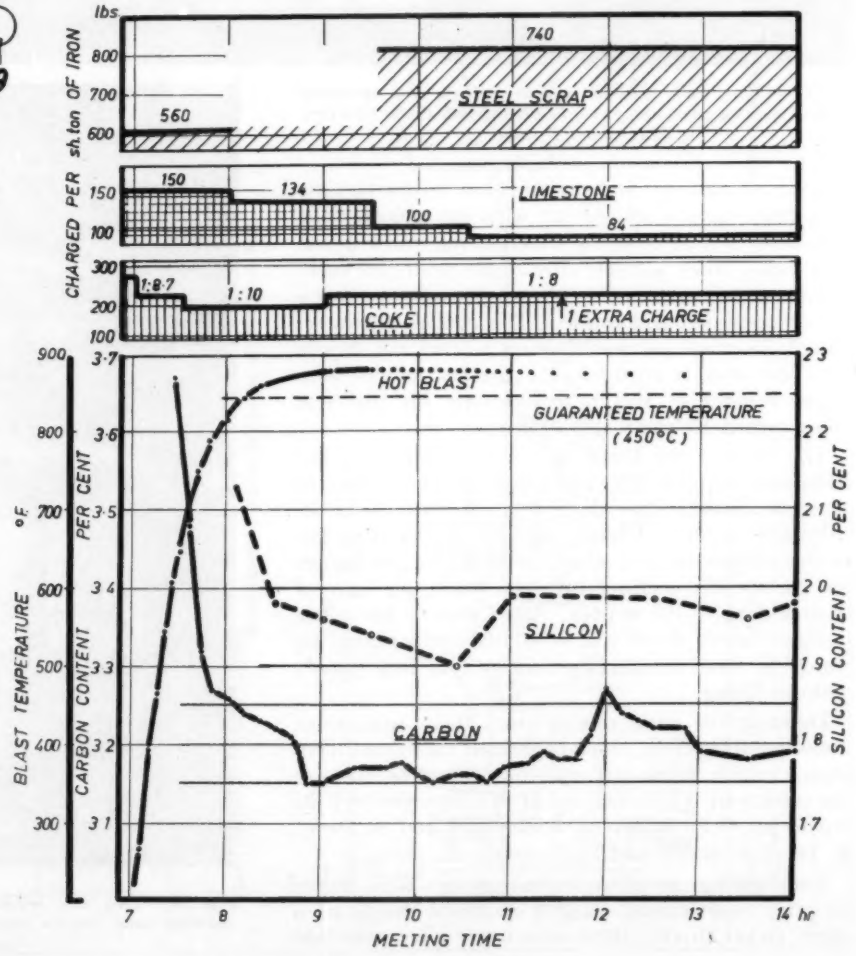


Fig. 15 — Metallurgical start-up of a 12-ton per hour hot-blast cupola melting unit for automotive castings.

parts of Germany, especially in the Ruhr, a water consumption as high as this adds up to considerable expense. Moreover, it must be borne in mind that an appreciable quantity of heat is permanently lost through this water, which at a water temperature rise of 86 to 122 F (30 to 50 C) can hardly be utilized again.

Therefore, a new cooling system was developed for these cases, in which cooling tubes are installed in the cupola parallel to the shell and the water in the cooling tubes is heated under generation of pressure and is removed as saturated steam. With the utilization of the heat of vaporization the amount of water that has to be continuously added to this cooling system comes to only one-tenth to one-twentieth of that required in the normal water-cooling designs. Another advantage of this system is that the heat lost in the cupola through cooling can be reapplied economically in the form of steam. Figure 14 shows the diagram of such a pressure-cooling system. Tuyeres, slag separators, or other components that are to be cooled can also be connected to this system.

This short survey of the development of the modern hot-blast cupola unit, described by the example of a system with which the writer is especially well acquainted and which he picked out just for this reason, it is hoped illustrates clearly the trend of the present, and presumably also future, development of the hot-blast cupola in Europe.

Operating Experience

Now what can we do with such a hot-blast cupola melting plant as a production unit and to what extent does it fulfill what we expect of it? Figure 15 shows the course of events on the first day of operation of a hot-blast cupola plant for the production of automotive castings with a melting rate of approximately 12 tons per hour. When placing the order the customer required:

1. a metal temperature at the pouring point of 2,558 F (1,420 C),
2. to use as large a quantity as possible of cheap charging materials, particularly steel scrap,
3. a maximum limit of 0.10 per cent in deviations of the analyses for carbon and silicon,
4. pearlitic structure in the cast product, and no detrimental effect on machinability through surface chill or excessive hardness at the corners.

In anticipation of the expected metallurgical behavior of the cupola, preliminary calculations were made, before putting the plant into operation. The first coke charge, the composition of the metal charge and the amount of fluxes to start with were determined from these calculations. With increasing blast temperatures the latter were then to be adapted to the metallurgical alterations in the melting process.

The result was the course of events logged in Fig. 15. The entire system was still somewhat moist, of course, and hence temperature was increased very slowly but steadily to the guaranteed temperature of 842 F (450 C) and was then kept at approximately 880 F (470 C). The carbon content had reached the upper limit of the required analysis after about 70 minutes and was kept within the set limits without difficulty. With the silicon it was somewhat more difficult to get into the requested range of tolerance in analysis, but this was also accomplished after about 1½ hr; silicon contents could then also be kept within the limits of 0.10 per cent. This meant that with regard to the analysis the customer's request was met within a few hours after the start of operations.

It is interesting to see how melting behavior of the cupola followed the principles which we have described earlier when we spoke about hot-blast cupola metallurgy in general. It is apparent that at about 10:30 o'clock the silicon content was at the lower permissible limit. In view of the structure desired it appeared necessary, however, to raise the silicon content to somewhere near the upper limit. An addition of ferro-silicon was to be avoided, so as not to lose the economic advantages over the cold-blast operation already achieved.

The percentage of steel scrap in the charge, which had been 8 per cent with cold blast, was therefore increased to 37 per cent, at the same time the addition of limestone was reduced and the coke charge slightly raised again. Metallurgically these measures altered the concentration relations in the cupola. The increase in the percentage of coke ash and the simultaneous reduction of the CaO proportion in the form of limestone increased the concentration of SiO_2 in the slag, i.e. it became more acid.

On the other hand the increase in the percentage of steel scrap in the charge meant a reduction of the silicon concentration in the metal. Therefore, conditions for a reduction of silicon from the slag were bound to become more favorable, more so as with the same blast temperature the increase in the coke charge led to an increase in the metal temperature (cf. Fig. 4) and to a decrease in the combustion ratio η_v. The steps taken were confirmed to be correct by the rise in the silicon curve. The carbon contents hardly changed, because the carbon pick-up was obviously checked by the reduction in basicity of the slag. The reason for the fact that at about 12:00 o'clock the carbon exceeded the upper limit in the analysis was that the cupola crew had put in an extra charge of coke from sheer habit; this had not been asked for and, therefore, could not be balanced.

At any rate, the example shows that it has, in fact, become possible to control and steer the melting procedure in a hot-blast cupola, even when—as in this particular case—the cupola in question is fully lined, which means that the lining erosion, which cannot possibly be calculated in advance, brings in a factor of uncertainty. Figure 16 shows that the metal produced met the requirements also with regard to the microstructure; there were no differences found in machinability.

Figure 17 shows the metallurgical course of events during the start of operations of a 12-ton per hour hot-blast cupola plant at Crane Ltd. in Great Britain. This unit was to produce *black-heart malleable iron* from the cheapest possible charging materials. The alteration in the additions of ferro-alloys which were made during this first melt are apparent from the

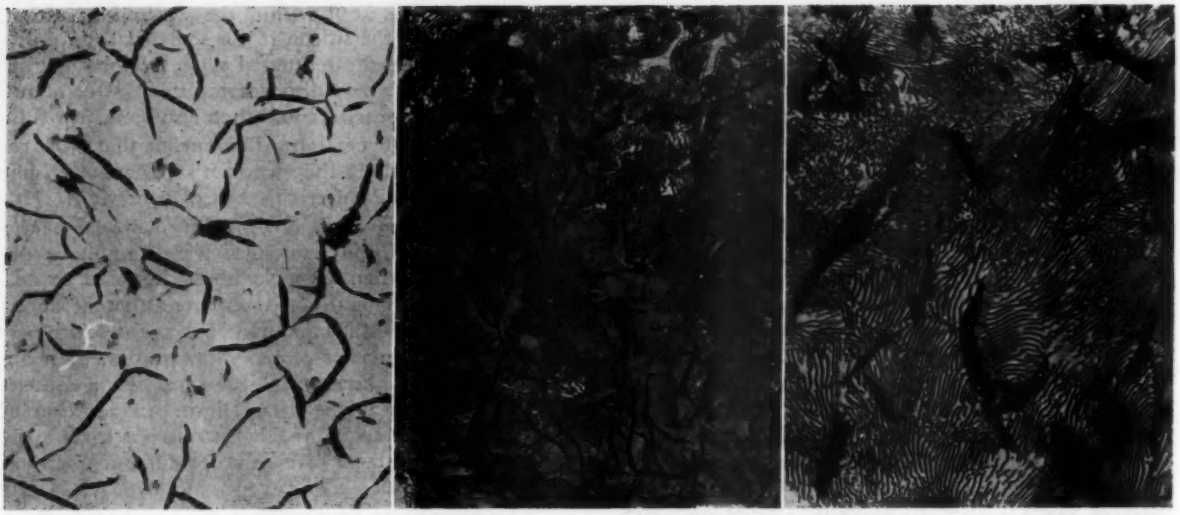


Fig. 16 — Microstructure of hot-blast cupola iron for automotive castings.

upper part of the illustration. Of particular interest in this connection are the smaller ferro-manganese additions, made possible by the lower manganese melting losses in the hot-blast cupola, despite acid slag, and the complete elimination of ferro-silicon additions after six hours of melting.

The middle section of Fig. 17 shows that there was a reduction in the percentage of pig iron in the charge, from 14 to 5 per cent, and at the same time an increase in the steel scrap charge to 32 per cent. The savings in costs, compared with cold-blast operation, which were brought about by these measures are illustrated in the lower part of Fig. 17. At the end of first day's operation a saving of $4.22 per ton of liquid

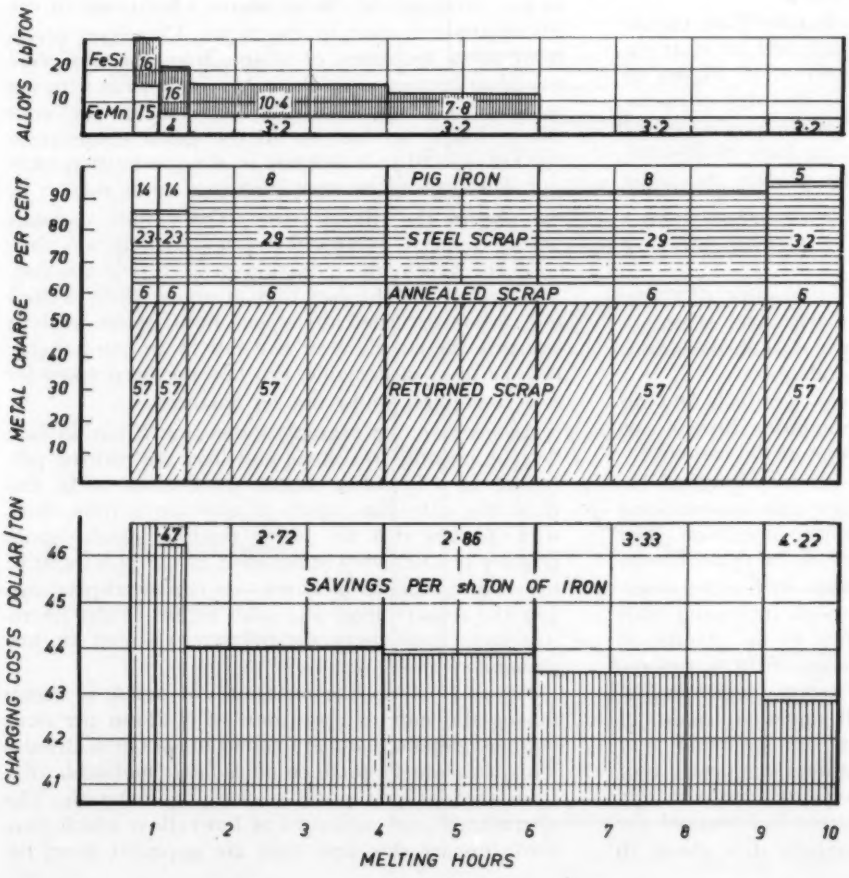


Fig. 17 — Start-up of a hot-blast cupola melting unit to operate on an economic charge for production of black-heart malleable iron.

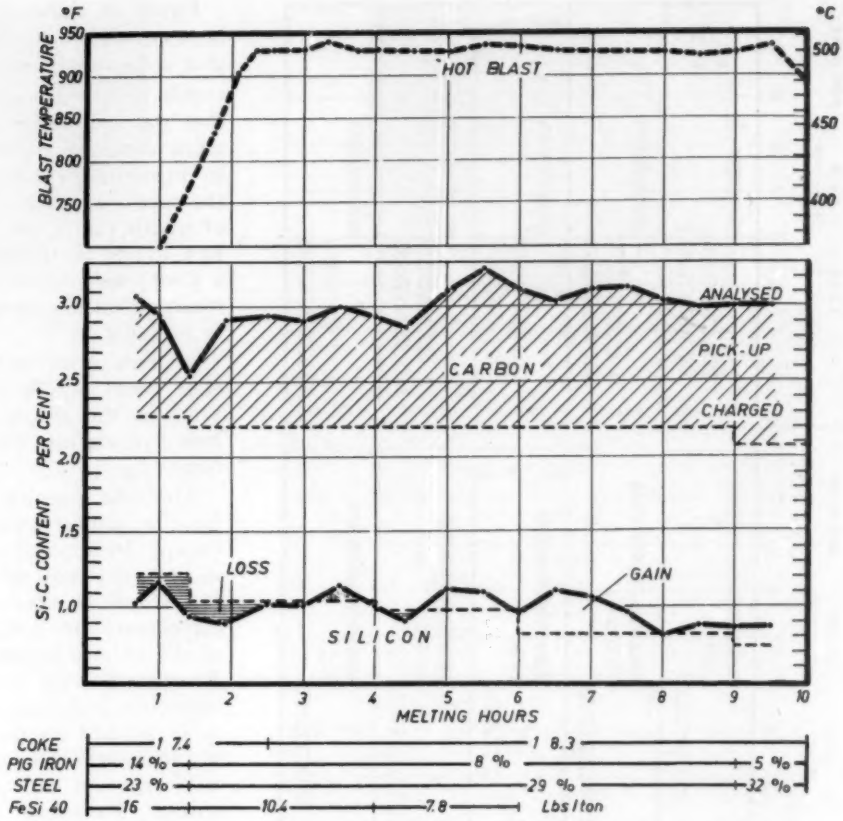


Fig. 18 — Actual silicon pick-up in a production of black-heart malleable iron.

metal had been achieved merely through the alteration in the metal charge. Further savings have been obtained in the meantime so that due to additional gains the installation will pay for itself within an unexpectedly short period.

Metallurgically, these economic advantages could be obtained only if with the utilization of all possibilities the cupola was expressly operated in the direction of *silicon pick-up from the slag*. Figure 18 shows the success of these endeavors. When all the steps had been taken, the carbon and silicon contents of the metal were only slightly different from the desired analyses. To do all this it was, of course, absolutely necessary to keep the hot-blast temperature at an even level so that undesired changes in the thermal, and consequently metallurgical, conditions within the cupola would be ruled out and that it would be necessary only to alter the concentration factors.

Figure 19, which concerns the same plant, shows that this result was not brought about by coincidence, but can be reproduced at will. For reasons of comparison, there are shown in the upper part of Fig. 19 the metallurgical data of the last week that the cupola was run on cold blast, and below this are data for melting days when the same cupola was operated on hot blast. Particularly apparent are the carbon pick-up, which is twice as high in the hot-blast cupola, and especially the silicon pick-up, in contrast to a considerable silicon loss in the cold-blast cupola. With regard to manganese, the reduced melting losses in hot-blast cupola operation are not clearly visible in Fig. 19 because melting losses and gains are expressed here in per cent of the analysis of the charge.

However, with cold-blast operation the melting loss of approximately 30 per cent was related to a charge of 15 lb ferro-manganese per ton, and that the same per cent melting loss with hot blast is related to only 3.2 lb ferro-manganese per ton, and that at the same time the total of manganese brought into the cupola through the steel scrap and pig iron is reduced, then this illustration also represents a saving in manganese. Incidentally, this form of illustration of the melting procedure in the hot-blast cupola clearly shows the correlation between silicon and manganese; higher silicon pick-up is connected with higher manganese melting losses, and vice versa.

The black-heart malleable iron produced in this hot-blast cupola did not show any differences in its micro-structure when compared to the previous cold-blast cupola iron, as is shown in Fig. 20. Neither were there any differences found during heat treatment and machining.

Operations With Poor Coke

The fact that the metallurgical proceedings in the hot-blast cupola can be guided and controlled also facilitates the use of lower quality fuels without detrimental effect on metal quality. In a special test series it was, therefore, investigated under what conditions an iron for use in automotive castings could be produced in an acid-lined hot-blast cupola with the use

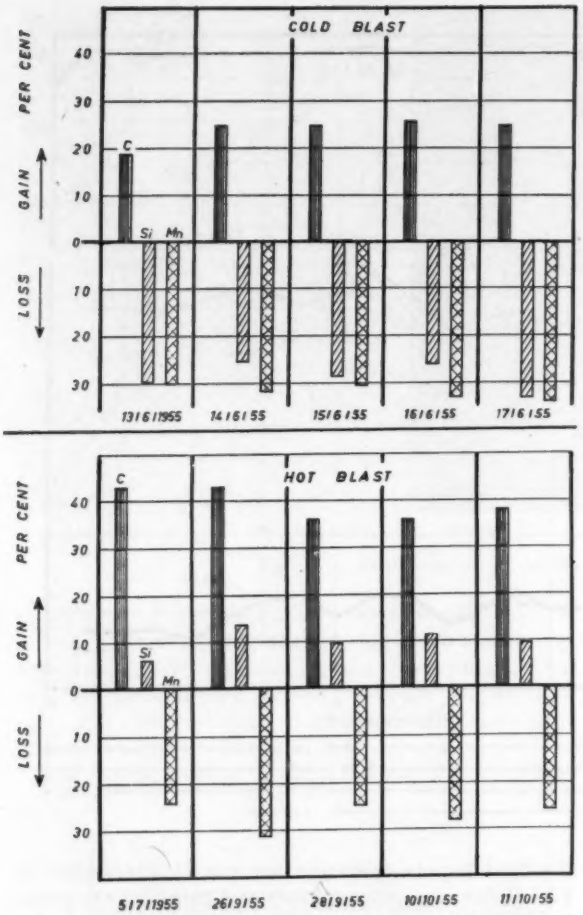


Fig. 19 — Melting behavior of one cupola with cold-blast operation and with controlled operation on hot-blast.

of only soft, high-ash coke. The coke taken for these experiments had C_{fix} — 77.3%
sulphur — 1.2%
ash — 19.4%
calorific value H_u 6.398 kg cal/kg coke.

The screen test showed 83 per cent over 1.57 to 2.36 in. (40 to 60 mm), shatter test 62.8 per cent.

In one series of test runs, the metal charge consisted of 40 per cent pig iron, 40 per cent bought cast iron scrap and 20 per cent return scrap, in another series it was 40 per cent pig iron, 30 per cent steel scrap, 10 per cent bought cast scrap and 20 per cent shop returns.

The cupola had a 24-in. inside diameter and was operated with hot blast of 840 F (450 C); tapping was done continuously through a siphon.

As a preliminary result of these experiments, which are still to be continued, it can be stated that it is possible to produce good quality automotive castings under these conditions. With the same blast volume and hot-blast temperature that would normally be used with high-grade foundry coke, the melting rate is somewhat lower, in fact, actually somewhat lower than would be expected from the low carbon content of the coke.

Figure 21 shows a few average values that have been established in comparison with the melting rates that are normally encountered in the same hot-blast cupola using normal foundry coke. The blast pressure that was required was remarkably high. There was a fan blower to produce the blast, and it was difficult to introduce the volume of blast quickly enough into the cupola to sustain fast combustion. For such types of readily caking coke it seems to be advisable, therefore, to operate the cupola with a positive blower like a Root's blower, which contains sufficient margin to overcome such cupola troubles as sticking, hanging, or bridging.

Analysis of the metal at the spout was held within close limits, after comparatively large initial fluctuations in the silicon contents had been overcome. Iron temperatures fell in line with the applied conditions.

The iron foundry *hot-blast cupola operated with basic or neutral lining* has not yet been accepted in Europe. Its application will probably remain limited to certain brands of high-quality cast iron, because the inherently more expensive lining costs have to be compensated by a correspondingly high selling price of the finished product. Experiments made by E. S. Renshaw[10] have become generally known in the meantime and have frequently been discussed. The same applied to the work by E. Piwowarsky and H. Schmidt[50] who made comparative investigations with acid and basic slag at the Aachen Foundry Institute. Following that series of experiments H. Schmidt[51] then compared the costs of basic melting with those of acid melting, both on the basis of German raw material prices, and he arrived at the conclusion just mentioned.

A recent publication by W. Heinrichs[42] on the practical experience of operating a hot-blast cupola lined with carbon gives some theoretical considerations, but in addition this unfortunately limits itself primarily to discussing the composition of the slag used. The cupola was run at Krupp, Essen, with basic slag. More informative is an investigational report which was given at the 21st International Foundry Congress 1954 at Florence by the Italian experts C. Bortolani, W. Demicheli and C. Longaretti.[20] This report is concerned with the production of cast iron for ingot molds from a basic-lined hot-blast cupola. With a basicity of 1.8 average and varying percentages of steel scrap in the charge they obtained the carbon contents shown in Table 2.

TABLE 2 — INFLUENCE OF PERCENTAGE OF STEEL SCRAP IN THE CHARGE ON CARBON PICK-UP IN A HOT-BLAST CUPOLA

Steel scrap, %	Carbon contents, %
Basic lined cupola	
40	4.13
65	3.79
100	3.52
Acid lined, same cupola	
30	3.68
50	3.63

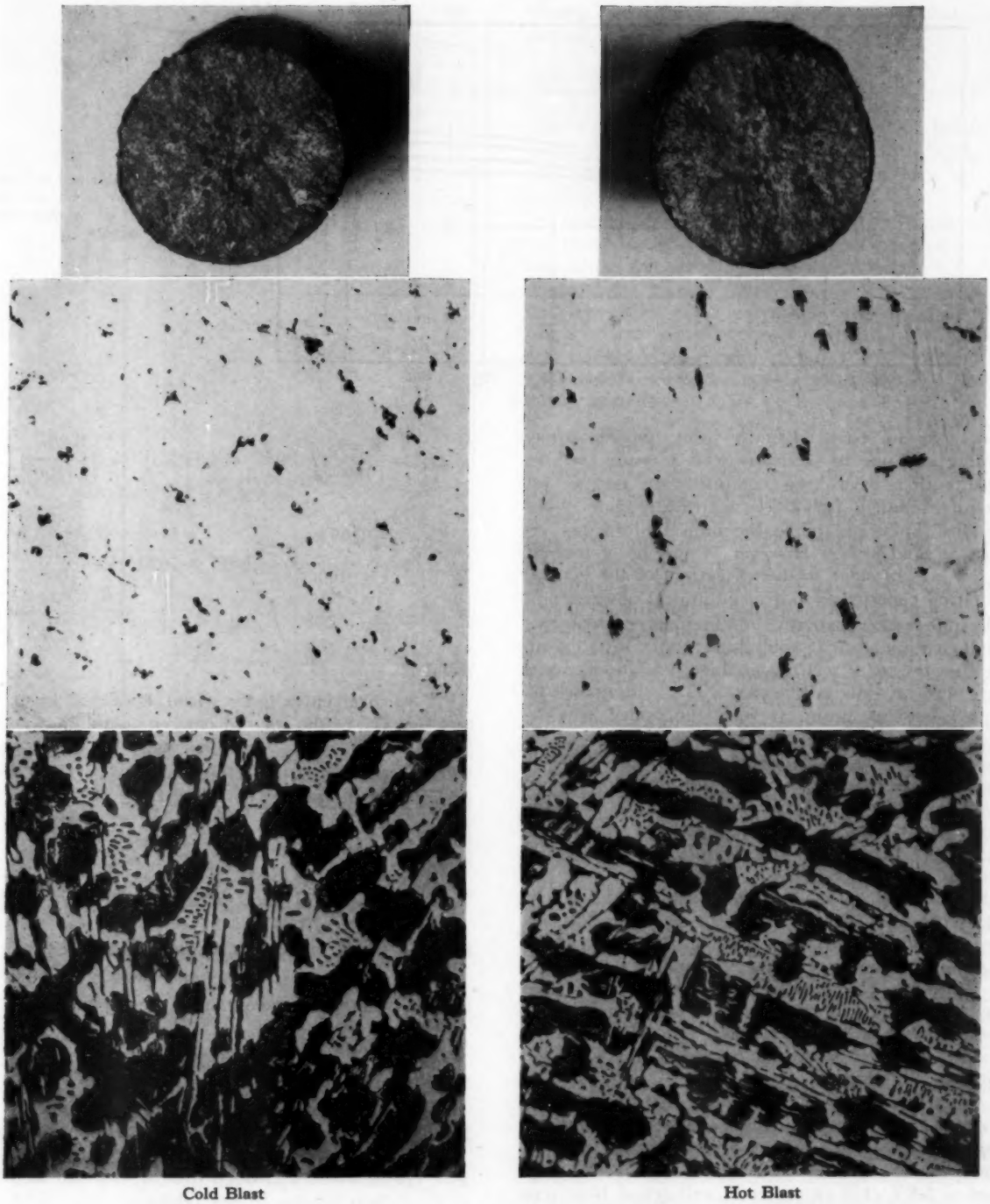


Fig. 20 — Fractures and photomicrographs of black-heart malleable iron produced in hot-blast cupola appear identical to that of iron produced previously in cold-blast cupola. Photomicrographs at 200X; Center, unetched; Below, etched.

The metal from the basic cupola was well suited to the production of ingot molds and had at least the same quality as the metal from the acid-lined cupola.

In continuation of the Piwowarsky investigations of the basic hot-blast cupola, M. G. Frohberg and W. Döbber[52] found that coating the coke, which for a long time held promise as a means to lowering the carbon dioxide reduction, did not effect an increase in the tapping temperature of the metal.

Hot-Blast Cupola With Cooled Melting Zone Without Lining

Particular attention has been paid during the last year or so to the *hot-blast cupola with cooled melting zone without lining.* As we have mentioned earlier, there are at present three fully automatic-controlled hot-blast cupola installations of this system in opera-

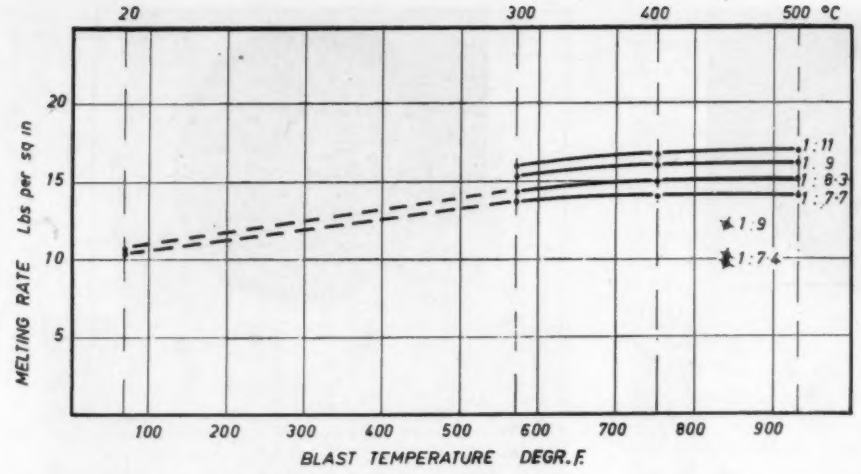


Fig. 21 — Influence of the quality of the coke on the melting rate of a hot-blast cupola.

tion, two in Great Britain using outside spray cooling and one in Germany with pressure cooling. Results that have been compiled are not as yet extensive enough to allow scientifically sound conclusions. These are to be the subject of further reports at a later date. However, it may be of interest at the present stage to discuss certain of the general operating experiences and metallurgical observations.

Despite rather extensive preliminary experiments conducted in a small cold-blast cupola installation, the first results of melting in these new cupola units were not at all satisfactory. The automatic-controlled hot-blast units as such functioned trouble-free, however, despite the coke charges being increased (1:67 to 1:5.3) the metal temperatures obtained were exceedingly low and did not exceed 2,640 F (1,450 C). Particularly annoying were the large fluctuations in metal temperature which with only short periods of shutting off the blast could amount up to 90 F (50 C).

From what has previously been said it need not be surprising that consequently the metallurgical behavior of the cupola kept varying all the time and that uniform analyses and metal qualities could not be achieved. The only positive result at this stage was that the cupola could, as had been expected, be operated with acid, neutral, and also basic slag and that to switch over during operation did not meet with any difficulty.

A first explanation of this difficulty was obtained from an approximate calculation of the *thermal balance* of the cooled cupola without melting zone lining. This calculation was based on a coke consumption of 1:6.7 (150 kg/t metal), and actual blast temperature of 932 F (500 C), a stack gas temperature at the cupola takeoff of also 932 F (500 C) and a combustion ratio of η_v = 40 per cent, which was calculated from the stack gas analysis. The consumption of cooling water was measured and the corresponding heat was arrived at as 12 per cent of the heat brought in by coke and hot blast. This figure corresponds approximately to the values given by Th. Kootz and H. Rellermeyer.[44] The effective heat actually delivered to the iron and slag has then to be calculated as follows:

$$Q_N = \frac{K \times k}{100} \Big[8080 + 4.44 \times i_w \left(1 + \frac{\eta_v}{100}\right) - 5.377 \times i_A \left(1 + \frac{0.65 \times \eta_v}{100}\right) - 1.867 \times 3020 \left(1 - \frac{\eta_v}{100}\right) - \frac{a}{100} \times 8080 \Big] \text{ kg cal/t iron} \qquad (1)$$

where:

K = coke charge in kg/t iron K = 150 kg/t

k = C_{fix} of the coke in per cent
k = 88% for Ruhr coke

8080 = calorific value of carbon in kg cal/kg C

4.44 = blast volume to burn 1 kg C to CO, in $Nm^3/kg\ C$[53]

$\eta_v = \frac{CO_2}{CO_2 + CO} \times 100\%$ η_v = 40%

5.377 = volume of stack gas at η_v = 0%, in $Nm^3/kg\ C$[53]

i_w = heat content of the blast, in kg cal/Nm^3
i_w = 160.2 kg cal/Nm^3

i_A = average heat content of stack gases, in kg cal/Nm^3
i_A = 168.6 kg cal/Nm^3

1.867 = carbon monoxide content of stack gases at η_v = 0%[53]

3020 = calorific value of carbon monoxide, in kg cal/Nm^3

a = heat losses through cooling, fraction of total heat intake, in per cent
a = 12%.

According to this formula, the volume of effective heat is

$$Q_N = 457{,}000 \text{ kg cal/t iron.}$$

This quantity of heat would be available to heat up iron and slag, to disintegrate the limestone and to compensate for radiation losses, etc. if the cooling losses calculated from water temperatures and water consumption per ton of iron are in fact only 12 per

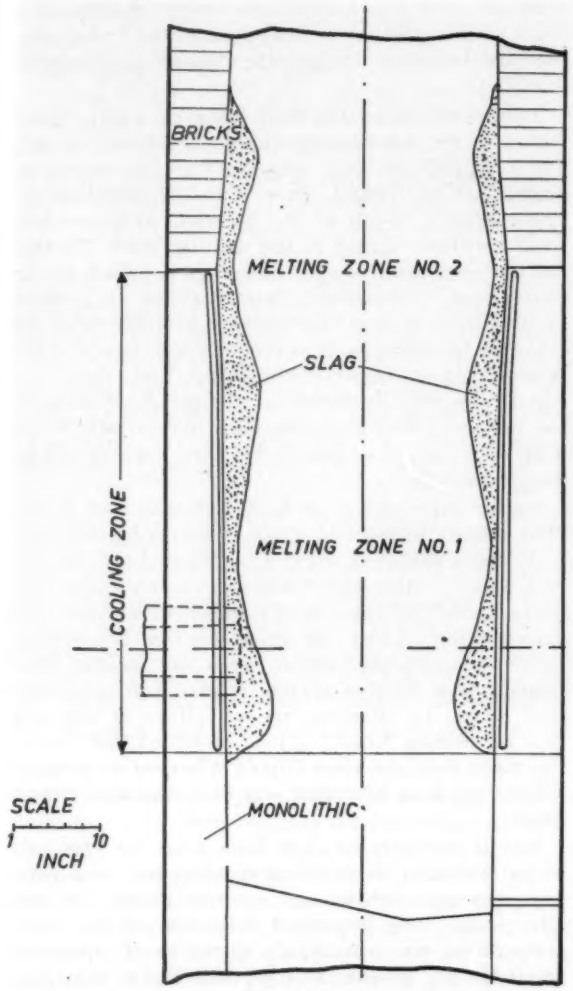


Fig. 22 — Cupola wall profile of a "cold going" hot-blast cupola without melting zone lining.

cent of the heat brought in through combustion of coke and through hot blast, i.e.

$$Q_K = 143{,}760 \text{ kg cal/t iron.}$$

Thermally the losses are then to be reckoned as

$$\frac{143{,}760}{8080 \times 0.88} = 2.2\% \text{ coke}$$
$$= 22 \text{ kg coke/t iron} \qquad (2)$$

so that for the melting operation there would still be 12.8% coke available.

However, there is a fallacy in this calculation. In a water-cooled hot-blast cupola without lining in the melting zone the quantity of heat released from 12.8 per cent of coke is weighted with heat losses from the stack gases in the form of latent and sensible heat, which would correspond to a coke charge of 15 per cent. It is necessary, therefore, to deduct from the amount of effective heat which we had just arrived at these additional heat losses which correspond to a coke charge of 2.2 per cent and to express these by the required amount of coke to compensate for them. Such a calculation gives a total of cooling losses, expressed in coke, of:

$$2.2 + \frac{67{,}960}{8080 \times 0.88} = 3.15\% \text{ coke} \qquad (3)$$

The water-cooled hot-blast cupola without lining in the melting zone that was operated with a 15 per cent charge coke performed metallurgically only as an 11.8 per cent charge coke!

This realization seemed to present an explanation for many unexpected results during the early melts, but it still was no solution to the problem. A coincidence then helped to clear up the problem, for at one time when melting synthetic pig iron from 100 per cent steel scrap, the cupola threatened to freeze and was therefore quickly dropped. After the cupola had cooled down, the cupola walls had the profile shown in Fig. 22, namely a rare and clear illustration of a double melting zone, such as W. Levi[4] had found in his investigations on the cooled cupola.

There was reason to assume that with a normal charge of pig iron and cast scrap there would also be a simliar phenomenon, even if not to the same extent. Obviously the cupola without lining did not melt uniformly over the whole cupola section. There had been some other observations which only now became significant. For instance, the fast dropping of the metal temperature when the blast had been shut off for only a short period, high stack gas temperatures at the takeoffs, and a large volume of heat taken up by the cooling water on the cupola shell. All these indicated a possibility that the cupola was melting essentially on the shell and not in the center.

The cooled tuyeres protruded only 3 in. into the cupola, and the blast did not penetrate into the center part of the cupola, but produced an annular combustion and melting zone, extending upwards, which at the top was limited by the combustion temperature of the coke. When this combustion temperature was no longer reached due to the powerful shell cooling effect, then the oxygen that was still not burnt climbed further up in the stack until combustion was possible again, i.e. above the cooling zone. With plenty of blast available, the volume of such oxygen could be sufficient to produce a second melting zone.

The extent to which this theory is correct is, of course, another question. In any case, it was assumed to be correct at that time and further steps were taken accordingly. The most obvious step, namely to have the tuyeres project farther into the cupola, could not be taken. For this reason it was attempted to alter the velocity of the blast in the tuyeres. The details of these investigations are to be the subject of a separate paper by D. Fleming.

We may, however, now mention the successful results that have in the meantime been obtained. It was found that with a definite blast velocity, which however is different for each cupola diameter, all the previously described difficulties could be overcome. The metal temperature increased by more than 212 F (100 C) and could readily be kept at an even level over long melting periods. Here again it was an absolute necessity to have the possibility of easily con-

trolling the blast temperature and to keep it uniform automatically at a certain level.

Thermally the resulting alteration of the combustion and melting conditions were revealed by a shifting of the cooling losses from the shell to the tuyeres, as is apparent from Table 3.

TABLE 3 — SHIFTING OF COOLING LOSSES IN WATER-COOLED HOT-BLAST CUPOLA WITHOUT MELTING ZONE LINING

Cooling		Cold Cupola	Hot Cupola
Shell,	kg cal/h	415,000	327,000
Tuyeres,	kg cal/h	550,000	618,000
Total,	kg cal/h	965,000	945,000

With an almost constant amount of heat consumed in water cooling the heat removed in the hot blast cupola has increased at the tuyeres and diminished at the shell. The melting zone has obviously shifted to somewhere nearer the tuyeres and is no longer spreading on the shell surface. The actual melting rates and metal temperatures in the hot cupola can only be explained, however, when it is assumed that the cupola has become "center going," that is, that it melts over its whole cross section.

As previously explained, the cupola that is improved in this manner is no longer as sensitive to periodic shutting off of the blast as the one melting on the shell. It would still be wrong, of course, to operate it exactly as a normal hot-blast cupola with lining in the melting zone, even after the described steps have been taken. A certain sensitivity to interruption of operation will no doubt always have to be tolerated, even though the metal temperature recovers quickly. This is only natural when you keep in mind that with such a cupola an amount of heat of 8,000 to 10,000 kg cal/t iron is being lost each minute to the cooling water, and that this continues as well during a period of blast-off, during which the combustion and therefore the production of heat in the cupola is interrupted.

Each interruption in the operation reduces the heat reservoir in the cupola, and when the blast is put on again these losses will always have to be recovered first before the previously existing thermal—and therefore also metallurgical—conditions have again been reached. It may be possible, within certain limits, to reduce the quantity of cooling water during an interruption, but, in view of the absence of lining on the cupola shell, this carries a definite risk which normally will be avoided. The best results, with the type of melting furnace we have described here, will probably always be obtained with continuous melting operation, and actually this is what the cupola was primarily designed for.

Metallurgically the water-cooled hot-blast cupola without lining follows the same fundamental principles that are applicable to the formerly normal type of cupola. It is not a basic cupola, but it can be operated with either basic or acid slag, that is, with any composition of slag that is technically practicable. However, the composition of the slag can now be controlled by the composition of the charge with this type of cupola, i.e. that after consideration of the coke ash, the slag analysis no longer changes as a result of uncontrollable pick-ups from the lining components, but only through the desired metallurgical reactions.

Time and again one finds that in international literature the water-cooled cupola is referred to as a "basic" cupola. In most cases this probably represents only a lack of thought. In view of the facts that (1) lining erosion is out of the question with a cupola cooled without lining in the melting zone, (2) that for the formation of slag limestone is added to the charge and (3) that this is done more for the purpose of obtaining a fluid workable slag rather than for tying up the silicic acid in the coke ash, it is not surprising that the cupola melts with a basic slag. It is equally possible, however, to charge in addition to the limestone also some materials that contain silicic acid, such as sand or gravel, and then the slag will no longer be basic.

In our experiments we have operated such a hot-blast cupola without lining for the production of a high-silicon cast iron, using a highly acid slag for several hours. Following this there was an extra coke charge added, coupled with removal of the acid slag from the well. Then everything was switched over to basic slag and the cupola went on melting low-sulphur iron from a charge of cast iron scrap and steel scrap. To illustrate the versatility of this new type of melting furnace it may be added that during the night shift the same cupola was used to produce refined pig iron from steel scrap for the other cupola units.

It will probably be clear from what we have said so far that after the teething troubles had been overcome the new melting unit worked trouble-free metallurgically. The important influence of the metal temperature was consistently shown in all operating results so far. Thanks to the possibility of obtaining high metal temperatures in this cupola—at a coke charge of 1:6.3 and hot blast of 932 F (500 C) there were uniform temperatures up to 2890 F (1590 C)

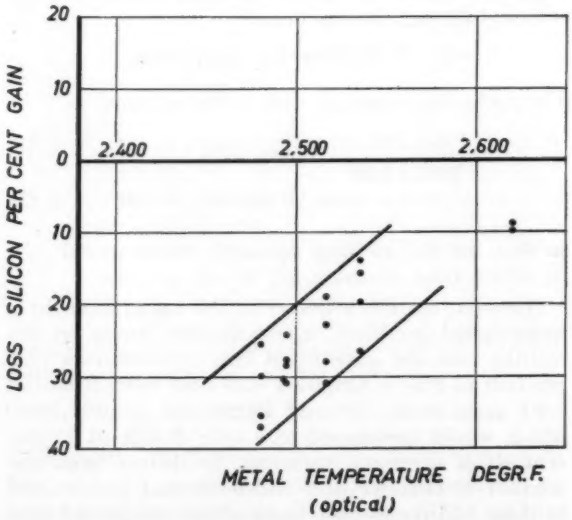


Fig. 23 — Influence of metal temperature on silicon melting losses of a hot-blast cupola operating with basic slag.

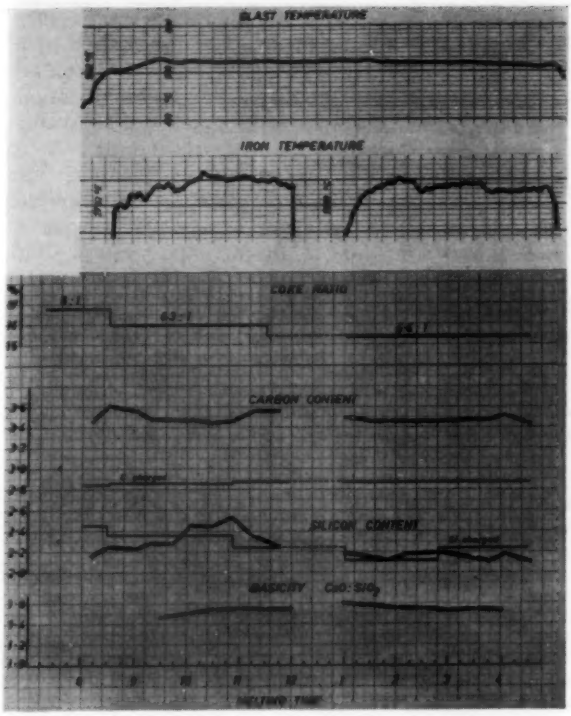


Fig. 24 — Basic melting in a liningless cupola without silicon loss.

actually measured—it is possible to keep the silicon losses low even when utilizing the advantages of basic melting.

Figure 23 shows, how, with steady basicity of the slag, the silicon losses correlate with temperature, i.e. how they diminish with increasing temperatures. To confirm this tendency a special experimental heat was melted of which the log is shown in Fig. 24. The basicity was kept practically equal (1.55) and also the blast temperature to achieve defined thermal, and consequently metallurgical, conditions in the liningless cupola To influence the iron temperature only the coke charge was several times altered.

The two silicon curves – silicon content of the charge and silicon in the metal – demonstrate that with high silicon input and still low metal temperatures there was a silicon loss while the actual silicon content of the iron was gradually increasing. A reduction of the silicon in the charge at that time did not show any noticeable effect. But as soon as the iron exceeded a temperature of about 2820 F (1550 C) there was suddenly a silicon pick-up.

Further reductions of the silicon charge and a decrease of the coke charge allowed to approach the desired silicon content of about 2.10 per cent without effecting the fact of a silicon pick-up. To confirm the assumption that, under certain conditions, it is more or less a matter of "know how" to prevent silicon losses even if the iron is melted under a middle-basic slag the silicon content of the charge was once again increased. Immediately there was again a silicon loss while the actual silicon content of the iron did not change. These experiences could be reproduced on production scale melting.

English cupolas without lining, which we already mentioned, operate at an average silicon loss of only 9 per cent and less with the use of a slag at a ratio of $CaO:SiO_2 = 1.23$. The sulphur contents there are about 0.05 per cent.

On the basis of their first operating experience Dobson & Barlow Ltd., have stopped charging pig iron in their cupola plant (with cold-blast operation the percentage of pig iron in the charge was 25 per cent) and are instead charging only cast iron scrap, return scrap and steel scrap. Utilization of all the metallurgical possibilities available in this new melting unit has not only kept the quality of the metal at the same level, but has actually improved it by two grades in British Standards. This is reflected in an increase in tensile strength from 12 tons per square inch to 14 tons per square inch and in considerably less trouble from varying wall thicknesses in the casting, along with improved castability.

Pre-melting Cupola for Steelworks

Development of this water-cooled hot-blast cupola without lining and the knowledge drawn from actual melting operations also meant a significant step forward toward the creation of a continuously working *pre-melting cupola for steelworks.* To supply open-hearth furnaces in non-integrated steelworks with liquid metal from cupola plants had already proven its value in times that rapid and large increases in production were necessary, provided that the economics and the melting process were of minor importance as, for instance, during the war. The disadvantages of the old cupola pre-melting unit, operating on cold blast, can be summarized as follows:

1. High coke consumption with 100 per cent steel scrap charge,
2. High carbon contents in the resulting liquid metal and, therefore, limitation of the amount of metal that could be charged into the open-hearth furnaces,
3. High sulphur contents of the pre-melt metal and, therefore, increased slag work in the steel furnace,
4. High lining costs with basic lining of the cupola.

The normal hot-blast cupola overcame part of these limitations. Consequently, in Europe, and particularly in Germany, the idea of pre-melting for non-integrated steelworks was soon resumed after the war. At the moment there are three hot-blast cupola plants in Germany and one in Italy operating in steelworks. The largest installation has a melting rate of 10 to 12 tons per hour. These cupolas, which are equipped with plants of system Schack-G.H.W., have furnished valuable leads, both on the economic and on the metallurgical-cupola-technical side of the integrated process in steelmaking.

It was obvious from the few reports published to date[54] that to guarantee trouble-free continuous operation it would be necessary to pay greatest attention to the cooling of the cupola, the dust problem in the recuperator, the life of the slag siphon and the lining of the receiver. Metallurgically it appeared to be necessary to reduce the carbon content of the liquid cupola metal to below 3.00 per cent.

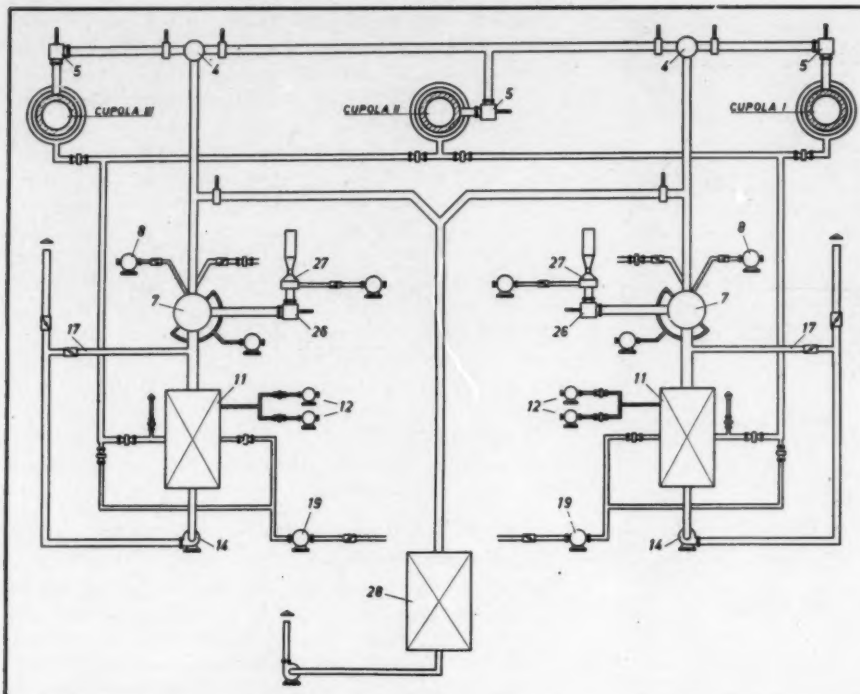


Fig. 25 — Diagram of an installed hot-blast cupola premelting unit of 1,000 tons daily melting rate for an open-hearth steelwork.

The steps taken in this direction by the author's hot-blast cupola system are probably clearly understood from what has been said. The question of the life of the slag dam seems also to have found a satisfactory solution. A pre-melting unit of this system with a daily output of 1,000 tons will go into operation in Germany in the very near future.

Figure 25 shows the diagram of this plant. Three cupolas of 25 tons per hour melting rate each are connected with two fully automatic-controlled hot-blast units. Preheating of the blast is done in cast iron recuperators by means of the cupola stack gases. The cupolas are of a shape that almost look like blast furnaces, and they are equipped with pressure cooling in the melting zone, which permits melting without lining. The heat taken off the cupolas by this cooling system is given into the work's steam grid, in the form of superheated steam at 597 psi (42 atm).

To utilize the amount of stack gas not required for blast preheating, there is a waste heat boiler in which steam at the same pressure and temperature as in the cooling elements is produced and is then added to the work's grid. This plant will probably be the largest and most modern plant of its kind in the world at the present time.

The metallurgist might be interested in this connection to know the results of some experiments with 100 per cent steel scrap in the charge, which were made in such water-cooled hot-blast cupolas without lining. The aim was to produce a liquid pre-metal of *lowest* possible *carbon content*. To facilitate further use of the metal so obtained— (it was cast into pigs), it was necessary, however, to have the silicon content somewhat higher than would have been desirable, i.e. the silicon content was approximately 1.15 per cent instead of 0.05 per cent. Each charge consisted of:

Steel scrap	—	1,232 lb
Silicon (in briquettes)	—	16 lb
Manganese (in briquettes)	—	2 lb
Charge coke	1:5.9 —	(17%)
Limestone	—	90 lb
Fluorspar	—	10 lb

The cupola was first brought up to temperature with a charge of pig iron and cast iron scrap and was then charged with 100 per cent steel scrap without an intermediate extra coke charge. With this latter charge, the metal analysis at the spout was taken each hour, and the figures were:

% C	% Mn	% Si	% S
3.10	—	—	
2.90	0.51	—	
2.82	0.71	0.95	0.115
2.50	0.42	1.05	
2.42	0.42	1.10	0.065
2.52	0.50	1.15	0.060
2.60	0.61	1.20	0.050

Leaving the initial conversion analyses out of consideration, the average carbon content was about 2.60 per cent with 0.06 per cent S. An increase in the amount of slag per quantity of iron, coupled with a simultaneous decrease in the basicity toward the acid side ($CaO:SiO_2$ = appr. 1.25), gave metallurgically favorable results. The slag moreover became more liquid and more reactive. This is in accord with the observations by T. M. Frazell and J. D. Sheley,[55] Since these results could be reproduced in a number of experiments, it is to be assumed that the prospects have improved for an increased use of hot-blast cupolas as pre-melting units in steelworks. The fact that with the operationally safe construction of such pre-melting units the possibility to erect non-integrated converter steelworks has come nearer into practical reach may be mentioned as a sideline.[35]

This survey of the development of hot-blast cupola melting in Europe and of the metallurgical possibilities available in the cupola operating with hot blast would not however be complete if the question of *ore reduction* in the cupola were not at least mentioned briefly. In Europe, M. H. Philipon,[13] J. Estivall,[13] and R. Doat[15,16] and others have already indicated the possibility of adding ore and slag to the gas producer with hot blast or to the hot-blast cupola. From India comes a report by R. A. McGregor[56] regarding satisfactory results obtained with a water-cooled cupola at a melting rate of 290 lb/sq ft/hr (approximately 4.1 t/m²h).

There has also been a recent report on the reduction of ore fines, with the use of low quality fuels.[49] In these latter experiments a cupola, with gas takeoff below the tuyere level in the well,[57] was used, with hot blast up to 1,110 F (600 C) and with Swiss anthracite containing 35 per cent ash, to produce a satisfactory iron from a charge of 50 per cent of very fine borings and turnings and 50 per cent ore of 28 per cent Fe content, without any detectable losses through dust. The sizes of the materials in the charge were extremely small;—more than 56 per cent of the charge had a size below 0.04 in. (1 mm) on the screen test. Experiments with charges of 100 per cent ores gave similarly successful results. The investigations are being continued in a new, improved experimental plant.

Summary and Conclusion

It has been endeavored to present a general survey of the development of hot-blast cupola melting techniques in Europe, and particular reference was made to the metallurgical and thermotechnical problems which have primarily influenced this development. Starting from fundamental metallurgical research and the requirements that follow from them and that have to be demanded of a modern hot-blast cupola plant, it was shown by way of example with one melting furnace system, to what extent these requirements can be met at the present stage of development.

Operating experience gathered both from hot-blast cupolas with lining and also without lining in the melting zone has shown that melting units can be controlled and regulated and have to be considered valuable metallurgical furnaces. Illustration of the metallurgical research work also furnishes a reply to a question which crops up time and again, namely: "Why is it that hot-blast cupolas in Europe use such high blast temperatures?" Not to save coke, but to make use, to the widest possible extent, of these metallurgical advantages which only come into full play with the use of higher blast temperatures in the cupola!

The design and construction of the European hot-blast cupola plants has no doubt become more complicated, but this does not apply to their operation, due to the use of automatic controls and similar means. The modern hot-blast cupola with blast temperatures of over 750 F (400 C) may require more care and scientific metallurgical observation than its modest brother operating on cold blast.

However, it pays back for this attention by maximum efficiency in quality and economics. The original crude remelting furnace of the ironfoundry has today grown up into a valuable metallurgical melting furnace which is continuously widening its application and makes its way into fields which could not possibly have been considered even only a few years ago.

Acknowledgment

It is a pleasant duty to express the sincere thanks to all who have so kindly assisted in this report, especially the Buderus'sche Eisenwerke in Germany, the International Research Group for the Hot-Blast Cupola, the Textile Machinery Makers Ltd. and Crane Ltd. in Great Britain, and last, but by no means least, Dr. A. E. Schuh, of U. S. Pipe and Foundry Co., U.S.A., at whose instigation this report was written and whose many helpful suggestions and critical review of the manuscript are deeply appreciated.

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DISCUSSION

Chairman: DR. A. E. SCHUH, U.S. Pipe & Foundry Co., Burlington, N.J.

Co-Chairman: R. A. CLARK, Electro Metallurgical Co., Union Carbide & Carbon Corp., Detroit.

H. W. GORMAN[1] *(Written Discussion):* The author and his associates are to be complimented on a thorough and carefully planned approach to the problem of developing hot-blast cupola melting techniques and the consequent satisfactory solutions of the problems encountered. Of particular interest in this connection are the results of the first day's operation of a new hot blast installation as shown in Fig. 15. It is impressive to say the least, to be able to successfully anticipate and predict the behavior of any newly installed melting equipment, particularly in view of the narrow target.

The author points out that blast conditions and lining erosion have serious effects on control. This is true also of cold blast operation where blast humidity is of definite importance. I would appreciate the author's opinion of the effect of humidity, if any, on the high temperature reactions in the hot-blast cupola and also the effect of purposely introducing moisture or steam into the blast.

On page 183 the author refers to an increase in steel scrap in the charge from 8 per cent to 37 per cent, a reduction in limestone and an increase in coke, at about 10:30 o'clock to increase the silicon content. We assume he is referring to the operation depicted in Fig. 15 and if so, question whether Fig. 15 can be interpreted as shown in the paper.

Referring to the double melting zone encountered in the unlined water-cooled cupola, the author states that the tuyeres could not be projected into the cupola more than 3 in. and the velocity of the blast was altered to correct this condition. What prevented extending the tuyeres into the cupola and what means was used to control the blast velocity?

MR. LOEBBECKE *(Reply To H. W. Gorman):* There is no personal experience available on the influence of higher moisture contents on the metallurgical reactions in the hot-blast cupola. It is known that, among others, the Russians are working hard along this line; however, reliable results have not yet been published.

It is correct that the increase in steel to 37 per cent refers to the start-up shown in Fig. 15. Unfortunately, the moment of the increase in steel was wrongly shown in the preprint; it should be at 9.30 hours, i.e. approximately one hour before the rise in silicon contents.

With regard to the deviation in blast velocity it should be explained that due to incorrect operation of the liningless cupola the tuyeres were so tightly welded into the cupola shell that they could not be moved. For this reason it was tried to obtain similar effects through the blast velocity as would no doubt have been achieved by further protruding tuyeres. Blast velocity was altered by inserting short sleeves into the tuyeres.

MILTON TILLEY[2] *(Written Discussion):* The article shows much thought and experimentation to throw light on the many problems presented in the use of hot blast in cupola melting, such as better fuel economy, hotter iron, less oxidation, faster rate of melt, more open tuyeres, less sulphur pick-up. All of these should reflect in less loss of castings in the foundry and better mechanical properties.

As for the economy, I think one can capitalize too strongly in attempting to improve all operating conditions so that between the capital expenditure and maintenance, the economy could suffer. From this standpoint, external firing with a relatively cheap fuel such as our natural gas might be considered, advantages being lower installation and maintenance cost and uniform temperature of the blast air. Thinking along these same lines, the higher temperatures, say over 750 F, require water

1. Supt. Foundry Control, Allis Chalmers Mfg. Co., Milwaukee, Wis.
2. Metallurgist, National Malleable & Steel Castings Co., Cleveland.

cooling, which may readily result in as much heat being extracted as put in by the preheated blast. Also, when the metal is subject to a pick-up of silicon from the refractories, control suffers.

I wish to express my appreciation for the valuable information presented from the experiences in Europe for it will guide our thoughts and actions on the subject in the future.

MR. LOEBBECKE *(Reply To Milton Tilley):* In connection with the economies of recuperative hot-blast cupola melting the question of maintenance costs was raised. It can only be repeated what was mentioned already in the report that additional essential maintenance costs have not as yet been found with the plants described here and have, therefore, had no influence on the economies of the plants. Neither does it comply with actual facts that water cooling is required with the use of higher blast temperatures of, for instance, over 750 F. This is proved by more than 400 cupola plants of different manufactures of this kind which have been in operation in Europe to entire satisfaction. It should be pointed out, however, that even if for instance the tuyeres are water cooled the heat losses through the cooling will by no means relinquish the effects of the higher blast temperature.

As was shown in Table 1 or respectively in the report on page 174, the influence of the amount of heat brought into the cupola by the blast is far less effective than the physical effect of hot blast, for example in the increase in combustion temperature of the coke. The highly preheated blast results in a shallower melting zone which is, moreover, drawn farther toward the tuyeres. This causes a more concentrated production of heat per unit of time and volume which again results in a higher temperature of the goods to be smelted. This effect remains also where a considerable amount of heat is taken off the outer zones of the cupola like, for instance, in a liningless cupola. It has been shown that also in such cupolas temperatures of more than 2900 F. could be obtained without difficulties.

S. TUNDER[3] *(Written Discussion):* I am very happy to confirm that the details and data dealing with the *acid* lined hot blast cupolas coincide entirely with our experience during the last 15 years covering the operation of more than 180 furnaces.

However, so far as hot-blast cupolas operated with *basic* slag is concerned, from our experience with all our furnaces of this type, the matter is really not so complicated as it appears from this paper. The problem of operation, both in a metallurgical and economic sense, can now be regarded as absolutely solved. The modern hot blast cupola requires no more careful handling or metallurgical observation than that required by a cupola operated on cold blast. With regard to metallurgical observation, one of our smaller customers sends his ladle tests for analysis to an outside laboratory about once a month, getting his results back in two to three weeks. Despite this, his rejects have been reduced by 75 per cent as compared to his former cold blast operation.

It is gratifying to know that Mr. Loebbecke is following our path in adopting the hot-blast cupola as a premelting installation for steel plants.

In the early days of our development and research work, we struggled to determine the way to run a hot-blast cupola with 100 per cent steel scrap charge and to get the lowest possible silicon content of 0.05 per cent. It is easy to produce an iron with 1 per cent silicon content as the melting time required is very much less than is the case in producing low-silicon iron. One of our first furnaces installed in a steel plant proved to us that when making the required low-silicon hot metal, the capacity of the furnace was 35 per cent lower than when making 1 per cent silicon iron.

This fact is not fully recognized in the planning of the three 25-ton furnaces mentioned in this paper. I would suggest that the calculation of the design should be revised in order to avoid the grave and expensive disappointments we have experienced. I should be very glad to place at the disposal of Mr. Loebbecke the construction details and calculations of a 26-ton furnace which we are building and which will commence operation in a few weeks. These calculations are the result of the careful and gradual development from 8 tons per hour through 15 tons per hour. These developments enable us to be absolutely sure that this furnace, when it commences operation, will actually, effectively and satisfactorily produce 26 tons per hour of low-silicon hot metal with the extremely low sulphur content which is necessary for the operation of the open-hearth shop.

In view of this, I would like to caution Mr. Loebbecke that it would be false to base his calculations on the trial analyses, as a metal with carbon as low as 2.5 per cent cannot be produced with the absence of silicon. In order to obtain low sulphur, as is necessary for the open-hearth practice, the carbon content is bound to increase to at least 3.5 per cent. It is, of course, possible to keep this carbon down at a low level, but in so doing, the sulphur and silicon content rises to such an extent that the economics of the duplex process of hot-blast cupola and open-hearth is radically disturbed.

The history and the outlook of the hot-blast cupola as presented in this paper is welcome. The hot-blast cupola is most certainly an excellent metallurgical apparatus with many-sided possibilities, but with regard to smelting ore it is perhaps inclined to lead us astray in our thinking. I should like particularly to stress this point. The hot blast cupola furnace must *not* be regarded as an apparatus for melting ore by a method which cannot be economically justified. It is certainly possible to smelt ore by this method, but then the cupola becomes a miniature blast furnace and, do you not think that we are putting our much vaunted progress in reverse gear by reducing the capacity of our furnaces by 80 per cent?

W. R. JAESCHKE[4] *(Written Discussion):* First, I would like to join the other discussors in complimenting the author on the excellence of his paper and its presentation.

The paper emphasized points of difference when operating cupolas with blast temperature below or above 750 F. The reason for this sharp line of difference is in the economics of hot blast equipment and operation. As you all know, our common metals of construction begin to drop in their physical properties markedly at that temperature range and for preheaters to serve at above 750 F or 800 F high heat-resisting metals of construction are required or maintenance costs could be excessive. Even the blast piping and windbox must be of high heat resisting metals or lined with refractory material. These extra considerations for use of high blast temperatures are reflected in higher initial costs of the equiptment and possibly in higher maintenance costs.

Another point of difference is that it is quite commonly known that all the metallurgical reactions in a cupola will proceed more rapidly and to a greater degree with the higher temperatures of the combustion resulting from higher temperature blast. If however, these reactions or their results are not controlled, they may prove to be an economic disadvantage rather than an advantage. The primary purpose of the foundry cupola is to produce iron satisfactory for castings to engineering specifications and the need for improved control becomes more and more apparent, as time goes on.

Also in this country, steel scrap is so high in price that some operators have indicated they are using more pig iron than formerly because the steel scrap is higher in price than the pig iron. One such case was reported to me from a district where market reports indicate that steel scrap is still priced advantageously below pig iron.

Under favorable market conditions, it would be possible to make savings as indicated in this paper, but then a fired or electrically-heated reservoir of large holding capacity would be required to produce the required uniformity in composition and temperatures to produce castings to specifications. This would increase the capital investment but otherwise show operating savings.

The author mentioned in his paper that F. K. Vial was among the first to recognize the metallurgical possibilities of melting with preheated blast, while others still considered coke savings as the most important economic possibility. Actually while F. K. Vial had originally intended to operate with blast temperatures of 800 F and reduce the sulphur in all scrap car wheel iron, the economics of preheater construction and maintenance, plus the advent of the soda ash desulphurizing process influenced a change to a blast temperature of 600 F. On my latest visit to the Griffin Car Wheel Co. I found them operating at 450 F to 500 F because of improved product control at that temperature. That is the experience also of other operators of hot blast cupolas in U.S.A.—better control at reasonable blast temperature and economic justification in appreciable coke savings, reduced oxidation, and better operation due to practically no bridging at the tuyeres.

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3. Technical Director, GHW, E. S. Harman Corp., Chicago.
4. Cons. Met. Engr., Whiting Corp., Harvey, Ill.

Operation with basic slags introduces new and additional variables that will require attention for control.

I agree with the author that more care and scientific metallurgical supervision is required when operating cupolas at above 750 F blast temperature and especially so when operating with basic slags.

One operator in U.S.A. operating at 600 F with basic slag has added three supervisors to his cupola crew; while another, the first to operate a liningless, water-cooled cupola with basic slags in U.S.A., has changed back to acid lining in the cupola well and acid slags to achieve the necessary control for his castings.

The author mentions the importance of a closely defined height of slag in the well at all times in order to control the metallurgical process in the cupola and claims that this is obtained through the use of a syphon type taphole. Such a tapping system is basically the same as our front slagging system and a closely defined height of slag can only then be obtained if the resistance to the blast remains constant. This resistance is a function of the blast volume and the coke size. Rarely in commercial cupolas in this country are either the blast volume or the coke size constant and thereby we have varying amounts of slag in the cupola well and consequently varying metallurgical conditions.

MR. LOEBBECKE *(Reply To W. R. Jaeschke):* The renewed question of maintenance costs has already been answered. In addition it may be pointed out that for temperatures up to approximately 1000 F it is not necessary to use heat-resisting materials for the blast piping and wind box. These parts are mostly insulated for economic reasons; the costs thus incurred should be negligible in view of the higher cupola temperatures obtained with hot blast and of the metallurgical advantages demonstrated.

It is certainly correct that with increasing blast temperatures and, therefore, increasing temperatures in the cupola all metallurgical reactions proceed more rapidly and that there are also such reactions as would not be possible with lower temperatures or in the cold blast cupola. Any fears that these reactions cannot be controlled should have been proved to be unfounded by the examples given. It may be added that all practical examples mentioned are based on analyses from metal samples off the spout. It is not true, therefore, that a balance of analyses takes place only in the forehearth.

The only comment to the mention of F. K. Vial can be that more than 30 years of technical development have passed since the time that Vial made his experiments so that what Vial considered to be impossible can be, and actually is, realized in our times.

The far-reaching control of the cupola with acid slag, which has been mentioned earlier, is also possible with basic operation as is shown in Fig. 24.

As regards the height of slag, the comments made by Mr. Jaeschke are principally correct. However, if we speak of a constant slag level then we mean a slag level which fluctuates to the least possible extent around a straight line level and is not uncontrollable like in the experiment shown in Fig. 2.

W. W. LEVI[5]: When operating a neutral-lined cupola with hot blast, the maintaining of close carbon control when producing irons with less than about 3.65 per cent total carbon is somewhat more difficult than in conventional acid practice. With our particular practice, the cupola is quite sensitive to shut-downs, changes in melting rate and the like.

When operating on the strongly basic side with hot blast, we produce irons containing well in excess of 4 per cent total carbon and under these conditions control of the cleanliness of the charge materials is essential. Introduction of such siliceous materials as sand, core butts, etc. must be prevented. Most contaminants which find their way into the cupola are acid in nature, and will upset the slag chemistry. The papers by Mr. Heinrichs and Mr. Loebbecke have illustrated the necessity of maintaining close control of slag composition.

MR. LOEBBECKE *(Reply To W..W. Levy):* The statement that a liningless cupola is more sensitive to interruptions, alterations in the melting rate, etc. is essentially confirmed by our own experience. Incidentally, this should not really come as a surprise because during each minute of standing time there are about 10,000 kgcal. taken off through shell and tuyere cooling while in the cupola there is no heat produced by combustion. For this reason a liningless cupola will always work best where it is operated continuously.

H. M. ASPINALL[6]: Is there any method by which the economics of a hot blast installation can be determined through inexpensive experiment without actually installing the equipment?

MR. LOEBBECKE: There is no such inexpensive method other than a review of the experience of others with such equipment.

W. A. PETERS[7]: What degree of accuracy is necessary in weighing the cupola charge, and does this vary for hot or cold blast operation?

CHAIRMAN SCHUH: The degree of accuracy in commercial cupola operation might be approximately 5 per cent. A higher degree of accuracy is needed in weighing materials which are unlike the product in chemical analysis. Thus, precise weighing of pig iron or scrap which is similar in analysis to the product is not as essential as is the case for steel scrap which is far different in chemistry. Of course, the weight of alloy additions must be controlled closely.

B. C. YEARLY[8]: It is common practice to melt malleable iron for cupola-air furnace or cupola-electric duplex processes in the acid-lined cupola. There might be difficulty in producing a low-carbon in the range of 2.6 per cent carbon, 1 per cent silicon iron in an acid-unlined cupola operated with high temperature blast. Is there any commercial production of low-carbon iron in this type of melting unit?

MR. LOEBBECKE: At the moment, there is no commercial production of malleable iron in the type of hot-blast cupola equipment described in this paper. For the present, I would not recommend acid unlined operation for the melting of high grade malleable. Such a plant is in use as a steel mill premelting unit. In this particular application, a low-carbon iron is melted but close control of the chemistry of the product is not so essential.

MR. LOEBBECKE *(Author's Closure):* It was not intended to record results of routine work for the production of cast iron of various qualities and analyses as has been done for years in hundreds of hot-blast cupola plants in Europe. The object was rather to show how easily and quickly these routine conditions can be achieved and what further metallurgical possibilities present themselves by the use of hot blast of higher temperatures. *Smelting in the hot-blast cupola of higher blast temperatures need not be an art! The fact that it can be made into an art by using new ways in metallurgical respect that were impossible to the cupola before shows only the value and importance of this new melting technique to foundry industry.*

Finally it may be pointed out that the objections made by S. Tunder against the operation of a liningless, continuously melting hot-blast cupola for the production of low-carbon iron for steelworks have been superseded by practical experience. The melting plant referred to (Fig. 25) was put into operation some months ago and has fulfilled all expectations mentioned in this report. The weekly average analysis of the metal molten from steel scrap was: C, 2.6%; Si, 0.3%; S, 0.06%.

There was no difficulty in using this metal in continuous operation in the open-hearth furnaces for the production of high-quality steel. A new large field has thus been opened to the hot-blast cupola.

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5. Chief Metallurgist, Lynchburg Foundry Co., Radford, Va.
6. Dominion Iron & Steel Ltd., Sydney, Nova Scotia.
7. Supt. Mfg., Laclede Steel Co., Alton, Ill.
8. Engr. Asst. to V. P., National Malleable & Steel Casting Co., Cleveland.

MALLEABLE BASE SPHEROIDAL IRON*

By

F. B. Rote,** E. F. Chojnowski*** and J. T. Bryce****

ABSTRACT

Malleable base spheroidal iron is a new high-strength alloy having graphite in a spheroidal form in a highly stable matrix. The principal alloy is sulphur, free or combined, added either during melting or as a ladle addition. Annealing of the sulphur-stabilized iron is accelerated by concurrent addition of a solid-state graphitizer. As sulphur content is increased above conventional founding levels, the internal graphite structure undergoes gradual transitions from temper carbon to rounded agglomerates, to duplexed spherulitic-temper carbon spheroids, to true spherulitic.

The addition of sulphur stabilizes carbide and permits the production of thick metal sections free of primary graphite. At the sulphur levels required to produce spheroidized graphite the fluidity of the alloyed iron is considerably greater than an unalloyed base, making pouring of thin metal sections feasible. The alloying elements are low cost and free of explosive reaction during ladle addition. Because sulphur produces a true alloying effect, no loss of effectiveness is encountered while the molten metal is held in the furnace or ladle. Excellent casting surfaces are produced in either green sand or shell molds.

The metal is heat treated by conventional anneal, quench and draw processes. During draw treatments it exhibits a strong resistance to second-stage graphitization or segregation of carbides. Uniform matrix structures are produced providing high tensile and yield strength and ductility. The modulus of elasticity is uniformly high at 25 to 27 x 10^6. Machinability is excellent at high strengths.

Introduction

Engineers currently recognize three graphite types in high-carbon ferrous materials which span a wide range of properties as a result of differences in the size and shape of the graphite, and variations in matrix composition and microstructure. Conventional gray iron contains graphite in a flake form which may vary with composition, cooling rate, and melting practice in quantity, size, shape, and distribution. This produces a variety of properties and, generally, a low level of ductility and toughness due to the notching effect of the flake shape.

Recently, a material containing graphite in the spheroidal or nodular form became commercially available with higher strength and toughness than are obtainable in conventional gray iron. This material, known as nodular iron, can be heat treated to produce high levels of strength equal to, or greater than, those obtainable in some steels. It is produced by the addition of rare earth metals or oxides to a low sulphur base iron, and can be cast with the spheroidal graphite in the free form. This material greatly extends the application of cast ferrous metals in modern industry, since wide ranges of properties, from high ductility to high strength, can be obtained by heat treatment and composition control.

For many years ferritic malleable iron, containing its graphite in temper carbon form, essentially rounded shapes with protrusions developed during formation, has been available to industry. More recently, materials bearing the general name, pearlitic malleable, have been developed in which the matrix is controlled with respect to carbon content and microstructure through alloying or heat treatment. This material also possesses a wide range of properties in essentially the same level obtainable in nodular iron. Ferritic and pearlitic malleable irons require heat treatment to convert combined carbon of the white iron casting, produced in the foundry, to temper carbon.

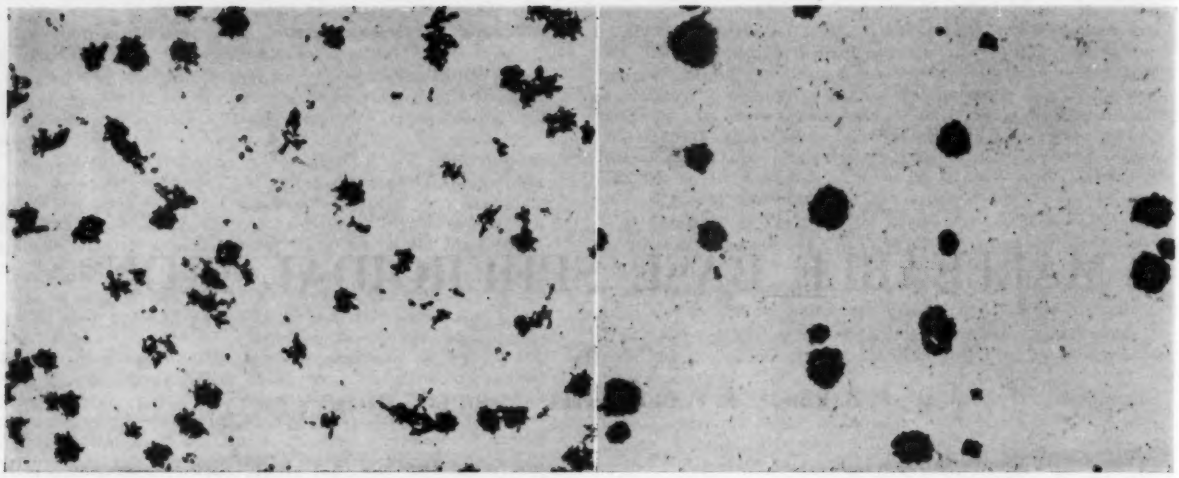
Recently, research at the authors' plant on improvements in ferritic and pearlitic malleable castings has led to the development of a new material referred to as malleable base spheroidal iron. Generally, its properties are similar to those developed in conventional pearlitic malleable with certain notable exceptions, which make it a unique material with considerable potential advantages.

Malleable Base Spheroidal Iron

Malleable base spheroidal iron is produced from the conventional duplexed metal used in most pearlitic and ferritic malleable production. The process, utilizing this base material, involves the addition of sulphur in the free or a combined form along with a solid-state graphitizer. As the sulphur content of the base iron is raised from conventional malleable levels, the graphite in subsequently annealed castings

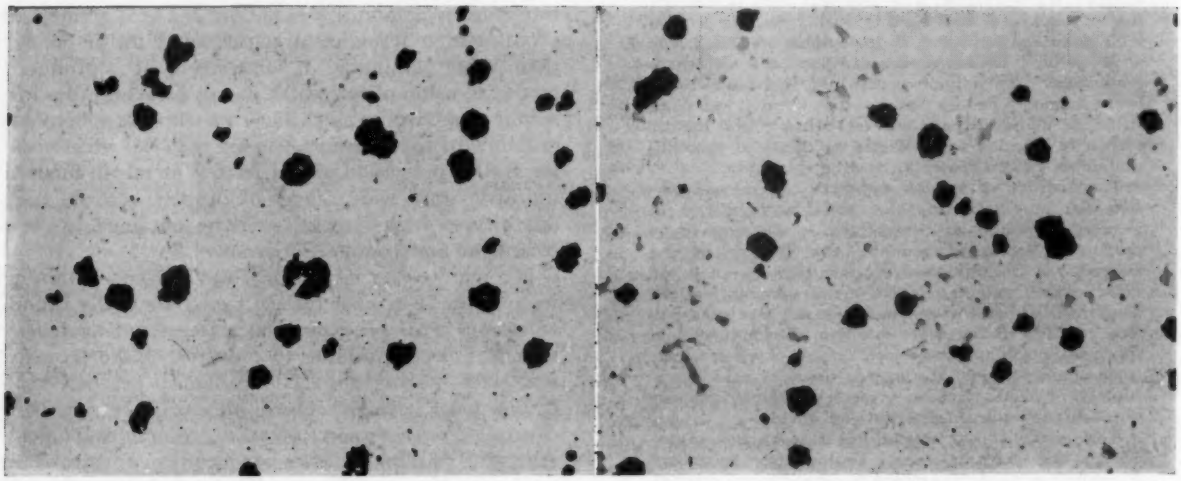
*U.S. Patent Applied For.

**Technical Director, ***Research Metallurgist and ****Chief Metallurgist, Albion Malleable Iron Co., Albion, Mich.



Conventional Malleable Iron Temper Carbon

Agglomerated Rounded Temper Carbon



Duplex Spheroids with Temper Carbon

True Spheroids

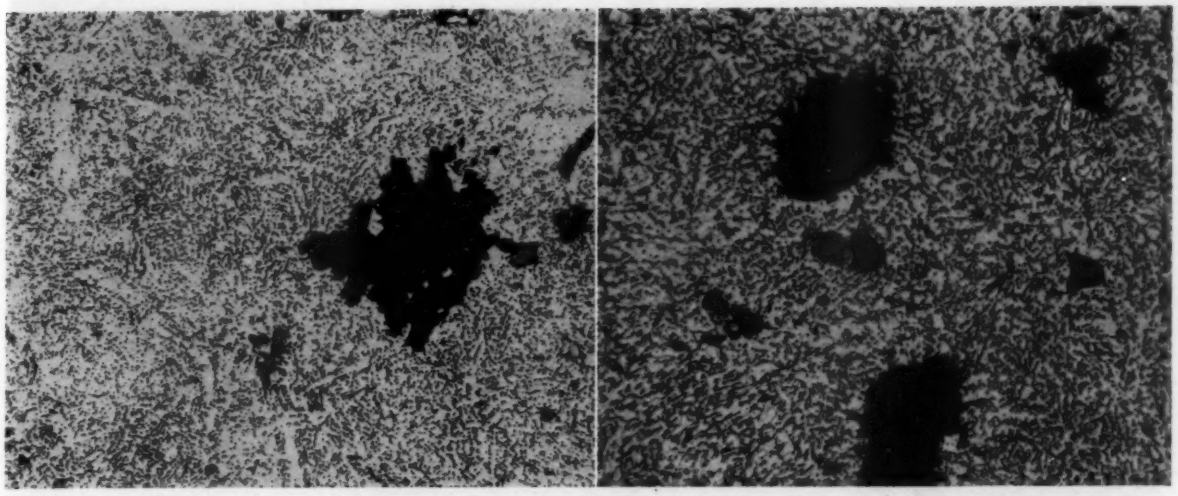
Fig. 1 — Change in graphite structure with increasing sulphur content in annealed malleable base duplexed iron. Unetched. Mag. — 100X.

undergoes a gradual transition from the conventional temper carbon shape to a true spheroidal shape. The first evidence of this phenomenon, with increase in sulphur content, is the agglomeration of the temper carbon nodules into a spheroidal shape with random structure. Further increases in sulphur produce a second transition, a duplexed carbon particle. The center of the particle is spherulitic, while the outer layers are randomly distributed carbon producing tight, rounded graphite. At still higher sulphur levels, true spherulites are formed. These structures are illustrated in Figs. 1, 2, and 3.

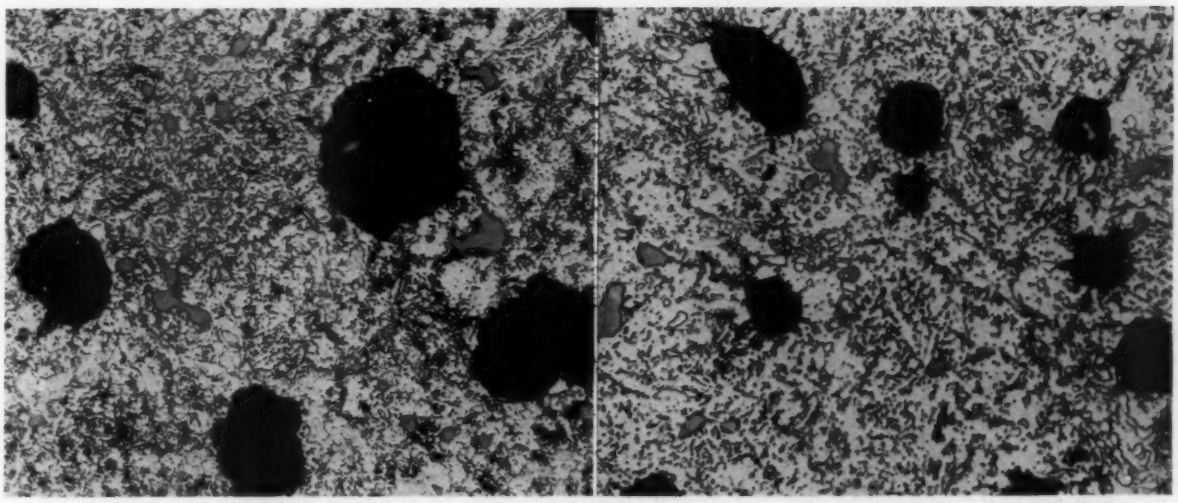
The graphite shown in the upper left photograph of Fig. 1 approaches a rounded configuration, except for the extensions on all surfaces produced late in the first-stage annealing process. In the upper right-hand photograph these extensions are considerably reduced and a more nearly spheroidal graphite pattern has been developed. However, as in the case of conventional temper carbon, the carbon distribution within the graphite particles has a random nature. In the lower left-hand photograph a tighter carbon particle produced by a combination of spherulitic and random distributions is illustrated. The lower right photograph shows the dense, ordered spherulitic structure.

The structural patterns of the graphite are demonstrated further in Fig. 2, which shows the particles at a magnification of 500 diameters, and permits visual identification of the four different configurations. It will be noted in the lower left-hand photograph that the centers of the graphite particles appear relatively smooth and dense, while the outer layers are rough and less dense. The lower right-hand photograph shows the completely dense structure of the spherulitic graphite in malleable base spheroidal iron.

The above is further demonstrated in Fig. 3, which illustrates the graphite structures as photographed under polarized light at 1000 magnification. The striated structure typical of spherulitic graphite is clearly visible in the center of the duplexed spheroid



Conventional Pearlitic Malleable | Agglomerated Rounded Temper Carbon



Duplex Spheroids with Temper Carbon | True Spheroids

Fig. 2 — Change in graphite structure with increasing sulphur content in hardened and drawn malleable base duplexed iron. Etched. Mag. — 500X.

and throughout the spherulites shown in the lower photographs.

By necessity, research work to date has been directed along many channels including:

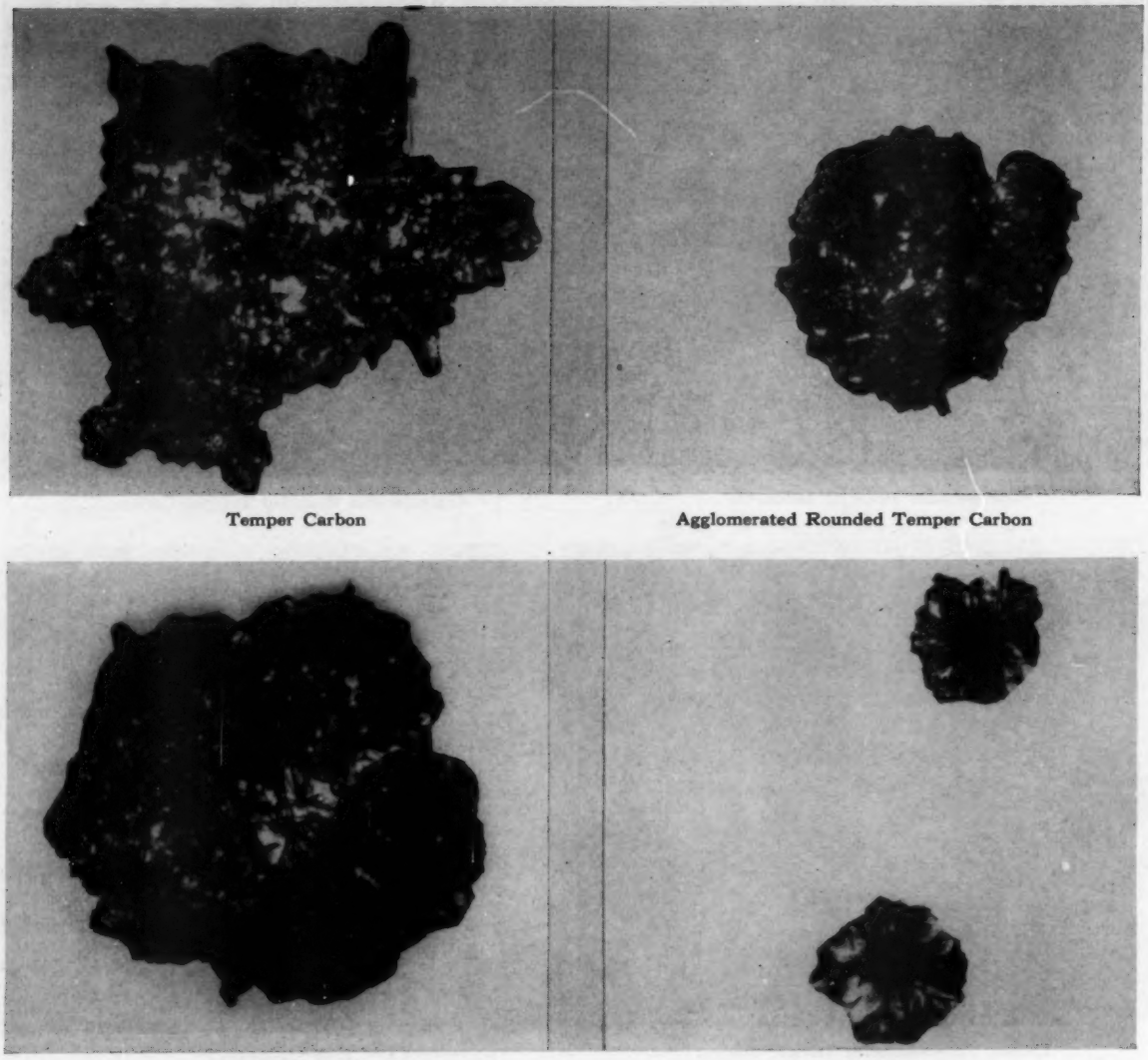
1. Control of iron composition to produce consistently each of the three spheroidal graphite types illustrated in the photographs.
2. Liquid and solid-state graphitizers to accelerate annealing of the sulphur-stabilized iron.
3. Control of annealing conditions to develop a desired combined carbon content for subsequent heat treatment.
4. Heat treatment control.
5. Evaluation of the properties of the material.
6. Production of commercial castings.

The data presented in the succeeding sections of this paper are taken from one phase of the work and are confined to a group of test bars and castings made from iron treated to produce a graphite structure which is of the duplexed type. Similar information for spherulitic and rounded agglomerated graphite types is being collected.

Experimental Procedure

A series of green-sand molds for ASTM malleable test bars and 1x4x4 blocks was prepared and poured with duplexed cupola-air furnace iron of the following base composition: 2.34 carbon, 1.65 silicon, 0.43 manganese, 0.143 sulphur, 0.03 phosphorus, and 0.03 chromium.

The base iron was treated with sulphur and a graphitizer to produce the duplexed type graphite. This particular base iron required approximately 0.40 per cent sulphur to produce the desired graphite structure. The castings were annealed in a radiant tube fired, atmosphere controlled pearlitic malleable annealing furnace. First-stage annealing was at 1720 F. The work temperature was dropped to 1550 F before discharge, when the castings were air cooled.



Temper Carbon | Agglomerated Rounded Temper Carbon

Duplex Spheroids with Temper Carbon | True Spheroids

Fig. 3 — Structure of graphite types in malleable and malleable base spheroidal iron. Polarized Light. Mag. — 1000X.

After air cooling, the castings were heated to 1575 F for 45 minutes at temperature, quenched in oil, then separated into groups and drawn at temperatures of 1100, 1200, 1300, and 1345 F for time periods between 1 hr and 19 hr at temperature. The test bars were turned to 0.505-in. diameter and tested with SR4 strain gages on one bar of each pair for stress-strain determination. The 1x4x4 blocks were retained for machinability tests.

Relation of Draw Time and Temperature to Mechanical Properties (Table 1)

As indicated above, four draw temperatures and several draw times ranging from 1 hr to 19 hr, were investigated. Figure 4 shows the relationship of time and temperature to brinell hardness for the oil-quenched malleable base spheroidal iron. The hardness after quenching in oil was 555 to 600 bhn (2.6 to 2.5 mm diameter). At all temperatures the hardness dropped considerably from the quenched hardness in 1 hr or less at temperature. The hardness obtained in even 16 hr at 1100 F remained relatively high, 269 bhn. At 1200 F and 1300 F the hardness developed with short-time draw treatments was in a level considered optimum for the best combination of strength and good machinability, and in times ranging from 4 to 16 hr controllable intermediate hardness levels were obtained. At 1345 F hardness decreased rapidly at short draw times, then leveled off at an intermediate level.

Two significant features of malleable base spheroidal iron are illustrated in Fig. 4. Draw heat treatments in the temperature range of 1200 to 1345 F produced essentially the same hardness, regardless of temperature, for draw times of 4 hr or longer. This indicates a lack of sensitivity of the material to draw temperature, and indicates the possibility of a high level of hardness controllability. In addition, it will be noted that the iron exhibited a high degree of stability on prolonged heating at draw temperatures in a range of

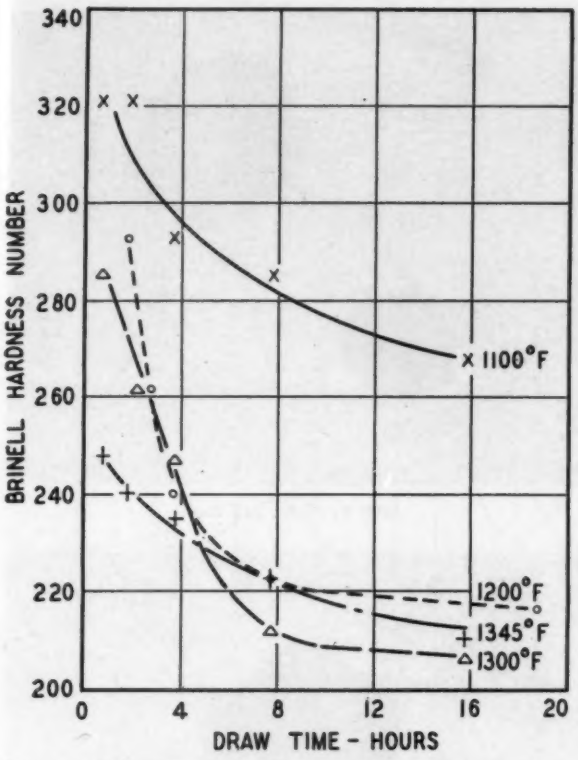


Fig. 4 — Time-Temperature-Hardness Relationships for Oil Quenched and Drawn Malleable Base Spheroidal Iron.

1200 F to 1345 F. This is evidenced by the flat nature of the curves after 8 hr at temperature.

An explanation for the hardness stability of the material on prolonged drawing is shown in Fig. 5, which illustrates the structures developed by long-time treatment at temperatures from 1100 F to 1345 F. Essentially no second-stage graphitization took place and the structures of the bars drawn at 1200 F to 1345 F are very similar, consisting of uniformly distributed spheroidized cementite. This indicates that the iron reached a stable condition requiring extremely long heating for further alteration. Consequently a desirable situation exists, as high levels of uniformity of matrix microstructure can be developed with concurrent high levels of properties and freedom from stress.

Relationships of tensile properties to draw temperature and time are shown in Fig. 6. Charts for 1200, 1300, and 1345 F with tensile strength, yield strength, and elongation plotted against time at temperature are included. For comparative purposes, the same properties for oil quenched and drawn pearlitic malleable handled under the same heat treatment conditions are shown in the charts for 1200 and 1300 F. Drawing malleable base spheroidal iron at 1200 F for extended periods produced a decrease in tensile strength from 120,000 psi to somewhat under 105,000 psi, while the yield strength decreased only slightly from somewhat over 90,000 to about 87,000 psi. The elongation increased moderately from 4 per cent after 2 hr at temperature to 5½ per cent after 19 hr at

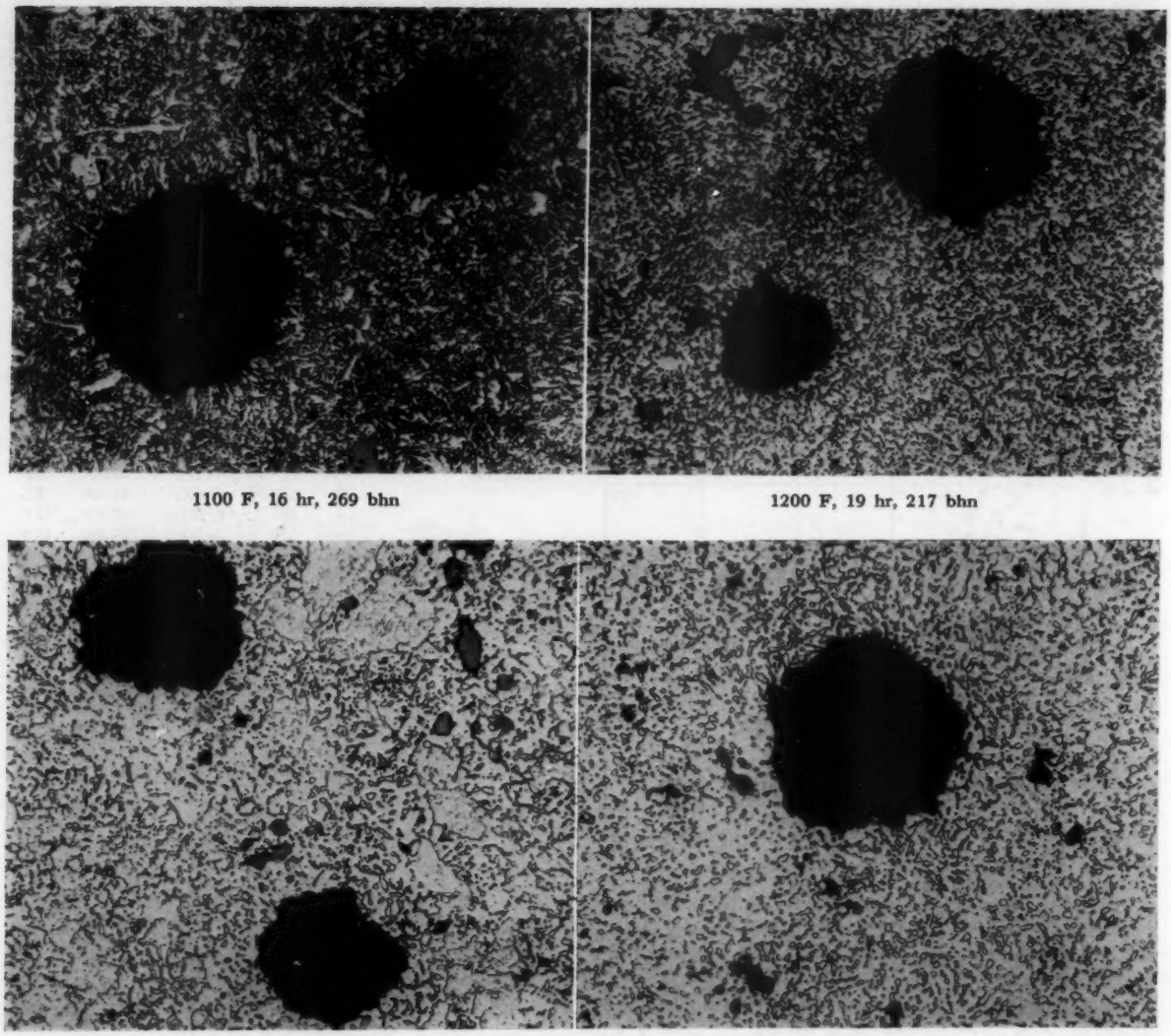
Table 1 — Time-Temperature-Properties Relationships for Oil Quenched and Drawn Pearlitic Malleable and Malleable Base Spheroidal Irons

Material	Draw Temperature, F	Draw Time, Hrs	BHN	Mechanical Properties (Averaged) Tensile Strength, psi	Yield Strength, psi	Elongation, %
Pearlitic Malleable	1100	1	321	123,000	111,000	2.5
		2	293	128,500	110,000	3.0
		4	277	117,300	99,000	4.5
		8	241	109,000	89,300	6.0
		16	228	104,000	75,300	7.5
Malleable Base Spheroidal	1100	1	321	123,500	101,500	2.5
		2	321	126,700	115,200	2.5
		4	293	112,000	97,300	3.5
		8	286	117,500	102,500	4.0
		16	269	101,000	97,300	4.5
Pearlitic Malleable	1200	2	255	108,000	90,200	5.5
		3	228	104,700	84,000	6.0
		4	217	97,400	81,100	6.5
		8	197	81,000	62,800	12.0
		19	179	56,200	39,200	16.5
Malleable Base Spheroidal	1200	2	293	117,600	92,000	4.0
		3	262	121,000	88,200	5.0
		4	241	116,000	88,100	4.0
		8	223	105,000	91,600	4.5
		19	217	104,500	87,900	5.5
Pearlitic Malleable	1300	1	196	109,000	78,000	7.0
		2½	183	89,500	70,600	8.0
		4	163	80,500	62,200	11.0
		8	131	60,800	45,100	14.5
		16	126	54,700	42,400	15.5
Malleable Base Spheroidal	1300	1	286	107,700	92,100	4.5
		2½	262	105,000	86,900	5.0
		4	248	109,500	79,600	6.0
		8	212	98,300	76,600	Flaw
		16	207	99,400	76,000	6.0
Malleable Base Spheroidal	1345	1	248	111,000	91,900	5.5
		2	241	106,500	90,200	5.5
		4	235	98,900	82,400	Flaw
		8	223	99,500	79,300	6.5
		16	212	97,300	77,300	7.0

temperature. After about 8 hr, little further change in properties occurred with increasing time at draw temperature.

Similar decreases in tensile strength and yield strength accompanied draw time at 1345 F. The tensile strength stabilized at between 95,000 and 100,000 psi, and the yield strength between 75,000 and 80,000 psi. The elongation increased somewhat more with the high draw temperatures, reaching a maximum of 7 per cent at 1345 F. The stability of properties with extended draw times parallels the hardness stability discussed previously.

The time-property curves for pearlitic malleable iron show high levels of properties at short draw cycles, and rapidly decreasing strength, and increasing elongation with extended draw. This is typical of pearlitic malleable and nodular irons which can, at elevated temperature, undergo concurrently two structural reactions: agglomeration of cementite, and second-stage annealing. Thus, with extended draw, a partial or fully ferritic structure may be produced having properties consistent with the structure and hardness.



1100 F, 16 hr, 269 bhn

1200 F, 19 hr, 217 bhn

1300 F, 16 hr, 207 bhn

1345 F, 16 hr, 212 bhn

Fig. 5 — Microstructure of oil-quenched malleable base spheroidal iron drawn for long time periods at various temperatures. Etched. Mag. — 500X.

As shown in Fig. 6, the high level of properties developed for pearlitic malleable is about the same as that developed by malleable base spheroidal iron. However, conventional pearlitic malleable is time and temperature sensitive, and requires good control during heat treatment. The lower sensitivity of malleable base spheroidal iron lessens the control requirement, and provides certain special advantages for this material. For example, since the matrix is uniform and temper resistant, selective hardening may be accomplished without fear of excessive softening beneath a hardened surface layer. Further, a high degree of uniformity in the heat treated structure can be expected as a result of the initial uniformity of the quenched and drawn base iron.

Figure 7 shows the composite chart relating tensile properties to brinell hardness for more than 500 bars tested in the present research. A range is shown for each property: tensile strength, yield strength, and elongation. Normal variations in research of this nature develop a certain spread in the properties. However, a variable not usually encountered in such work is actually present in a study of malleable base spheroidal iron, and produces a somewhat greater spread than would be expected. This variable is associated with differences in internal graphite structure, despite a spheroidal shape. It has been determined that some influence of graphite structure is reflected in the tensile properties of test castings. The relatively wide ranges in graphite structure of the bars for which data are included in Fig. 7, from rounded agglomerated temper carbon to spherulitic, partially account for the range in properties.

At high levels of hardness, malleable base spheroidal iron shows about the same range of properties as commercial pearlitic malleable produced by oil quenching and drawing. At intermediate hardness ranges, 200 to 240 bhn, much higher properties are obtained in the malleable base spheroidal iron. This is due to the uniformity and stability of the alloyed matrix.

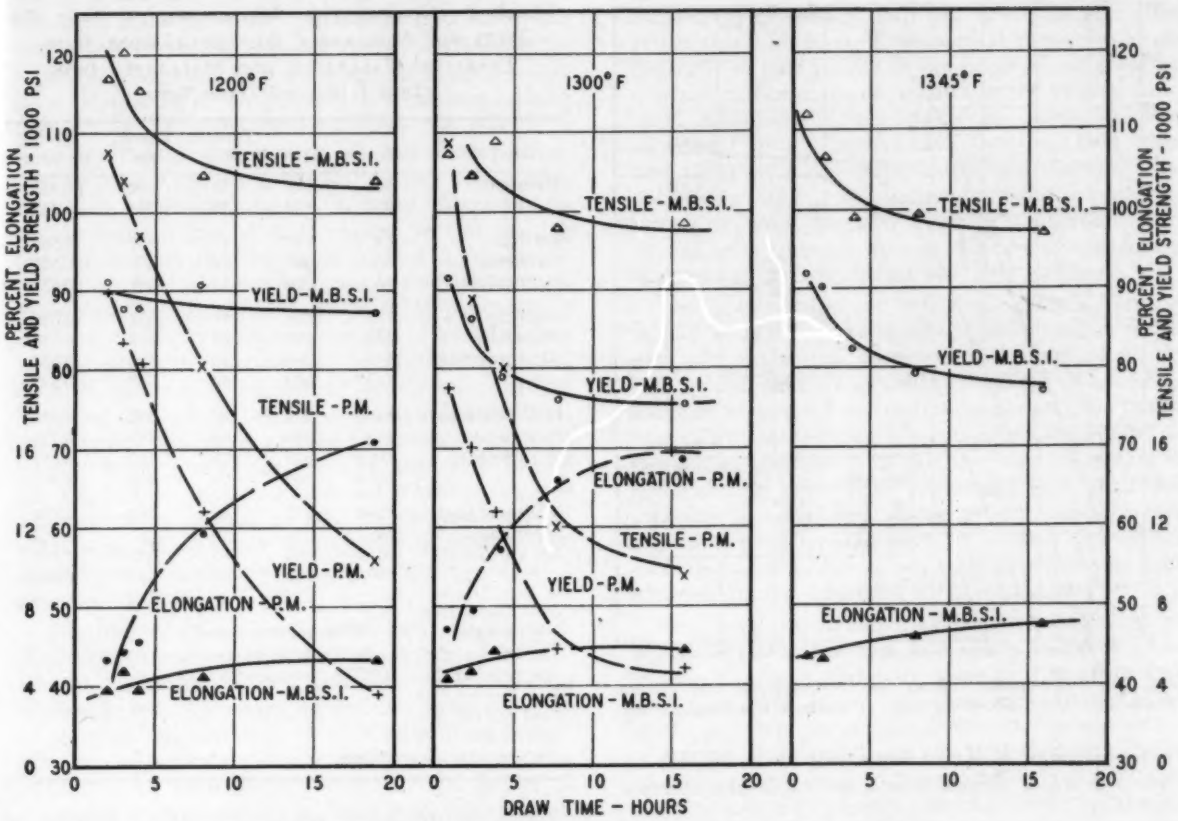


Fig. 6 — Time-Temperature-Properties Relationships for Oil-Quenched and Drawn Pearlitic Malleable and Malleable Base Spheroidal Iron.

Modulus of Elasticity

In the design of many machine structural components, a uniformly high rigidity in the structural material is essential. Materials exhibiting high levels of properties and low modulus of elasticity suffer in such applications. Because of the relatively low and closely controlled carbon content of malleable iron, this material provides an excellent base for high strength derivatives produced by heat treatment or alloying. The excellent base material, in combination with the sulphur alloy – which provides a high degree of uniformity in the matrix structure – produce high elastic modulus castings. This is demonstrated in Fig. 8, which shows stress-strain curve for malleable base spheroidal iron at four hardnesses between 321 bhn and 179 bhn. Tensile strength ranged from 123,500 to 95,900, yield strength from 118,500 to 68,000, and elongation from 2 per cent to 7 per cent. Elastic modulus was uniformly in a range of 25.4 to 27.0 x 10^6. This is an important property of the new material.

Machinability

A limited amount of quantitative machinability test data has been completed on malleable base spheroidal iron. Considerable experience in turning and finishing test bars and test castings has been obtained in the laboratory, and it is indicated that freely machinable structures are obtained at hardnesses lower than about 280 bhn. At the same hardness,

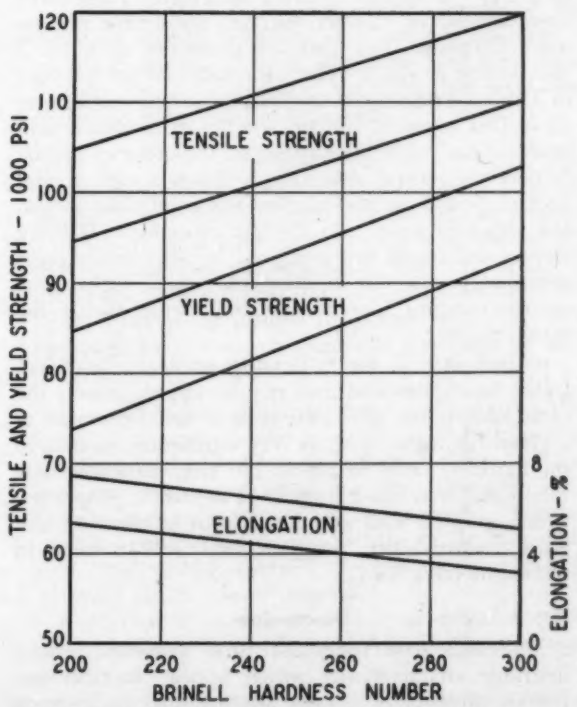


Fig. 7 — Brinell Hardness-Mechanical Properties Relationships for Oil Quenched and Drawn Malleable Base Spheroidal Iron.

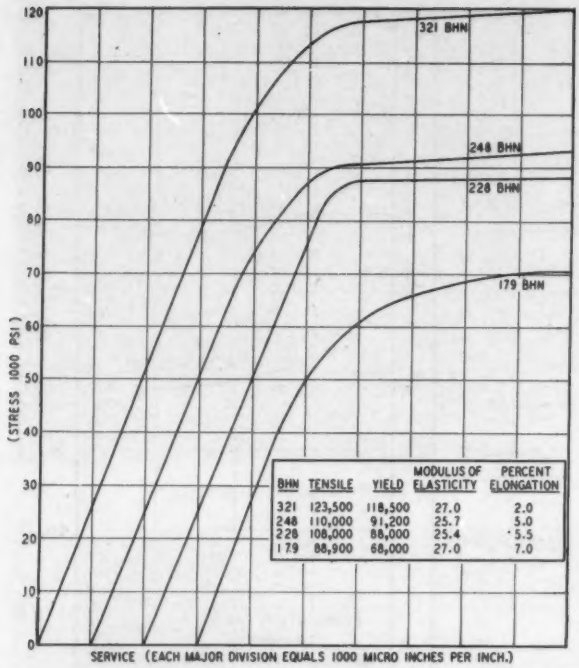


Fig. 8 — Stress-Strain Curves for Oil-Quenched and Drawn Malleable Base Spheroidal Iron at Various Hardness Levels.

lathe operators have not been able to distinguish between pearlitic malleable and malleable base spheroidal iron.

Drilling torque and thrust tests have been conducted at University of Michigan on 1x4x4-in. blocks alloyed to produce a duplex graphite structure. These have been limited in number, but are ample for preliminary purposes. Test data are presented in Table 2 and plotted in Fig. 9. The test conditions are detailed in Table 2 with respect to drill and testing conditions. As in past research, control samples of malleable and pearlitic malleable were run, and the order of testing each of the several materials drilled was varied from drill to drill to secure random test conditions should any variables enter into the test procedure. Drilling torque and thrust are measured by SR4 strain gages mounted to the vise holding the work. Strip chart recorders record both torque and thrust during the actual drilling.

As indicated in Fig. 9, pearlitic malleable and malleable base spheroidal iron require approximately the same horse-power per cubic inch of metal removed at a given hardness. This is very significant, in that in the hardness range of 200 to 240 bhn, malleable base spheroidal iron has a high level of tensile properties, which coupled with a high modulus of elasticity and good machinability, promises considerable value in the engineering field.

Discussion

Malleable base spheroidal iron possesses certain desirable characteristics which appear to offer important advantages in some applications. In addition to its effect on graphite shape, the addition of sulphur in the free or combined form stabilizes cementite during solidification, permitting the production of thick metal sections. The application of solid-state graphitizers accelerates first-stage annealing but does not materially affect carbide stability at the lower

TABLE 2 — COMPARATIVE MACHINABILITY TEST RESULTS FOR MALLEABLE BASE SPHEROIDAL IRON, PEARLITIC MALLEABLE, AND MALLEABLE IRON
(Drill Torque and Thrust Tests)

Material	No.	BHN	Torque, lb-in.	Thrust, lbs	Power, hp/cu in.
Malleable	A	131	35	280	0.423
	B	137	32	280	0.387
Pearlitic	1A	187	47	330	0.568
Malleable	2A	202	48	430	0.581
(Air Cooled)	3A	229	52	397	0.628
Pearlitic	1-O	207	52	357	0.628
Malleable	2-O	255	63	557	0.763
(Oil Quenched)	3-O	269	73	620	0.883
	4-O	269	60	580	0.727
Malleable Base	80A	235	57	380	0.688
Spheroidal	10A	262	60	493	0.727
(Air Cooled)	8WA	269	73	550	0.882
	1WA	293	73	540	0.882
Malleable Base	80-O	235	55	368	0.664
Spheroidal	8W-O	262	58	527	0.702
(Oil Quenched)	10-O	269	63	527	0.763
	1W-O	293	69	515	0.833

Drill Conditions:
13/32 diameter — 120° chisel edge angle
30° helix angle — 10° lip relief angle
118° point angle — 0.067 in. average web thickness
Test Conditions:
Feed = 0.007 ipr
rpm = 630
Cutting velocity = 67 fpm

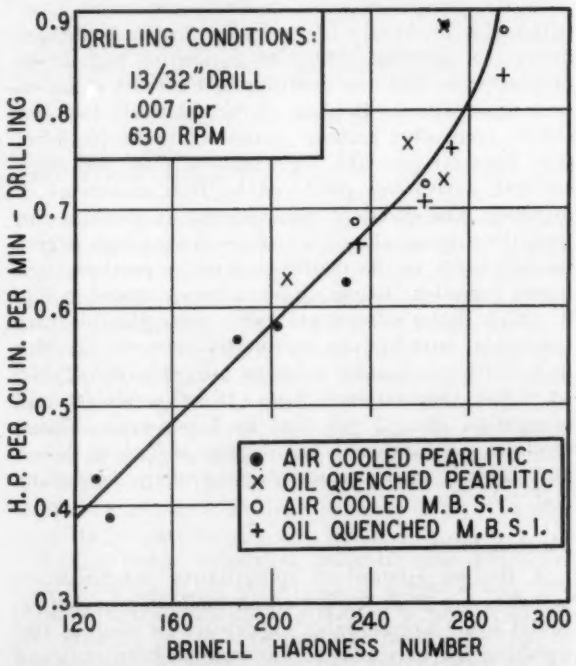


Fig. 9 — Relationship of Drilling Horsepower to Brinell Hardness for Malleable, Pearlitic Malleable, and Malleable Base Spheroidal Irons.

temperatures of second-stage annealing, thereby permitting formation of uniform matrix microstructures with relatively high temperature draw heat treatments.

Alloying for control of the graphite structures may be accomplished in the primary or secondary melting furnaces or by ladle addition, since the control is a true alloying, rather than an inoculating effect. The materials employed are low in cost and readily available. Alloying during melting develops no visible reaction, nor does it require special melting practice beyond the degree of control normally exercised in a duplexing operation. The addition of alloying materials to the ladle is accompanied by a reaction similar in appearance to that observed during inoculation with ferrosilicon. With ladle addition, however, there is a sulphurous odor which requires adequate exhaust to avoid unpleasant atmospheric conditions. No such gaseous evolution occurs in metal alloyed during melting. In experiments conducted to date, there has been no evidence that the alloying effects of sulphur and the necessary graphitizers wear off with time in the ladle or furnace, nor is there a section sensitivity. This is typical of a stable alloying condition.

The fluidity of malleable base iron alloyed with sulphur is considerably greater than that of the unalloyed base. A control spiral used at the Albion Malleable Iron Co. shows an increase from about 14 in. in unalloyed iron to up to 30 in. in sulphur alloyed iron poured under the same conditions. Thin metal sections have been poured with a high degree of success. Excellent casting surfaces were obtained in both green sand and shell molds.

Considerable research has been done to develop plausible theories for the mechanism of spherulitic graphite formation in high carbon irons. Generally, it has been conceded that commercial irons with this type graphite require a low sulphur base with rare earth alloying additions. The present development, however, discloses the plausibility of forming controlled spherulitic or spheroidal graphite structures in a higher sulphur material. As a matter of fact, unbalance in sulphur content, toward levels above those heretofore considered commercial in ordinary gray, nodular, or malleable iron production, is an essential. It has been found that the addition of sulphur, or sulphur combinations, will permit the controlled and predictable formation of a variety of spheroidal graphite structures, ranging from agglomerated rounded temper carbon to spherulitic. Because the iron, as produced for this paper, is an annealed material, a high degree of structural uniformity can be achieved. This has long been an important advantage in ferritic malleable and pearlitic malleable irons, and carries into the malleable base spheroidal irons.

The exact influence of graphite shape within the range which can be produced by sulphur alloying has not been clearly evaluated. At certain hardness levels, the mechanical properties of pearlitic malleable iron and malleable base spheroidal iron are similar. This is most clearly indicated in the range of 241 to 285 bhn. At hardnesses below 241 bhn, malleable base spheroidal iron possesses markedly higher levels of mechanical properties than pearlitic malleable iron. While this high level of mechanical properties is undoubtedy in part related to the shape of the graphite, it is unquestionably also related to the matrix structural uniformity characteristic of malleable base spheroidal iron. Experiments involving long draw treatments at high temperatures demonstrated that the matrix stability was such as to avoid second-stage annealing and excessive carbide spheroidization. Therefore, a uniform matrix with the characteristics of medium to high carbon steel was produced.

The combination of moderate carbon content, structural stability, and structural uniformity produces desirable mechanical properties including a high degree of uniformity and a high modulus of elasticity. Further, because of the matrix uniformity, malleable base spheroidal iron reacts very uniformly to preferential surface hardening, in that the carbides remain in place during draw heat treatments, and it is not necessary in subsequent flame or induction hardening to depend on diffusion of carbon for attainment of uniform high carbon with accompanying high hardness in the quenched surface structure.

Research has indicated that the end-quench hardenability of Albion's base iron and malleable base spheroidal iron is about the same as that of an SAE 4140 steel. This makes the materials amenable to oil hardening throughout a moderately heavy section.

Because of the stabilizing effects of elevated sulphur content, malleable base spheroidal iron has a very high resistance to primary graphite formation during solidification. The base iron has a normal sectional limitation of approximately 1⅛ in. for a completely white structure. Upon addition of sulphur to produce either duplexed or spherulitic graphite, the range of section thickness that can be tolerated is greatly increased. Castings 4 in. thick have been made free of primary graphite in the hard iron state, and uniformly spheroidal as annealed.

Conclusions

The information presented in this paper is limited generally to malleable base iron alloyed to produce a duplexed spheroidal graphite structure. Additional data available show somewhat different levels of properties with different graphite structures, and will be reported in the future. On a basis of the information presented, certain conclusions regarding the new material seem justified:

1. A spheroidal graphite shape with controllable internal structure can be produced by the addition of readily available and low-cost alloying elements. Conventional melting practice for malleable iron, and low-cost melting materials can be employed. Alloying can be accomplished during melting or by ladle addition. No violent thermal reaction is encountered. Alloying effects do not wear off.

2. Malleable base spheroidal iron is stable structurally, and can be treated to high levels of mechanical properties.

3. Heavy casting sections can be produced from the material and subsequently annealed to produce a

uniform spheroidal graphite structure. Castings up to 4 in. thick have been produced successfully.

4. Malleable base spheroidal iron possesses a unique characteristic of high resistance to second-stage annealing, which permits attainment of uniform spheroidized structures through high-temperature heat treatments.

5. The structural uniformity of the material lends itself to ready uniformity in surface hardening treatments, which permit formation of hard surface layers backed up by a strong, tough base material. There is little evidence of a softened subsurface layer.

6. Excellent machinability at high levels of properties can be achieved because of the matrix uniformity which makes it unnecessary, as in other graphitic ferrous materials, to maintain a balance of hard and soft matrix constituents to develop a desired average in hardness.

7. High uniform elastic modulus is obtained in malleable base spheroidal iron through the combination of a rounded graphite in a relatively low carbon graphitic material annealed for uniformity of graphite structure and alloyed and heat treated for uniformity of matrix constituents.

Acknowledgments

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Appreciation is due the Management of the Albion Malleable Iron Co. for financial support, encouragement, and for permission to publish the results of this work.

ADDENDUM

Influence of Sulfur on Graphite Structure

Sulfur is a carbide stabilizer during solidification and in subsequent annealing. Chemically it is a scavenger for manganese; or in the case of other ferrous materials, manganese is a scavenger for sulfur. The two elements combine in the ratio of their atomic weights which is approximately 1.71. By the addition of sulfur either during melting or in a ladle, essentially all manganese is combined in a manganese sulfide complex, and remaining sulfur is present in the form of an iron-sulfide complex. The structure of the iron-sulfide is such that graphite formed during annealing, and apparently nucleated on the iron-sulfide complex, is of a spherulitic type.

The base iron for malleable and pearlitic malleable production at the authors' plant contains an average of 0.44 per cent manganese, which is a balance at two times the sulfur content plus ten to fifteen points. Under ideal conditions, this amount of manganese can be balanced by 0.26 per cent sulfur to form manganese sulfide and result in a metallic matrix free of both manganese and sulfur. In a balanced condition with respect to manganese and sulfur, the graphite structure on annealing is of an intermediate type between typical malleable iron temper carbon and a spheroidal type. When the sulfur content exceeds by about 0.05 per cent that amount required to balance the manganese, the rounded agglomerated type of graphite is produced. At this sulfur level, little if any spherulitic carbon has been observed.

With a further increase of 0.05-0.10 sulfur (.36-.41 sulfur with a .44 manganese base), the duplexed graphite structure occurs. Some carbon is in the form of apparent rounded agglomerates, and a small amount in the form of spherulites. The bulk of the carbon is duplexed. At still higher sulfur content, .45 sulfur and above, a spherulitic graphite structure is produced. Experiments at the authors' plant have included sulfur contents as high as 1 per cent. The resulting castings required excessive annealing times.

Naturally, different base manganese contents would require different amounts of sulfur to achieve the essential chemical balance which permits the rounding of the temper carbon. Other conditions resulting from variations in melting practice might also vary the sulfur contents required for graphite shape control. These matters are under investigation at the present time.

Acceleration of First Stage Anneal

As would be expected, the utilization of sulfur contents in excess of those which can be balanced by manganese retards graphitization both during solidification and during subsequent annealing. The former, retarded graphitization during solidification, is a desirable characteristic of the material, since it permits production of heavy-section castings which at conventional sulfur levels would mottle or become gray. However, retarded annealing rates materially increase the cost of production. To counteract this condition solid-state graphitizers can be, and are employed to bring the annealing time within practical limits. The most widely used of this type of accelerator is boron added to the iron either during melting, or as a ladle addition. Because of the potent effect of the sulfur as a carbide stabilizer, a certain amount of silicon, which will also accelerate graphitization during annealing, can be added without fear of primary graphitzation. Other graphitizers might also be used.

One series of experiments utilizing the normal malleable-base iron produced at the authors' plant involved the addition of both silicon and boron to test ladles of iron which were then poured into standard malleable test bars. The bars were annealed then reheated, oil quenched and drawn at 1200 and 1300 F, and tested to determine the effects of sulfur, silicon, and boron on mechanical properties and response to draw heat treatment. Results are presented graphically in Figs. 10 and 11. These show that increases in sulfur and silicon contents under the heat treating conditions increased the yield strength with little change in hardness or elongation. Increases in boron content by ladle addition of 0.0012 per cent to 0.0048 per cent resulted in decreases in yield strength with little consistent affect on hardness or ductility.

Detailed studies of annealing characteristics of the materials for which mechanical properties are presented are incomplete. However, it is indicated that

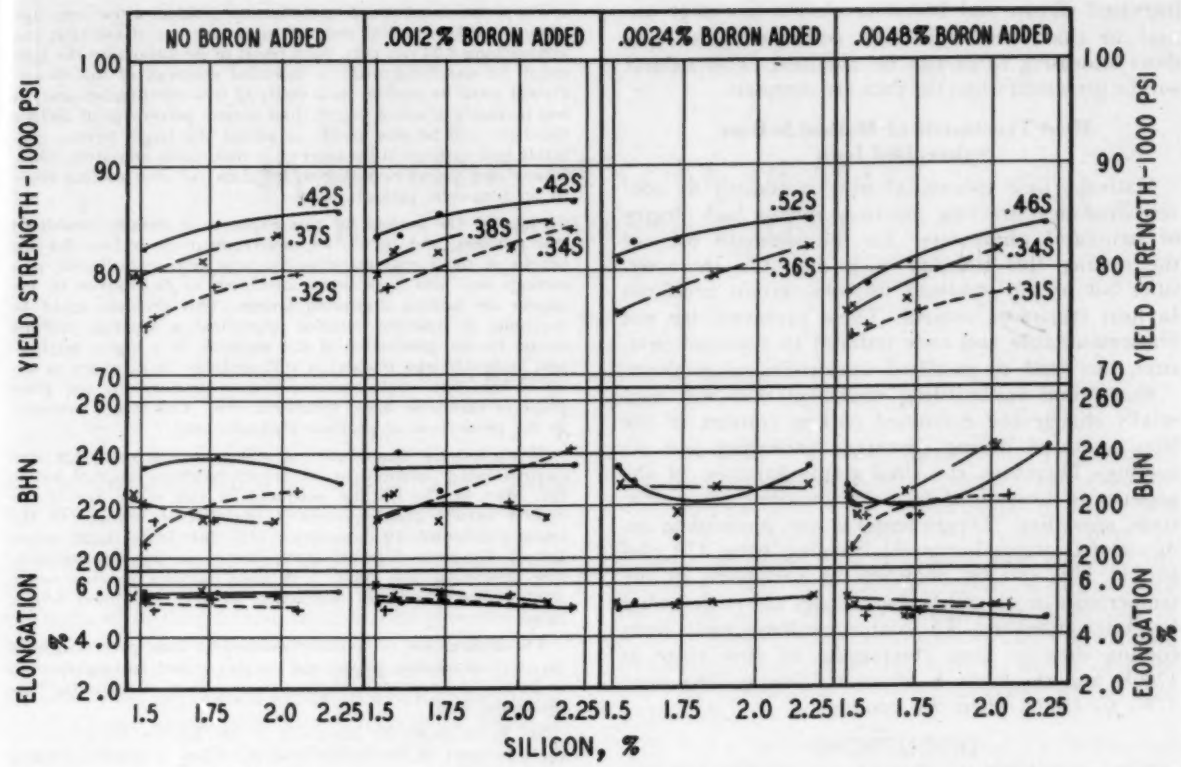


Fig. 10 — Influence of Si, S, and B contents on tensile properties of malleable base spheroidal iron (oil quenched and drawn 6 hr at 1300 F)

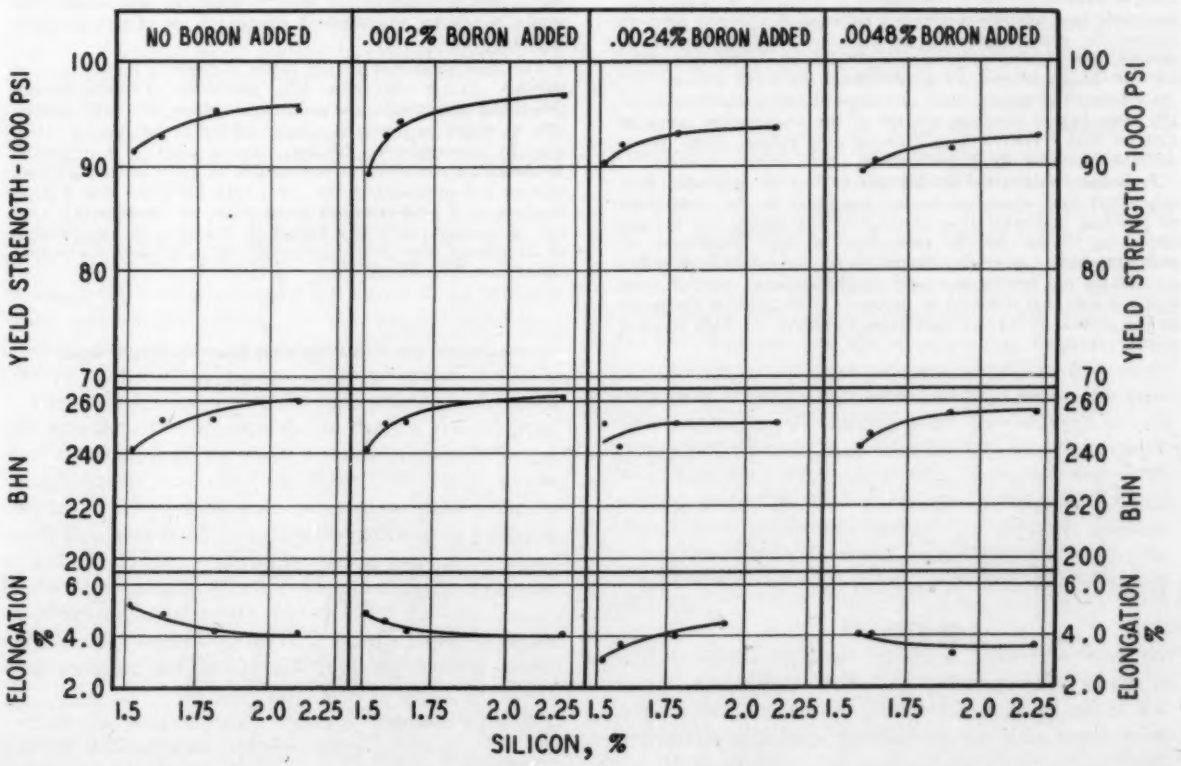


Fig. 11 — Influence of Si and B contents on mechanical properties of a 0.52% S malleable base spheroidal iron (oil quenched and drawn 6 hr at 1200 F).

increased silicon and boron accelerate first-stage anneal in that even with high sulfur contents very short annealing times can be obtained. This subject will be presented when the data are complete.

Heat Treatment of Malleable-Base Spheroidal Iron

Malleable-base spheroidal iron responded to conventional heat treating practices with a high degree of structural uniformity. To the ultimate user of the casting this stability is a desirable characteristic, but to the producer imposes certain problems in heat treatment control. These problems are not unsurmountable and once reduced to standard practices, can lead to excellent uniformity of products.

Short time austenitizing treatment does not materially change the combined carbon content of the iron produced during first-stage annealing and air cooling. Therefore, the final stable hardness of the product is determined by prior handling after first-stage annealing. Experiments as yet incomplete indicate that stable hardnesses ranging from 170 bhn to 241 bhn can be obtained by variations in the temperature from which the castings are cooled after first-stage annealing. The higher hardness results from cooling directly after completion of first stage at 1700F, and the lower hardness by slow cooling from 1700 to 1550F, then air cooling.

DISCUSSION

Chairman: B. C. YEARLY, National Malleable & Steel Castings Co., Cleveland.

Co-Chairman: C. K. DONOHO, American Cast Iron Pipe Co., Birmingham, Ala.

H. R. SAURER[1] *(Written Discussion):* It seems to the writer that malleable base spheroidal iron is a material that should compete successfully with large steel castings and forgings because it possesses the physical properties of pearlitic malleable iron without the danger of primary graphitization in heavy sections.

Uniformity of matrix after prolonged drawing and exceptionally close brinell hardness control in the temperature range of 1200 to 1345 F after a long drawing time further make this a desirable material for large castings.

Problems could arise in the production of malleable base spheroidal iron which we do not encounter in the production of pearlitic malleable iron castings in the cupola air furnace duplexing system. As the percentage of daily production of malleable base spheroidal iron increases, the problem of sulfur control for the ferritic malleable would certainly increase. The writer is sure that it would be necessary to de-sulfurize the metal in the cupola air furnace operations to reduce the high residual sulfur.

The writer faced this de-sulfurization problem a few years ago when his coke was bad and the sulfur content of the iron was approaching 0.20 per cent. As a result of de-sulfurizing the base metal, he was faced with an increased tendency of this de-sulfurized metal to mottle. As a result of this mottling tendency it was necessary to add a higher than normal percentage of carbide stabilizer, and he was unable to anneal the larger ferritic malleable iron castings in his short-cycle continuous annealing ovens. It was even found necessary to lengthen the slow cooling curve in his long-cycle periodic ovens.

Probably the authors of this paper have already considered this problem, and the writer merely mentions it here for the benefit of those who manufacture large ferritic malleable iron castings and who have never attempted to de-sulfurize in the cupola air furnace duplexing system. This problem could be overcome if sufficient business warranted a separate melting system for the production of this material. In a highly mechanized malleable iron foundry it will probably be necessary to set-up at least one whole conveyor line to produce any one given phase of malleable based spheroidal iron. This is not necessary in the production of pearlitic malleable iron.

At the present time there are many pearlitic malleable iron castings being produced in the brinell hardness range of 163 to 207. Most of the castings produced in this range are of the smaller variety and represent a considerable tonnage in the foundry business. In accordance with the Draw Time versus Brinell Hardness Number curve shown in Fig. 4, malleable base spheroidal iron shows a definite hardness stability which would make it costly or impractical to reach this lower brinell range.

The authors are to be complimented on their presentation of this very interesting paper, and we are indeed looking forward to further developments of the various phases of malleable base spheroidal iron.

MR. ROTE *(Reply To Discussion By Mr. Saurer):* We appreciate the comments of Mr. Saurer who has taken a sound operating approach to the problems which might be encountered in the production of malleable base spheroidal iron in conjunction with normal malleable or pearlitic malleable iron. There is no question of the multiple problems to be encountered were high sulfur returns remelted for straight malleable production. This would necessitate separation of returns, or use of separate melting facilities.

The data presented in the paper indicate a high degree of stability during the draw heat treatment of malleable-base spheroidal iron. Under the conditions utilized, it would be difficult to secure hardness in a range of 163 to 207 brinell. However, by adjustment in the heat treating practice, it is possible to reduce the combined carbon content to lower levels than those secured and presented in the paper with the result that a stable hardness at a predetermined lower range for example, 177 - 187, can be secured. This is achieved by lowering the temperature of air quench after first-stage annealing to produce lower combined carbon contents.

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1. Metallurgist, The Dayton Malleable Iron Co., Dayton, Ohio.

EVALUATION OF COKE QUALITY BY A COMPRESSIVE TEST

By

D. E. Krause* and E. A. Lange**

Suitable evaluation of the quality of foundry coke as a means of predicting its behavior in the cupola has been of considerable interest to foundrymen for many years. Foundry coke has a number of measurable properties, many of which are related to its behavior in the cupola. Many properties of coke are interrelated so that a relation of a coke property, not considered directly applicable to the performance in the cupola, may sometimes be found. These relations are sometimes fairly good although a change in coking practice may appreciably alter the relation of such a coke property to cupola performance.

Aside from the effect on the economy of the cupola operation, the foundryman is principally interested in the quality of coke from the standpoint of metal temperature and carbon pickup. Iron temperature at the spout is very important as inadequate temperature will result in a greatly increased number of casting defects. Since carbon pickup in the cupola is also dependent on temperature, temperature is of first importance. Carbon pickup is important as the mechanical properties of the iron are so dependent on carbon content.

As a result of much practical experience, foundrymen have learned the strength of the coke to be a highly important property.[1] Although the manner in which the strength of coke influences metal temperature and carbon pickup by the iron is somewhat obscure, the relation of coke strength to metal temperature and carbon pickup is quite good.

The strength of the coke has much to do with the amount of coke required to melt a ton of iron. As the strength of the coke increases, losses in handling decrease, and crushing of the coke charge in the cupola is reduced. Cupola practice subjects coke to conditions which do not directly concern its function as a fuel. These conditions involve impact or other forms of crushing which coke must resist in order to perform satisfactorily.

Although standard ASTM shatter and tumbler tests measure somewhat different mechanical properties of coke, they both measure the resistance to degradation whether it be by impact and shock or by abrasion. Both tests require a fairly specialized piece of equipment. Very few foundries have attempted to install this equipment and establish the necessary correlation of the results obtained with operating data. There is considerable difference in opinion among coke producers as well as among foundrymen concerning the relative merits of these two tests for evaluating the performance of the coke in the cupola.

A review of the many tests used for evaluating foundry coke was published in 1947 with an extensive bibliography.[2] Comparison of foundry cokes which performed better than average, or considerably below average, consistently showed differences in strength and ignition temperature. Measurement of either property alone did not seem to be adequate. From the standpoint of ease of testing and investment in equipment required, the compressive strength test was investigated as a means of evaluating the strength of coke. The ignition temperature test used is similar to that first described by Boegehold.[3] Although this test has some shortcomings, it has proven to be of sufficient value to warrant its use.

Comparatively little information is available on the compressive strength test for coke. That which is available is not too encouraging. Some work done by two members of the Connecticut Foundrymen's Association on the compressive strength of foundry coke was reported by them to that group in June 1951. Shorthly thereafter, work was started by the Gray Iron Research Institute, Inc. on the subject of compressive strength tests and during the intervening period 86 lots of foundry coke from 18 producers were tested for compressive strength, ignition temperature, volatile matter content, and ash content. The purpose of this paper is to report the results obtained.

Procedure

The results reported by the Connecticut Foundrymen's Association Coke Committee were obtained from 1-inch cubes cut from the lumps of coke. Examination of a large number of test data made avail-

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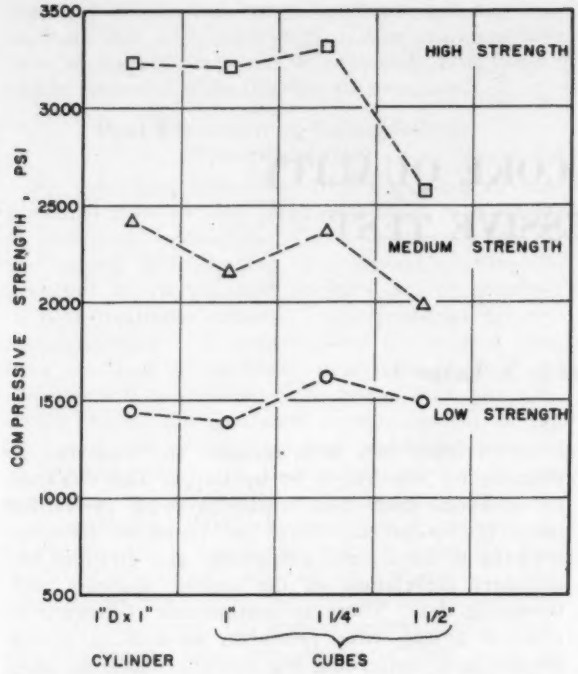


Fig. 1 — Effect of test piece size on compressive strength value of high, medium, and low strength cokes. Data are averages of ten or more test pieces.

able through the courtesy of one of the coke producers indicated that coke cubes obtained from the lumps midway between the wall end of the lump and the seam end gave strength values quite close to the average of values obtained from cubes in all three locations. The wall end of the coke usually showed higher strength than the seam end. In view of these findings and subsequent confirmation, the results reported in this paper were from specimens cut about midway from the wall end to the seam end of the lump.

Some preliminary work in preparing specimens for tests included the use of hole saws, band saws, and hack saws, and demonstrated that cutting of cubes with an abrasive masonry saw using water to keep down the dust was by far the fastest and most convenient method. There are a number of manufacturers of these saws and many foundries have one of these saws available for cutting refractories for use around and in the cupola. It was found desirable in the laboratory to replace the wooden top of the moving tray of this type of saw with an aluminum plate in order to obtain more accurate dimensions.

After some preliminary work, it was found most desirable to carefully select from 6 to 10 lumps of coke showing both wall and seam ends from a shipment of coke. One-inch thick slices were cut from these lumps halfway between the wall end and the seam end and parallel to the wall face of the lump of coke. By the use of proper guides on the table of the saw, 1-in. cubes were cut from the 1-in. thick slices. From 25 to 35 cubes were obtained from a lot of coke in this manner. The wet cubes were dried for approximately 3 hr at 250 F.

The cubes were allowed to cool before testing. If the cubes are cut dry, no subsequent drying would be necessary. The cubes were tested in a conventional screw-powered testing machine with steel pressure plates, one of which had a spherical seat. If a conventional testing machine is not available, a simple hydraulic jack press with a pressure gage can be satisfactorily used. Such presses are relatively inexpensive. In many cases, the cubes after drying were weighed and measured for the purpose of determining their density.

Effect of Test Piece Size

Shortly after the project was started, the influence of test piece size on the strength values obtained was investigated. One-inch diameter cylinders 1 in. high were prepared by means of a hole saw in a drill press and cubes with edge lengths of 1 in., 1¼ in. and 1½ in. were cut with an abrasive masonry saw. It was found difficult to obtain sound cubes much larger than 1½ in. because of fissures always present in coke. Lumps of low, medium and high strength coke were selected and from 10 to 20 pieces of each size of test specimen prepared. The results are shown in Fig. 1.

The lower strength obtained with the 1½ in. cube was felt to be due to fissures which sometimes were not too apparent on the surfaces of the cube. Since it was easier to obtain sound 1-in. cubes than the larger sizes, all of the remaining results reported in this paper were from 1-in. cubes.

Reproducibility of Test

Foundry coke is without much doubt one of the most difficult of foundry materials from which to obtain a representative sample. A certain amount of non-uniformity in a shipment of coke is difficult to avoid because of the nature of the process by which the coke is produced. Even with careful selection of coke lumps, considerable spread in values may be expected. The selection of not less than 10 lumps of coke and the testing of at least 35 cubes should however give a fairly reliable value. Testing of specimens from 10 lumps of coke does not necessarily indicate the reproducibility of the test as there is no way of knowing from what parts of the oven the various lumps came. There could easily be real differences in the strength of the coke in the various lumps although all of them came from the same oven.

One approach to obtaining information on this

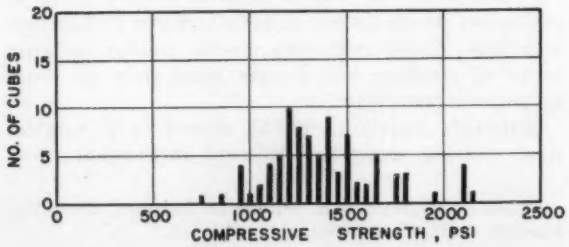


Fig. 2 — Frequency distribution of compressive strength values for 88 one-inch cubes from one lump of coke.

subject was to select a rather large piece of coke (12 in. x 16 in. x 7 in.) and cut as many cubes as possible from a region halfway between the wall end and the seam end of the lump of coke. This large lump of coke yielded 88 one-inch cubes with compressive strengths shown in Fig. 2. This large lump of coke was not considered of particularly good quality as the lump at the seam end appeared appreciably undercoked. Although the 1 in. cubes were obtained from a 2-in. band about halfway between the wall and seam ends of the lump, there was considerable difference in appearance in the cell size and density of the coke which is reflected in the fairly large spread in strength values. A higher quality coke would have shown less scatter in strength values.

Relation of Compressive Strength to Other Properties

Visual inspection of the cut surface of cubes from the various cokes tested revealed an appreciable difference in cell size and uniformity of structure. The stronger cokes usually appeared finer grained and had better defined edges which resisted rubbing off with the fingers to a greater extent than the weaker cokes. The finer grain of the stronger cokes gave the impression of greater density. Dimensions and weights of the cubes were obtained before subjecting them to test and the density calculated as gram/cc. The relation of the average density thus calculated and the average compressive strength for each of the cokes tested is shown in Fig. 3. Although there is some tendency for denser cokes to show a higher compressive strength the relationship is poor. The comparatively low density of beehive coke is quite apparent.

The ash content is thought by some to have an influence on the strength of the coke but the data

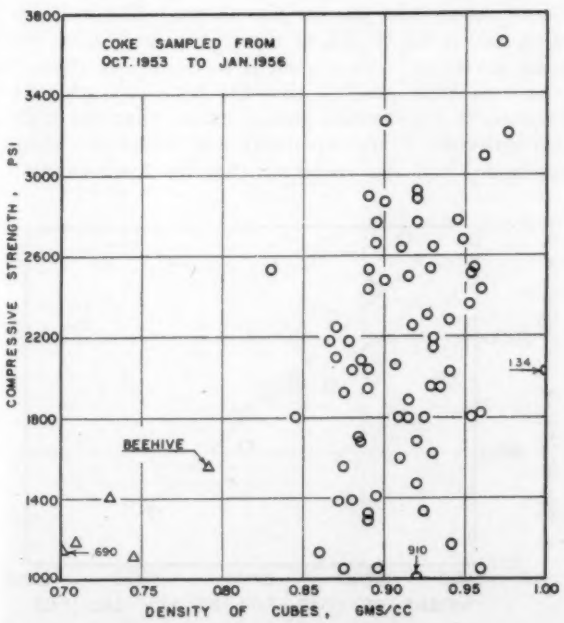


Fig. 3 — Average values of compressive strength and density of 1-in. cubes of cokes.

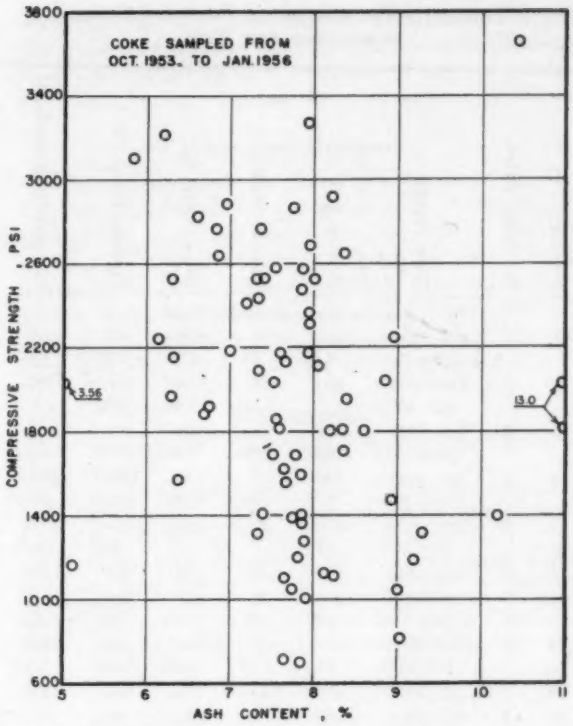


Fig. 4 — Compressive strength and ash content of foundry cokes.

shown in Fig. 4 indicate only a very general mild influence. There is some tendency for higher ash cokes to have somewhat lower compressive strength, but with the large variety of cokes tested, there are factors which appear to have a more important bearing on the strength of the coke than the ash content.

The relation of ignition temperature, as determined by a modified Boegehold test, to the compressive strength is shown in Fig. 5. There is no apparent relationship although cokes having compressive strengths in excess of 2200 psi had ignition temperatures above 940 F.

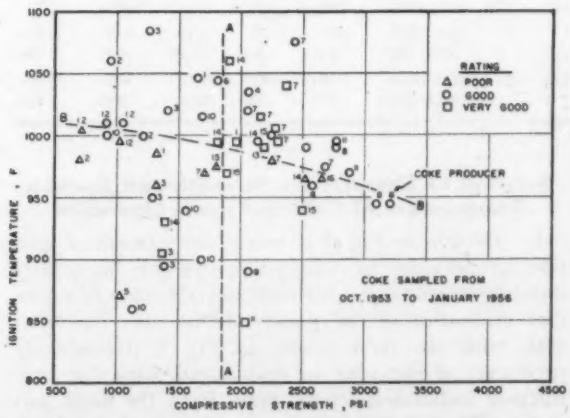


Fig. 5 — Relation between ignition temperature, compressive strength, and cupola performance of foundry cokes.

TABLE 1 — PROPERTIES OF FOUNDRY COKES SUBMITTED FOR TEST

Producer	No. of lots tested	Time Period	Density, g/cc	Volatile Matter, %	Ash Content, %	Ignition Temp., °F	Compressive Strength, psi
Regular Oven Foundry Coke							
1	3	Feb. 1954-June 1955	.920-.928	.35-.54	6.30-9.30	985-1040	1320-1960
2	3	Dec. 1953-Apr. 1954	.897	.57-1.23	7.85-7.89	950-1000	700-1210
3	6	Apr. 1954-June 1955	.871-.895	.36-2.99	7.34-7.91	895-1085	1010-1410
4	2	Feb. 1954-Aug. 1954	.926-.973	.47-.56	7.95-10.44	1025-1035	2310-3670
5	2	Sept. 1953-Mar. 1955	.678-.870	.85	8.05	870-960	1034-2110
6	1	Apr. 1954	.909	.44	7.56	1045	1805
7	9	Jan. 1954-June 1955	.884-.976	.42-.77	6.38-8.95	970-1075	1710-2680
8	21	Nov. 1953-Dec. 1955	.845-.976	.45-1.74	5.85-8.85	890-1050	1690-3270
9	1	July 1953	.875	3.15	7.75	880	1050
10	5	Oct. 1953-Feb. 1955	.860-.930	.70-3.48	7.51-7.67	860-1015	910-1690
11	3	Sept. 1953-May 1955	.847-.920	.85-1.25	6.83-6.95	850-995	2040-2890
12	4	Jan. 1954-June 1955	.960	.48-1.00	7.67-9.04	995-1010	710-1050
13	1	Nov. 1954	.930	1.05	7.00	990	2190
14	8	Nov. 1953-Dec. 1955	.877-.955	.51-1.74	7.40-7.93	930-1060	1390-2530
15	5	Mar. 1954-Nov. 1954	.830-.915	.45-1.13	5.56-8.60	940-1000	1800-2530
16	1	Apr. 1954	.930	1.43	8.37	965	2650
Special Oven Foundry Coke							
1a	1	Feb. 1954	.941	.78	5.13	955	1170
1b	2	Jan. 1955-Feb. 1955	.940-.960	.54-.82	13.00-13.05	990-1010	1820-2030
6a	2	Jan. 1954-May 1955	1.14-1.34	.31	3.56	1065-1075	860-2030
8a	1	Nov. 1954	.945	.91	6.60	950	2820
Beehive Foundry Coke							
17	1	Mar. 1954	.791	.42	6.40	1085	1570
18	2	Apr. 1954-Nov. 1955	.690-.745	.47-.63	8.13-8.19	955-1095	1110-1160
18a	2	May 1954-Nov. 1954	.710-.730	.65-.88	9.20-10.2	990-1040	1190-1410

Relation of Compressive Strength and Ignition Temperature of Coke to Cupola Operation

In addition to Fig. 5 showing the relation of ignition temperature to compressive strength for a large number of cokes, it also indicates the cupola operators' evaluation of the quality of the coke. The numerals with the data points in Fig. 5 indicate the producers of the coke so that points with the same number indicate the coke came from the same producer. These numbers correspond with those shown in Table 1. The cupola operations represent coke ratios from 6:1 to 14:1, hot and cold blast, and appreciable differences in melting rates.

The evaluation in cupola operation is based entirely on the operator's opinion as to whether a particular coke gave average performance, worse than average or better than average. It is entirely possible that an operator accustomed to using high-quality cokes may rate a certain coke as poor, whereas another operator accustomed to poor grades of coke might rate the same coke very good if he had the opportunity to use the same coke. Unfortunately, some of these foundries are widely separated and it was not feasible to interchange some of the cokes of specific interest in this investigation.

In Fig. 5 the vertical line A—A was drawn so that the same number of points are on the left side as there are on the right side. It will be seen that there are eight cokes rated as poor on the left side and three cokes rated as poor on the right side. Three cokes to the left of line A—A are rated as very good, while ten cokes on the right side are rated very good. A more critical examination of the cupola operations disclosed that some coke rated as poor to the right of A—A were used by operators accustomed to high quality cokes.

In considering ignition temperature of the cokes it will be seen from Fig. 5 that none of the cokes above line B—B were rated as poor. Although the data show a wide scatter, they do seem to show that a lower ignition temperature can be tolerated if the strength is high. There is a fair relationship between ignition temperature and volatile matter content, the ignition temperature decreasing as the volatile matter increases.

During the course of certain experimental work with a cupola lined to 54-in. diameter, a relation between the compressive strength of the coke and the weight of the coke required to make up the bed to 60 in. above the tuyeres was found. This relation is shown in Fig. 6. All of these cokes were from the same producer. The cupola is mechanically charged and it is quite possible that the lower-strength coke crushed to a somewhat greater extent than the high-strength-coke. From experience with cokes of various strengths, it is also suspected that the lower-strength

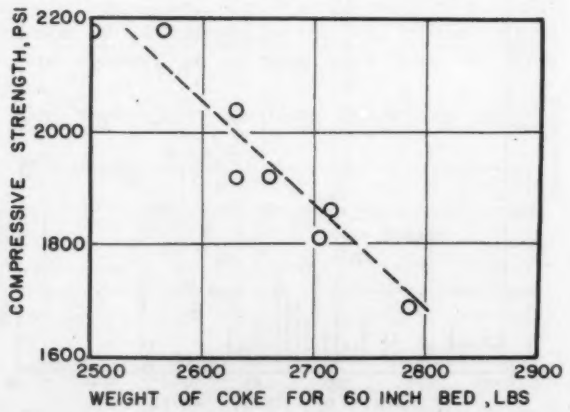


Fig. 6 — Effect of compressive strength on weight of bed coke.

coke had the corners rounded off more than the higher-strength coke which would allow a greater density in packing of the coke when it is dropped into the cupola. Unfortunately, this relation was not found until some time after the work was completed and the coke samples used up.

Properties of Cokes Tested

As a matter of general interest, some of the properties of the cokes tested, the number of lots of coke from a given producer tested and the dates covered by the tests are shown in Table 1.

Summary

Although the compressive strength test for evaluating foundry coke has some shortcomings, it should be considered as a useful test. Since the amount of coke actually tested is relatively small as compared with a shatter test, great care must be exercised in regard to obtaining a representative sample. Although some sort of strength test has thus far proved to be the most useful of the many coke tests devised, the quality of foundry coke should not be judged solely on its strength. A combination of some type of ignition, reactivity, or combustibility test with a strength test will probably be required to define the quality of foundry coke completely. The problem of proper sampling of a shipment of coke is a large one, and an easy solution does not appear probable.

Acknowledgment

The authors wish to thank the members of the Cupola Committee of the Gray Iron Research Institute for supplying the cokes for test as well as the cupola operators' evaluation of the cokes. They also wish to thank Messrs. Thibeau, Hasselman, and LeRoy of the Connecticut Coke Co. for their helpful suggestions regarding the method of sampling the coke and preparation of the test specimens.

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STRESS-STRAIN RELATIONSHIP FOR GRAY IRON

By

E. M. Stein*

ABSTRACT

Typical stress-strain relationships for medium- and low-strength gray irons are presented. The stress-strain curves are bowed from the origin, showing that these irons do not have a true modulus of elasticity. However, a definite relationship was found for all of the irons investigated upon plotting the percentage of ultimate strength against strain. This relationship suggests that stress analysis of gray iron should be interpreted in terms of the percentage of the ultimate strength rather than in terms of stress.

Introduction

Most commonly used structural metals have a well-defined modulus of elasticity. A significant part of the stress-strain curve for these metals is linear. Obtaining an accurate value of the elastic modulus for such metals is, therefore, quite simple. Design of structural elements such as beams, columns, and rings can be accomplished without too much difficulty by means of elementary stress analysis as long as the loads utilized are within the elastic region. Gray iron, however, is different than most metals, in that the stress-strain curve has no significant linear portion. Since there is curvature even at very low values of stress, there is no true modulus of elasticity.

One method of coping with the difficulty of evaluating materials such as gray iron is to utilize what is sometimes called a "relative" or a "reduced" modulus. Substitution of a relative modulus for an elastic modulus is, of course, an improvement since it recognizes the disparity between the actual behavior and an elastic behavior. A stiffness comparison between one iron and another can be obtained from the relative moduli. Two of the more familiar relative moduli are the tangent modulus and the secant modulus. The tangent modulus at a given stress is the slope of a tangent to the stress-strain curve at that stress. The secant modulus at a given stress is the slope of a line connecting the origin and the curve at that stress.

Experimental Procedure

In a study that was made recently at Battelle of gray iron properties, strain data were obtained for over 200 tensile specimens. Two electric-resistance strain gages* were mounted opposite each other and were connected in series so as to average the bending effects. Two temperature-compensating gages were used, also connected in series. Strains were measured with an SR-4 strain indicator.

The bars, which were 0.800 inch in diameter, were loaded at a platen speed of 0.005 inch per minute, and unloaded to 1000 psi at the same speed, from progressively higher loads. This low loading rate permitted accurate strain readings to be obtained to the point of

*Electric-resistance strain gages were chosen because optical methods lack sufficient precision and because mechanical extensometers must be removed before fracture to prevent their being damaged.

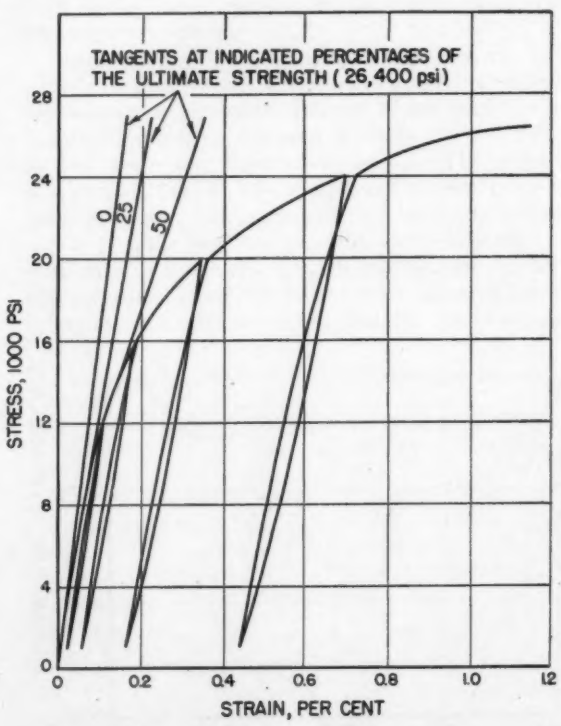


Fig. 1—Loading and unloading curve for a medium-strength gray iron.

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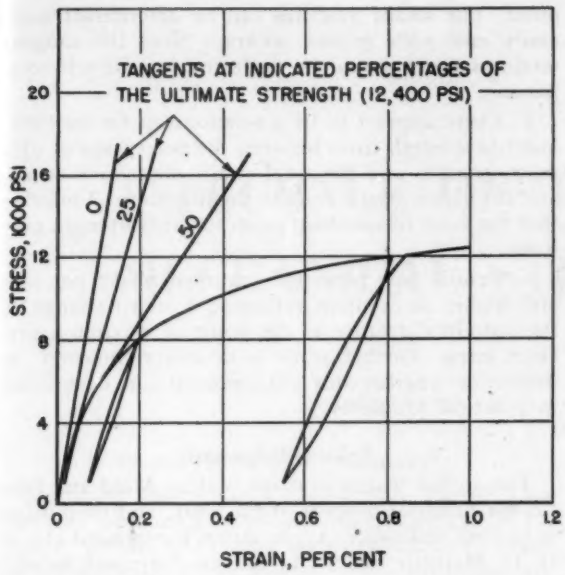


Fig. 2—Loading and unloading curve for a low-strength gray iron.

fracture. The purpose of maintaining a slight minimum load was to prevent a change of alignment upon reloading.

Results

Figures 1 and 2 show typical stress-strain relationships obtained for medium-strength and low-strength irons. The tensile strengths of the two bars for which stress-strain curves are shown were 26,400 psi and 12,400 psi, respectively.

The total strain at any stress is composed of elastic (recoverable) strain and plastic strain (permanent set). For example, in Fig. 1 the total strain is 0.70 per cent at a stress of 24,000 psi. The projected unloading curve from this stress intersects the zero stress line at 0.43 per cent strain. The total strain for a stress of 24,000 psi thus is composed of 0.43 per cent plastic strain and 0.27 per cent elastic strain. Characteristic hysteresis loops were observed for all specimens. These loops, for any one iron, became more pronounced with increasing stress.

Tangent moduli at 0, 25, and 50 per cent of the ultimate strength are shown in Figs. 1 and 2, and values of these moduli are presented in Table 1. The rapid change of tangent moduli values with increasing stress is the reason that constant moduli values should not be used in evaluating the properties of gray iron. At any given stress on a nonlinear curve, the secant modulus is always higher than the tangent modulus.

TABLE 1 — TANGENT MODULI FOR MEDIUM- AND LOW-STRENGTH GRAY IRONS AT VARIOUS PERCENTAGES OF ULTIMATE STRENGTHS

Per Cent of Ultimate Strength	Tangent Moduli, 10^6 psi	
	Tensile Strength, 26,400 psi	Tensile Strength, 12,400 psi
0	16.5	12.0
25	11.6	7.2
50	6.0	3.0

It should be emphasized that since these moduli vary with stress, it is essential that, for any relative modulus value, the corresponding stress be stated.

Figure 3 shows the relationship between the percentage of ultimate tensile strength, and the strain, for the two bars under discussion. These curves indicate that the stress at a given total strain represents approximately the same percentage of the ultimate strength, irrespective of the strength. The elastic and plastic strain curves appear also to be independent of the ultimate strength. It is interesting to note that calculations based on data obtained by other investigators[1] indicated that curves for high-strength irons follow closely those observed for medium- and low-strength irons. Thus, evaluation of all gray irons usually can be interpreted more accurately in terms of percentage of available (ultimate) strength, rather than in terms of stress.

The implications of this property are interesting. It says that for gray iron the stress at a given strain divided by the ultimate strength is constant. Mathematically, a stress-strain curve may be written as

$$S = A\ F(e),$$

where S is the stress that is a function, F, of the strain, e, and A is a constant. For different stress-strain curves, F is usually different. A possible example would be

$$S = A\ e^n,$$

where n is a constant for a given material.

The interesting feature about gray iron is that both low- and medium-strength types can be fairly well represented by an equation of the type

$$S = Su\ F(e),$$

where the constant $A = Su$, and the same function F, applies for both low- and medium-strength irons.

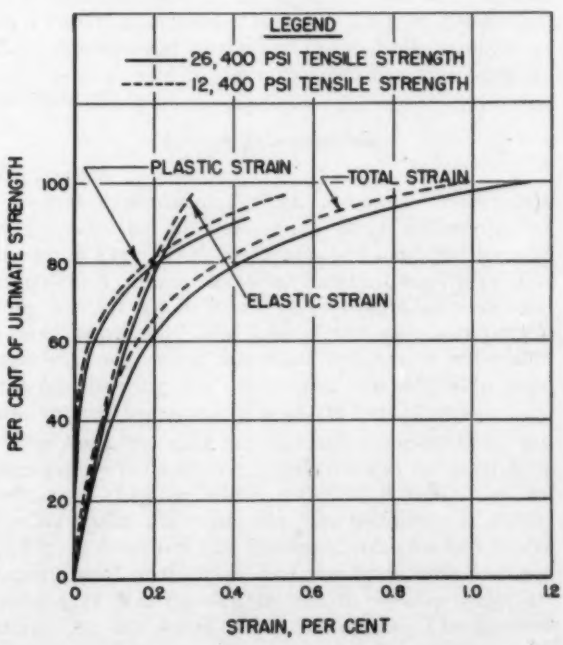


Fig. 3—Relationship between per cent of ultimate strength and strain.

This property may be quite useful. One example can illustrate that this is true. A solution for bending of a beam with a curved stress-strain curve is available[2]. In the derived expression for bending moment, it is merely necessary to change the value of S_u in going from the solution for, say, a low-strength to a medium-strength iron. This allows a significant simplification, particularly when numerical integrations are necessary.

One aspect of the stress-strain curves is especially intriguing and may possibly be used to practical advantage. It is observed that upon unloading from strains of approximately 0.2 per cent and higher, then reloading, there is a sharp change in curvature at the point of maximum previous stress; this behavior is common to most materials that do not experience rapid recovery. Using this property, it may be possible to determine the maximum load which a gray iron part has reached in service, if it has at any time been strained to 0.2 per cent or higher. To determine the maximum stress attained in such a part, a stress-strain curve would be obtained on a bar machined from the part. An alternative method would be to conduct the test on the part itself, in which case the maximum load reached in service would be obtained rather than stress.

Conclusions

The following conclusions are indicated by the results obtained:

1. Stress-strain curves for gray irons are curved from the origin. Thus, these irons have no moduli of elasticity in the sense that steel (for example) has.

2. Relative moduli (tangent and secant) are useful only for comparing the stiffness of one iron with another. The secant modulus can be determined more easily and with greater accuracy than the tangent modulus, so it should be preferred for stiffness comparison.

3. There appears to be a relationship for medium- and low-strength irons between the percentage of ultimate strength and the total strain, the elastic strain, and the plastic strain. Data in the literature[1,3] indicate that the same relationship exists for high-strength gray irons.

4. Tensile bars previously strained to 0.2 per cent and higher show upon reloading a sharp change in stress-strain curvature at the point of maximum previous stress. Further work is necessary, however, to determine whether this phenomenon can be utilized in practical problems.

Acknowledgment

The author wishes to thank Vulcan Mold and Iron Co. for financial support of the work, and permission to publish this paper. Appreciation is expressed also to H. O. McIntire and R. L. Carlson for many helpful suggestions, and to R. E. Bradford and E. W. Tucker for their assistance in obtaining the data.

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INFLUENCE OF TEMPERATURE ON MECHANICAL STRENGTH OF COKE

By

John Varga, Jr.*, and H. W. Lownie, Jr.*

Introduction and Summary

Results of new research have shown that mechanical tests conducted at room temperature cannot be used to predict the extent to which coke will break up mechanically at the service temperatures encountered in cupolas. The rate at which coke breaks up, and the total amount of breakage which occurs, are different at room temperature than at the higher temperatures encountered in service.

Work done in the past on evaluation of coke properties and published reviews of this subject by various individuals have all stressed that the two most important properties of coke for cupola use are (1) its size, and (2) its strength. For example, according to Mulcahy,[1] "The physical properties of the coke, however, are the ones we are most concerned with, for these factors in most instances exert by far the greatest effect upon the performance of the coke. The principal factors in this category are as follows: size, strength, weight per cubic foot, porosity and cell size, and, of these, if we consider size and strength alone, we will have an index of coke quality more important in general than all of the other factors combined."

The same points are emphasized by Krause,[2] Krause and Lownie,[3] and Carter and Carlson.[4] In 1951, a survey conducted by Lownie[5] contained the following statement, "Many of the foundryman's operating difficulties could be reduced if he were supplied with a larger and stronger coke. At least, higher strength and less undersize are two things that most foundrymen want." Investigators in Europe and England[6,7,8] also state that high strength is a major requirement of a good foundry coke.

Evaluation of coke strength has been done in the past primarily by the use of shatter tests and tumbler tests *made at room temperature*. Results obtained from the ASTM shatter tests have shown very good correlation with the breakdown a coke receives in shipment and handling *before* it is charged into the cupola. Correlation of tumbler-test results with cupola operations have not usually been so good as for the shatter test. A thorough coverage of the background and development of shatter and tumbler tests is given in the work of Mott and Wheeler.[9]

It was recognized a few years ago that the lack of correlation between coke tests and the performance of coke in service might be caused by the fact that the tests are made at room temperature, while the production operation is carried out at least partially at high temperature. High-temperature tests are used to advantage in evaluating metals to provide information on high-temperature creep and resistance to oxidation. Why not, then, use tests at elevated temperature to evaluate coke? Why not test the coke at the same temperatures it encounters in service, or at least determine whether the strength of the coke is the same at high temperatures as it is at room temperature?

The only work found published in the literature on the high-temperature strength of coke is that of Isarov,[10] who was concerned with individual pieces of coke rather than with aggregate coke as it is used in a cupola operation. The work done in this present investigation was concerned with bulk samples of coke as it is used in the cupola, rather than with individual pieces.

Initial Investigation

Equipment

Initial experimental work at room temperature was conducted in the simple piece of equipment shown in Fig. 1. The apparatus is a modified form of tumbler. A section of 10-in. standard steel pipe 4 ft long was set up to rotate end-over-end on centrally-placed trunnions. One end of the pipe was permanently closed, while the other end had a removable cap for charging and discharging the coke. The pipe was rotated by means of a variable-speed drive.

For tests on coke at elevated temperatures, the steel-pipe section of the apparatus was enclosed in a refractory chamber which could be heated. This assembly (with the refractory top removed) is shown in Fig. 2. A control thermocouple was attached to the outside wall of the pipe and the leads were brought outside of the furnace by means of slip rings attached to the water-cooled trunnions. The control

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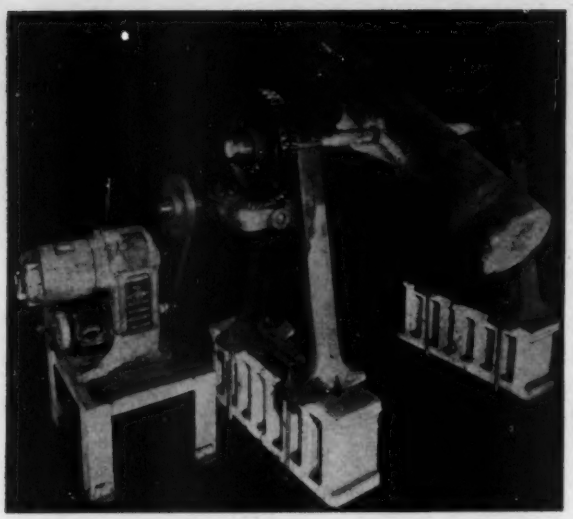




Fig. 1 (Left) — Original equipment for testing coke at room temperature. Fig. 2 (Right) — Original equipment for testing coke at elevated temperature.

thermocouple was calibrated against a reference thermocouple placed in a sample of coke inside the pipe during heating and holding periods.

Experimental Work

Preparation of Samples of Coke for Testing. The ASTM-recommended method for preparing a sample of coke for tumbler testing was found to have some disadvantages for this work, so an alternative method was devised. The ASTM method specifies that all of the coke in the test sample be of one uniform size, regardless of the size range of pieces of coke in the shipment. To prepare samples for the current work, the distribution of particle sizes in the sample was identical with the distribution in the gross lot of coke being studied. The ratio of the amount of coke retained on each test screen was the same in the 20-lb test samples as in the large lot of coke under investigation.

Operating Procedure. The following procedure was used for each test of the initial investigation. The pipe was turned to a horizontal position, the test sample charged and distributed as uniformly as possible over the length of the pipe, and the pipe closed. The refractory cover was then placed on the heating chamber, and the chamber heated to the desired testing temperature.

When the testing temperature was reached, the charge was held for a soaking period of 20 minutes. The pipe was then rotated at 10 rpm for the desired number of revolutions. Rotation was then stopped, and the heating chamber cooled. The coke sample was discharged, screened and then often recharged to the pipe for further degradation. In this program, screen analyses of the coke was obtained at zero, 100, 300, and 600 revolutions of the pipe.

For all of the tests described in this paper, the coke was screened through a series of sieves, the coarsest of which had openings 3 in. square, and the finest of which had openings $\frac{1}{8}$-in. square. For use in this paper, reference will be made usually to only the cumulative percentage of coke retained on a 1-in. screen after testing. This percentage corresponds to the ASTM "Stability Factor" (ASTM Standard Tumbler Test for Coke, Designation D294-50).

Results of Initial Investigation

The results of the initial program showed that for each of three cokes investigated, the amount of degradation of the coke in this test increased as the service temperature was increased. A typical result is

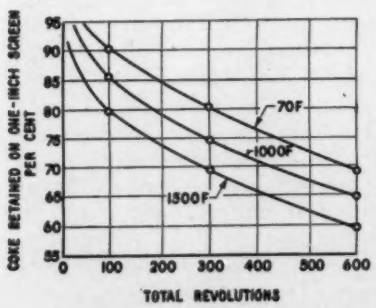


Fig. 3 — Effect of testing temperature on strength of Coke A.

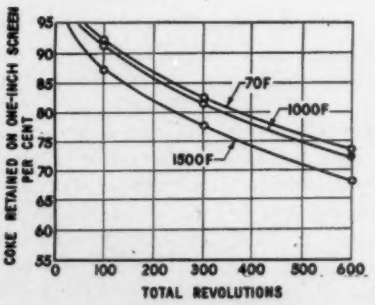


Fig. 4 — Effect of testing temperature on strength of Coke B.

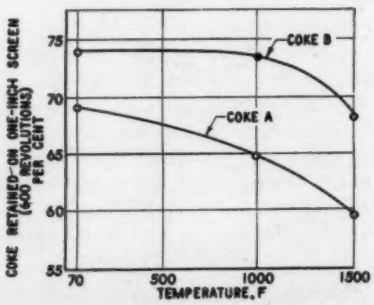


Fig. 5 — Effect of testing temperature on strength of Cokes A and B.

shown in Fig. 3 for Coke A. Notice that the difference in the degree of degradation of this coke at the three temperatures of testing was almost constant. The situation was slightly different for Coke B, as shown in Fig. 4. Coke B showed little difference in degradation between room temperature and 1000 F. When the testing temperature was increased to 1500 F, however, this coke degraded much faster than at 1000 F.

The effect of testing temperature on the degradation of these two cokes after 600 revolutions is shown in Fig. 5. The difference in strength of these two cokes, and the difference in their resistance to degradation at high temperatures, are readily apparent. Also, it can be seen that Coke B at 1500 F was as strong as Coke A at room temperature.

Once it was established that the mechanical strength of coke did change with testing temperatures, it was decided to extend the investigation to other cokes and to higher testing temperatures.

Extended Investigation

Equipment

The steel-pipe apparatus used for the initial investigation was satisfactory up to a testing temperature of 1500 F. At temperatures above 1500 F, the pipe scaled rapidly and the water-cooled bearings distorted. Also, the refractory chamber with its removable top was unwieldy and difficult to handle when hot. Furthermore, the design of the assembly was such that the large amount of heated refractory made working conditions uncomfortable during charging and discharging of a sample. All of these disadvantages of the first equipment were overcome by building a new unit using a refractory tube surrounded by a combustion chamber which rotated with the tube.

For the second design of a testing apparatus, a tube of silicon carbide was selected as the testing chamber because the tube had the hot strength and high thermal conductivity required for the high-temperature tests. The tube measured 10 in. ID, 13 in. OD, and 54 in. long. Three inches of the length of the tube were used for support at each end, leaving a length of 48 in. to be heated to the testing temperature.

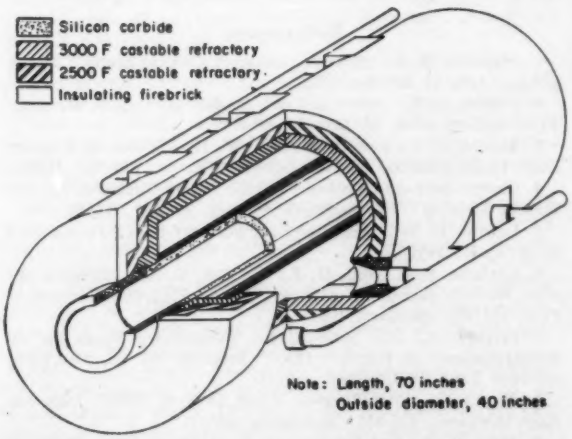


Fig. 6 (Above) — Cutaway view of high-temperature coke testing equipment. Fig. 7 (Right) — High-temperature coke-testing equipment.

A cutaway view of the testing unit is shown in Fig. 6. The outer steel shell was lined with three refractories as shown. The combustion chamber was heated with natural gas using two manifolds of three burners each. The burners fired tangentially into the combustion chamber. The unit ready for operation is shown in Fig. 7.

Temperature control of the unit was maintained with two thermocouples in the combustion chamber. One thermocouple was located 12 in. from each end of the combustion chamber, and the thermocouples were diametrically opposed. The leads were brought out of the furnace through slip rings and were attached to a recorder. The control thermocouples were calibrated against a reference thermocouple located inside the silicon carbide tube.

Experimental Work

Preparation of Samples of Coke for Testing. The samples of coke for testing were prepared in the same manner as described for the initial investigation.

Operating Procedure. The unit was heated to the desired testing temperature and the prepared sample of coke was charged into the silicon carbide tube. After standardized holding period to bring the coke up to temperature, the unit was rotated for the desired number of revolutions. The hot coke was then discharged into a steel container. The container was immediately closed with a tight-fitting lid and cooled in running water. The cooled coke was screened, and then often recharged to the tube for further degradation. As in the initial investigation, screen analyses were usually made after 100, 300, and 600 revolutions of the tumbler.

Results of Extended Investigation

A considerable number of cokes of different types were tested at room temperature, 1500 F, 2000 F, and

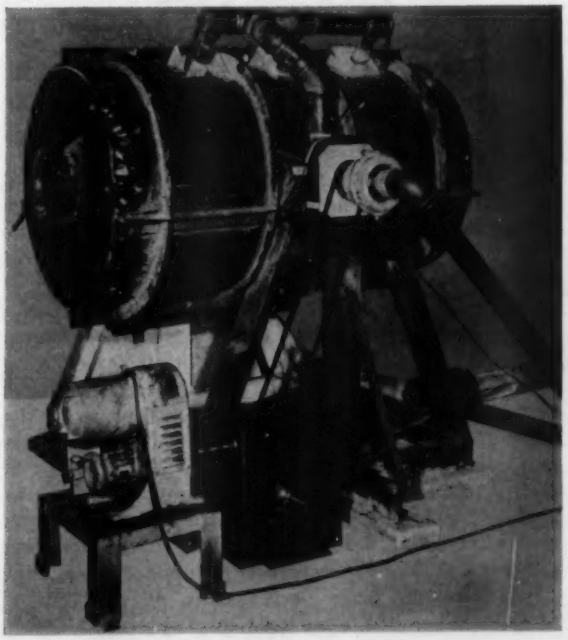
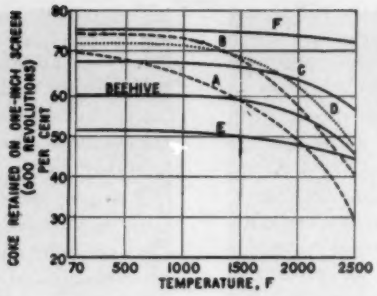
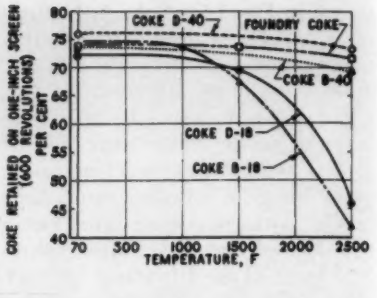


Fig. 8 (Left) — Effect of testing temperature on strength of seven cokes.

Fig. 9 (Right) — Effect of testing temperature and coking time on strength of several cokes.





2500 F. The results showed that the adverse effect of higher temperatures on the strength of the coke continued for testing temperatures up to 2500 F, but that some cokes withstood the higher temperatures much better than others. The effect of testing temperature on the rate of degradation of several cokes is shown in Fig. 8.

Four of the cokes shown in Fig. 8 had room-temperature strengths that were almost the same. Each of these four had about 70 to 75 per cent of the sample retained on a 1-in. screen after testing at room temperature. When tested at 2500 F, however, these four cokes showed a wide difference in strength. For example, based on the amount of coke retained on a 1-in. screen after testing at 2500 F, Coke F showed little increased degradation at the higher temperature, and Coke A degraded so much that only about 28 per cent of the sample was retained on a 1-in. screen. Obviously the relative mechanical strengths of these four cokes differed more from each other at elevated temperatures than at room temperature. Cokes A, B, and D appear from a room-temperature test to be almost as strong as Coke F. At 2500 F, however, Coke F had a marked advantage over the other three.

The reverse situation is also illustrated in Fig. 8. Cokes D, E, and beehive had room-temperature strengths that varied over a wide range of about 20 per cent retained on a 1-in. screen. The spread at 2500 F between these three cokes was only about 3 per cent. Coke E was by far the weakest coke of the entire group of seven when tested at room temperature, but when tested at 2500 F, Coke E made a respectable showing, better than Coke B which was one of the strongest at room temperature.

These results and many others show quite clearly that tests at room temperature on coke cannot be used to predict the strength of coke at the higher temperatures encountered in cupola service.

The effects of coking time and testing temperature on the rate and amount of degradation of two cokes are shown in Fig. 9. The numbers following the coke designations B and D indicate the coking time in hours. A foundry coke with a 28-hr coking time is shown for comparison. Coke B, coked for 18 hr, had a room-temperature strength of 74 per cent (retained on a 1-in. screen), but its strength decreased drastically to 42 per cent at a testing temperature of 2500 F. Increasing the coking time to 40 hr, using the same coal mix, produced no change in the room-temperature strength, but improved the strength at 2500 F to 69 per cent. Coke D, when coked for 18 hr also had a high room-temperature strength, 72 per cent retained on a 1-in. screen. The effect of testing temperature was again quite pronounced, and the strength of this coke decreased to 46 per cent at 2500 F. When the same coal mix was coked for 40 hr, the room-temperature strength increased slightly to 76 per cent. More importantly, however, the coke now stood up well at high temperatures so that 74 per cent of the coke was retained on a 1-in. screen after testing at 2500 F.

Those results, which show the effect of coking time on the mechanical strength of oven coke, illustrate one important use for the high-temperature test method.

Conclusion

The purpose of this research to date has been to develop a method for testing the mechanical strength of coke at the temperatures likely to be encountered in service. No attempt has yet been made to correlate the results of these tests with the performance of the coke in a cupola. Obviously, however, this is a step which should be taken.

From the work that has been completed, it is possible to draw the following major conclusion:

> The mechanical strength of coke at room temperature gives no indication of its mechanical strength at the higher temperatures encountered in cupola operation.

The method of testing described here is adaptable to a wide variety of materials. For example, anthracite, char, and special types of coked products have already been tested to yield information that has been useful, and often surprising.

The authors are indebted to the United States Steel Corp. for permission to publish this paper, since much of this discussion is based on work done at Battelle Memorial Institute under their sponsorship.

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DISCUSSION

Chairman: W. W. LEVI, Lynchburg Foundry Co., Lynchburg, Va.

Co-Chairman: E. J. BURKE, Hanna Furnace Corp., Buffalo, N.Y.

Secretary: C. F. WALTON, Gray Iron Founders' Society, Cleveland.

B. P. MULCAHY[1] *(Written Discussion):* This attempt by the authors to devise a hot strength test for coke is to be highly commended, in fact, any new test procedure which will reveal more refined information or some new indices is very desirable.

In attempting to evaluate coke, i.e., to determine in advance of use how a coke will perform for some particular process, it must be remembered that almost all principal tests now employed are destructive; by this we mean that the test is designed along lines hoping to simulate certain isolated conditions encountered in actual use. The test described by this paper is of this character and as such must be examined to see if it does reproduce actual use conditions.

It is very unfortunate that the authors did not include in their data the complete test results of the cokes tested at "room temperature", because I am sure they would have found some very significant correlation between some of the factors regularly determined and their hot strength data.

Early in the paper the authors state "Results obtained from the ASTM shatter tests have shown very good correlation with the breakdown a coke receives in shipment and handling before it is charged into the cupola. Correlation of tumbler test results with cupola operation have not usually been so good as for the shatter test."

The authors then procede to use this very factor of tumbler test to establish and delineate the cokes under hot conditions.

If we critically examine the test we must reach the conclusion that it is predominantly one of shatter rather than "a revised tumbler test", because the coke is rotated end-over-end in a tube length of 48 in. representing its potential drop.

Even the conventional Tumble Test is recognized to be a combination of shatter and attrition, with the shatter effect predominating. This was clearly shown by the high correlation found between the shatter test and the tumbler test in the work of the Coke Evaluation Project No. 43, by the American Iron and Steel Institute, and American Coke and Coal Chemicals Institute. In the present instance it is felt that the apparatus engenders shatter to an even greater extent.

Our conventional tests do distinguish cokes to a better extent than indicated by the authors. For example, it is well known that the Volatile Matter index is highly significant when properly interpreted. As V.M. increases (within normal operating limits) the shatter test strength goes up, which is readily understood since fissuring has not reached the advanced state. At the same time, however, the tumbler index shows a decided weakening. We also know that such a coke in a cupola "decrepitates" due to re-carbonization of the individual piece. Many samples of coke from the "drop" have been re-analyzed to establish this point.

Many years ago we conducted a rather extensive series of tests on several selected types of coke ranging from a V.M. content of 0.50 per cent to 9.0 per cent. These cokes were immersed in molten iron from the cupola with temperatures ranging from 2400 F to 2850 F and held in insulated ladles. In every instance we found this same disintegration of the coke containing the high V.M. content.

The testing of coke to determine fundamental properties is very difficult for the simple reason that the tests are destructive and therefore impossible of reproduction on the same sample.

These remarks are not derogatory toward this work reported, and it is sincerely hoped that the work will continue, but with a full appreciation of the conditions of the test and, of course, a true interpretation of the results.

At the present time we have an active sub-committee of ASTM, D-5, XXII engaged in a complete review and study of all Physical Tests of Coke. The tests under consideration are not only those currently listed by ASTM, but also all other tests of which we are aware. These latter include those from this country as well as all European practice.

1. President, Fuel Research Laboratory, Indianapolis, Ind.

We fully appreciate the difficulties encountered by the authors in conducting the tests they describe, but suggest that they study the results in direct comparison with the conventional tests as a matter of rigorous appraisal, because it must be remembered that our industry has accumulated vast information on coke properties and any identification must start from this point. We frankly admit that at the present time we do not completely identify a coke but at the same time it must be recognized that many of the factors measured are fundamental characteristics.

MESSRS. VARGA and LOWNIE *(Authors' Reply To Written Discussion By Mr. Mulcahy):* Mr. Mulcahy's first point is that it is a disadvantage of the test which the authors have described that the test is destructive. We agree that it is destructive, and wish that it were otherwise. Many of our recognized tests on materials are, however, destructive, but are still popular and widely used. The conventional tensile test on steel or iron is just one example of another test which is destructive, but is widely used because it is the best we have for the purpose.

As a second point, Mr. Mulcahy says that the test "must be examined to see if it does reproduce actual use conditions". We have given a lot of thought to this for several years, and are convinced that the test probably *does not* reproduce actual use conditions. We do feel, however, that at least on the matter of temperature, the test approaches some conditions of use more closely than other tests which have been proposed for coke. Continuing the analogy with the tensile test for steel, the conventional tensile test is widely used in spite of the fact that few *service* applications for metals involve the particular type of loading used in the *test*.

Mr. Mulcahy states that if we had correlated the results of our tests with "other factors regularly determined" (such as Volatile Matter, for example), we probably would have found a correlation between the current conventional tests and the new tests. The authors wish to state that the absence of data in the paper on such correlations should *not* be taken to mean that such correlations were not attempted. The absence of the data, however, *does* mean that we were not able to establish such correlations in our work. We preferred to delete from the paper those results which led to no useful conclusion.

In his discussion, Mr. Mulcahy has made the erroneous assumption that our test was basically a shatter or tumbler test involving impact of the coke in falling from one end of the tube to the other. We agree with most of Mr. Mulcahy's comments on shatter and impact, but they do not apply to the work that is reported here. Our test is largely an *abrasion* test. At the rotational speed of 10 rpm (as given in the paper), the coke *slides* from one end of the tube to the other.

The effect of abrasion is shown by the discharged coke which consists of fairly large pieces of well-rounded coke, and a lot of dust. At other speeds of rotation not described in the paper, say at 15 rpm, the coke does drop from end to end in the tube. Under these conditions, we do have some impact and shatter, and some of the remarks made by Mr. Mulcahy are applicable. Just as Mr. Mulcahy implies, we *also* are not enthusiastic about shatter and tumbler tests of conventional types. For this reason, we have developed the abrasion test described in the paper.

Finally, Mr. Mulcahy says that the industry has accumulated vast information on coke properties and that any future identification must start from this point. We agree that there is a lot of information, but we still don't know how to use it *to predict* beforehand whether a particular coke will give us the performance we want. Bluntly, we consider the currently available information inadequate. We agree that we can weed out many unacceptable cokes on the basis of present tests such as those for volatile matter, ash content, room-temperature shatter, etc. However, we have never learned how to use the existing tests to *guarantee* beforehand that we will obtain in service the type of performance we want. There seems to be a missing link somewhere.

The work just reported was devoted to an attempt to determine whether the mechanical strength of coke at elevated temperatures is that missing link. So far, the authors have not recommended the test for general use. There is more work to be done. Even in its present state, however, the test has already been used to get some people out of trouble,

and to keep some other people from getting into trouble. We are sure that you will hear more about high-temperature tests for coke.

D. E. KRAUSE[2] *(Written Discussion):* This paper represents a worthwhile contribution to the subject of testing of coke. It would be of considerable interest to extend the investigation of high temperature strength of coke to foundry cokes. Since the conclusions drawn were aimed at cupola operation, some of the statements made and conclusions drawn are open to question.

In the description of the preparation of the samples, it was stated that the distribution of particle sizes in the sample was identical with that in the gross lot of coke. It is subsequently stated under operating procedure that coke was screened through a series of sieves, the coarsest of which had 3-in. square openings. There is very little coke used in cupola melting operations where all of the coke would pass through a 3-in. opening. For many of the larger cupolas, the coke is of such a size that many of the pieces would not be able to pass each other in a 10-in. diameter retort.

The following major conclusion, "The mechanical strength of coke at room temperature gives no indication of its mechanical strength at the higher temperatures encountered in cupola operation" is not supported by the data shown. Since only one coke is identified as foundry coke in Fig. 9, the only conclusion which could be drawn from Fig. 9 is that the strength of foundry coke is approximately the same at 2500 F as at room temperature.

Since very little foundry coke is produced with a coking cycle as short as 18 hr, the other data in Fig. 9 indicating an increase in strength with longer coking time would suggest that foundry cokes may not show as great a decrease in strength by raising the testing temperature from room to 2500 F as found for blast furnace cokes. The all inclusive conclusion is therefore not justified.

Although there is no justification for assuming the strength of coke at high temperatures to be the same as that at room temperature, the cupola operator has an opportunity to partially evaluate the high-temperature strength by correlating cupola operating data. It has been shown by a number of investigators that most of the resistance to air flow through the cupola is developed in the coke bed. For a given type of charge material, blast rate and tapping temperature, the blast pressure will depend primarily on the size of the particles in the coke bed and the height of the bed.

If a screen analysis of the coke as it goes into the cupola was made and correlated with blast pressure, higher than normal pressures would indicate a decrease in coke size in the cupola. A coke with low strength will usually show a high rate of degradation at room temperature so that higher blast pressures are nearly always encountered when using such cokes. It is not too often that high blast pressures are obtained with a coke which appears to be strong at room temperature. The results in the paper indicate that at such time when high blast pressures are encountered with a coke of satisfactory size at room temperature, the strength of the coke at 2500 F may be low.

MESSRS. VARGA and LOWNIE *(Authors' Reply To Written Discussion By Mr. Krause):* The authors appreciate the helpful comments and suggestions made by Mr. Krause.

The authors agree that the reported data would have been of more immediate value to foundrymen if this investigation had included more foundry cokes, and the authors have made a statement to this effect in the first paragraph of their Conclusion. Most of the work was done on blast-furnace coke because it was of most immediate interest, but the methods are, of course, applicable to foundry coke.

2. Technical Director, Gray Iron Research Institute, Inc., Columbus, Ohio.

The paper is intended to present two items of new information, (1) a description of a new method for testing the mechanical strength of coke, and (2) data to show that the mechanical strength of coke is influenced by testing temperature. In his comments, Mr. Krause offers useful observations on foundry coke in particular. As shown by the title, the paper deals with coke in general, and is not limited to foundry coke. It is, as Mr. Krause suggests, entirely possible that foundry coke may behave differently than blast-furnace coke at elevated temperatures.

The size of the cokes evaluated during this program, and the selected capacity of the equipment which has been described, were set by the requirements of the job to be done. The description of this particular equipment in the paper should not be taken to imply that other equipment of the same type to handle larger sizes of coke cannot be constructed and used.

T. H. BURNS[3] and R. A. DYKE, JR.[4] *(Written Discussion):* Messrs. Varga and Lownie have done an excellent job in determining the effect of temperature on the mechanical strength of coke and correlating these results with coking time and temperature at the coke oven.

We have been investigating hot temperature properties of coke since early this year and have attempted to evaluate hot impact strength test results with coke performance at the same cupola. We have found in our tests that there is a correlation between screen size after hot impact and coke performance. We hope to make this test accurate enough to use hot impact results to forecast the amount of a coke required to give equivalent day to day performance. We shall be glad to report our findings when we have completed our study.

We feel that the foundryman needs an impartial, quick test that can be used as a guide to predict cupola performance of coke as received at the foundry.

MESSRS. VARGA and LOWNIE *(Authors' Reply To Written Discussion By Messrs. Burns and Dyke):* On the written discussion submitted by Central Foundry Co., and a supplementary letter sent to the authors with more details, show that Mr. Burns and Mr. Dyke are thinking along the same lines as we are, and that we both have the same objective in mind, although our techniques for testing at high temperatures are different.

Early in our work (several years ago), we considered (among others) the method of testing used at Central Foundry, but we rejected the method in preference to the method we have described. This rejection, however, was arbitrary, because we have never tried their method. For all we know, it may give better results than our method.

Certainly it is to be hoped that the results of the work at Central Foundry can be presented to AFS when they are completed. By the same token, Battelle considers that its work on this subject is incomplete, and that the research should be extended by everyone who is interested, so that more information can be presented at a later date.

SECRETARY WALTON: A practical test for the hot strength of coke is available to the foundryman. Just before charging the cupola, after the bed is in, place a flat plate on the bed with a steel rod attached to it at one end. This rod will be upright near the wall of the cupola. Place a loose-fitting scrap pipe over the rod to prevent binding. The top end of the rod is observed while the cupola is being filled.

This is an ingenious method but be sure to check it with good practice since all coke beds will settle 12 to 18 in. on normal burning in.

3. Technical Director, The Central Foundry Co., Holt, Ala.
4. Chief Metallurgist, The Central Foundry Co., Holt, Ala.

SERVICE LIFE OF IRON CASTINGS CAN BE AFFECTED BY THEIR THERMAL CONDUCTIVITY

By

J. A. Davis*, H. W. Deem*, and H. W. Lownie, Jr.*

ABSTRACT

Thermal conductivity can have a major influence on performance of a casting in service. Differences in thermal conductivities of irons may affect design specifications in many applications involving heat transfer.

Quantitative data are given on the coefficient of thermal conductivity of five irons between 150 F and 600 F. If thermal conductivity of common unalloyed flake-type pearlitic iron at 200 F is taken as 100 per cent, then alloyed flake-type irons at 200 F have thermal conductivities of about 65 per cent, and an alloyed nodular pearlitic iron has a thermal conductivity of less than 50 per cent.

Four service conditions which make thermal conductivity an important factor in design and service life are discussed.

Foundrymen, engineers, and designers can easily get themselves into difficulty by ignoring the thermal conductivity of gray iron.

When dealing with gray iron, designers and engineers usually do not give thermal conductivity even a passing glance. It is common knowledge that the thermal conductivity of gray iron is lower than that of carbon steel, but that it is high enough for most types of service where gray iron is needed.

In the usual course of events, foundrymen have little reason to worry about the differences in thermal conductivity between different grades of iron. Not only do simple unalloyed gray irons differ little in thermal conductivity, but gray irons of many different compositions have served meritoriously for years in heat-transfer service of many different types.

Modern industrial progress, however, constantly demands the use of materials of higher strength. Let us show a hypothetical example of where these demands can lead. A vital part for an internal-combustion engine is cast from an unalloyed gray iron with a tensile strength of 30,000 psi. The casting has given good service for many years, but has some tendency to firecheck, oxidize, and warp because one part of the casting is subjected to alternate heating and cooling. The heat at the hot face is dissipated by means of a water jacket in another part of the casting. Then a change in the design of the engine suggests that an iron of higher tensile strength be used, and, since we all know that alloying usually improves the heat resistance of gray iron, it is decided to kill two birds with one stone by switching from the unalloyed iron to an alloyed iron with a tensile strength of 60,000 psi. This looks sensible to everyone concerned, so it is disappointing to find that actually the new iron does not stand up so well as the original iron.

In this predicament, a logical move seems to be to change to the iron which everyone *knows* will do the job. Hang the expense, use nodular iron! With sighs of relief all around, an alloyed nodular iron casting with a tensile strength of 90,000 psi is put into service—but now the part overheats and firechecks, the engine runs too hot, and the part is completely unsatisfactory. Now we are forced to a last resort—try the original unalloyed iron again! Much to everyone's surprise, it outperforms the high-priced premium-grade irons. Why?

The addition of almost any alloying element to gray iron will decrease its thermal conductivity. Furthermore, a change from flake graphite in the iron to

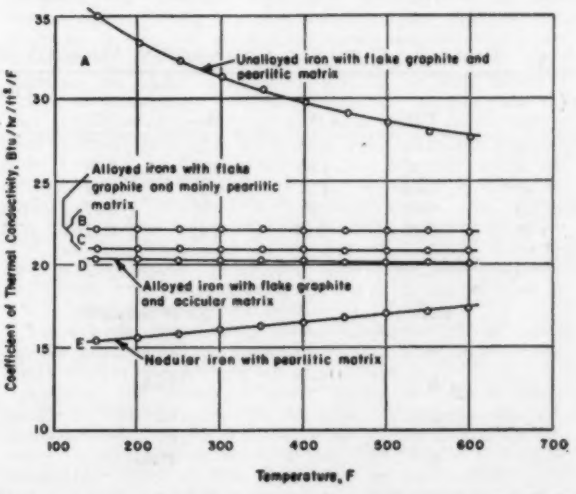


Fig. 1 — Effect of temperature on thermal conductivity of five cast irons.

*Battelle Memorial Institute, Columbus, Ohio



Fig. 2 — Typical microstructure of unalloyed Iron A. Flake graphite in a predominantly pearlite matrix. Mag. — 500 X.

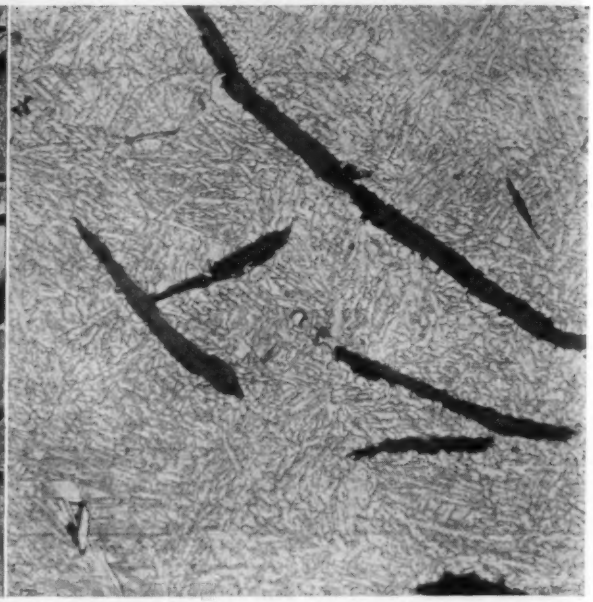


Fig. 3 — Typical microstructure of alloyed Iron D. Flake graphite in a matrix of acicular ferrite. Mag. — 500 X.

spheroidal graphite causes an additional drop in thermal conductivity. To express the comparison in round numbers, some alloyed gray irons at 200 F have only about 65 per cent of the thermal conductivity of ordinary unalloyed flake-type iron at the same temperature, and alloyed nodular iron has only about 50 per cent of the thermal conductivity of ordinary unalloyed flake-type iron.

Actual measured thermal conductivity values for five cast irons are shown in Fig. 1. Notice that the difference in thermal conductivity among the irons is greatest at about 100 to 300 F, and that the thermal conductivities approach each other as the temperature is increased. At about 1500 F or 1600 F the irons used for illustration may have almost equal values for thermal conductivity.

The values shown in Fig. 1 were determined according to the method described by Van Dusen and Shelton.† The chemical analyses of the irons are given in Table 1, and the typical microstructures of three of the irons are illustrated in Figs. 2, 3, and 4.

Applications in which the thermal conductivity of cast iron can exert an important effect on the performance or life of a casting will usually involve all or most of the following conditions:

†M. S. Van Dusen and S. M. Shelton, "Apparatus for Measuring Thermal Conductivities of Metals Up to 600 C." National Bureau of Standards, *Journal of Research*, vol. 12, 1934, pp. 429 to 440.

TABLE 1 — CHEMICAL COMPOSITION, MICROSTRUCTURE, AND TENSILE STRENGTH OF FIVE CAST IRONS

		Composition, per cent								
Iron	T.C.	Si	Mn	P	S	Ni	Cu	Mo	Mg	Sn
A	3.50	2.25	0.70	0.07	0.11	•	•	•	•	•
B	2.92	1.80	0.71	0.08	0.11	2.05	0.10	0.68	•	•
C	2.89	1.51	0.87	0.09	0.10	1.76	0.53	0.60	•	•
D	2.70	2.00	0.95	0.07	0.11	2.20	1.05	1.00	•	•
E	3.50	2.10	0.41	0.11	0.02	2.25	0.88	1.05	0.03	0.05

•only residual amounts

Iron	Type of Graphite	Type of Matrix	Tensile Strength, psi
A	Flake	Pearlite	25,000 in 1.2-in. bar
B	Flake	Mostly pearlite, some acicular ferrite	45,000 in 3-in. wall
C	Flake	Mostly pearlite, some acicular ferrite	45,000 in 3-in. wall
D	Flake	Acicular ferrite	60,000 in 3-in. wall
E	Nodular	Pearlite	80,000 in 3-in. wall

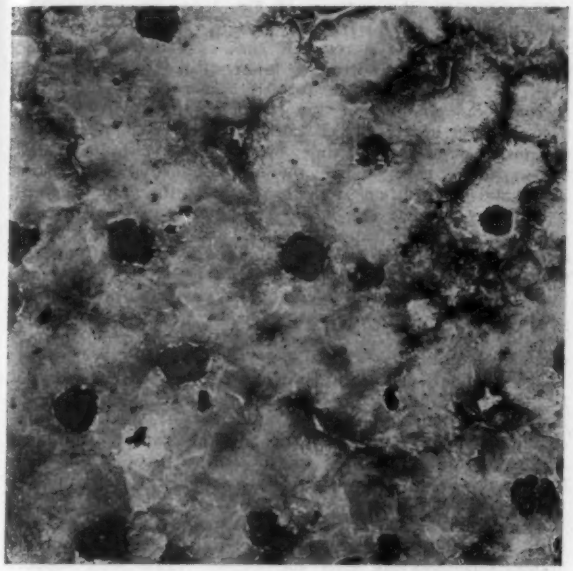


Fig. 4 — Typical microstructure of alloyed Iron E. Nodular graphite in a matrix that is predominantly fine pearlite. Mag. — 500 X.

(1) The flow of heat is unsteady or cyclic; often one face of the casting is alternately heated and cooled (as on the hot face of the pistons in a two-cycle diesel engine).

(2) Metal sections are thick (as in hollow water-cooled rolls used to form glass).

(3) There is rapid transfer of heat to the iron or away from the iron (as when the surface of a brake drum is heated by friction).

(4) There is a fairly large difference in temperature between the hot and cold faces of the casting.

If these conditions are not encountered in a particular application, the thermal conductivity of the iron may have little influence on the rate at which heat is transferred, or on the performance or life of the casting. As an example of this, consider an ordinary steam radiator used for space heating. The thermal conductivity of the metal itself is a minor factor—a radiator of high-conductivity copper would have little advantage thermally over a radiator of low-conductivity stainless steel. The thermal resistance of the thin metal wall is almost insignificant in comparison with the thermal resistance encountered in the slow-moving air and steam on each side of the metal. Furthermore, the flow of heat is fairly steady and involves only a small difference in temperature.

Iron castings are used in many applications which involve all or most of the four critical conditions listed above. In these applications, the thermal conductivity of the iron can have an important effect on performance and service life. These applications fall into two general types:

(1) Applications where we desire to transfer heat through the iron so that the heat can do a desired job on the cooler side of the casting wall. Such applications include rolls of many types used for heating or drying, large boilers, molds for plastics or glass, electrically heated patterns for shell molding, and laundry irons.

(2) Applications where we desire to transfer heat through the iron so as to get it away from a hot face. Such applications include brake drums, rolls and molds for forming glass, automotive engine blocks, cylinder linings, diesel pistons, hot-work rolls, ingot molds, permanent molds, and so on.

If the application is of the first type, where we are trying to put heat through the iron and *into* a product of process, a lower thermal conductivity may cause us to get less useful heat where we want it. Our process may run too cold. If the application is of the second type, where we are trying to move heat through the iron to get it *away* from a hot area, a lower thermal conductivity may increase the tendency for the heated face to heat check, oxidize, grow, and warp. Our process and our iron will run too hot.

By intelligent use of the higher strength of alloyed irons and nodular irons, it is sometimes possible to reduce the thickness of the casting wall. The use of the thinner wall will tend to offset the effect of the lower thermal conductivity of these irons. Often, however, the use of the thinner wall will nullify the advantages of using the stronger iron in the first place.

Experience has shown that, when an application involves the transfer of heat through an iron casting, the thermal conductivity of the iron must be considered. This is particularly true when a change in specification is being made to call for the use of a stronger iron or a more heavily alloyed iron.

DISCUSSION

Chairman: C. K. Donoho, American Cast Iron Pipe Co., Birmingham, Ala.

Co-Chairman: W. R. Jennings, Deere & Co., Waterloo, Ia.

Secretary: S. G. Johnson, Jr., International Harvester Co., Indianapolis, Ind.

D. E. Krause[1] *(Written Discussion):* This paper, through discussion of hypothetical examples, raises some interesting questions. Of the data reported, particularly the thermal conductivities in Fig. 1, the unusually high thermal conductivity of iron "A" is of considerable interest. The tensile strength of iron "A" is given as 25,000 psi in a 1.2-in diameter bar which is somewhat low for an iron having a carbon equivalent of 4.25 per cent. This suggests that the iron may have an appreciable amount of ferrite in the matrix although it is stated that the matrix is predominantly pearlitic.

In view of a silicon content of 2.25 per cent, it is surprising that the conductivity of iron "A" at 200 F should be about 15 per cent higher than that of an iron having the same carbon equivalent but a silicon content of about 1.50 per cent. Since it is known that an increase in silicon content will decrease the thermal conductivity, the data seem to indicate that composition of gray iron cannot be used as a basis for estimating thermal conductivity.

There is an obvious error in the designation of the values on the ordinate of the graph in Fig. 1. The thickness was omitted and it should be per foot of thickness.

Some of the statements made should have been better qualified. The statement, "It is common knowledge that the thermal conductivity of gray iron is lower than that of carbon steel—" is open to question. Unalloyed iron "A" has a thermal conductivity equal to that of an 0.08 per cent carbon steel and about 10 per cent better than a 0.23 per cent carbon steel. Published data on thermal conductivity of unalloyed

1. Technical Director, Gray Iron Research Institute, Inc., Columbus, Ohio.

gray iron having 4.0 per cent T.C. and 1.5 per cent Si indicate such an iron to have a thermal conductivity equal to a 0.23 per cent C steel.

The use of the term "acicular ferrite" in Fig. 3 and Table 1 is rather unusual as the type of structure illustrated is usually designated as simply an "acicular matrix". The term used in the paper gives the impression that the matrix of iron "D" shown in Fig. 3 is essentially free of combined carbon which is quite unlikely.

It is unfortunate that service data relating thermal conductivity with performance of the castings were not included. Although the arguments supporting the possibility of thermal conductivity of iron playing an important role in service life of gray iron sound reasonable, there are other factors such as modulus of elasticity and coefficient of expansion which influence service life of castings subjected to cyclic heating and cooling. It certainly should be worthwhile investigating the thermal conductivity of gray irons giving either unsatisfactory service life or unusually good service life.

MESSRS. DAVIS, DEEM and LOWNIE *(Authors' Reply To Written Discussion By Mr. Krause):* We would like to thank Mr. Krause for his comments, and agree generally with the various points he has made.

The tensile strength listed in Table 1 for Iron A was estimated from its composition and microstructure. The estimate of 25,000 psi may be slightly low, as suggested by Mr. Krause, but is expected to be within 3000 psi of the true strength. The other tensile strengths listed in Table 1 were actually measured.

The units for the ordinate of Fig. 1 are incomplete. The units should be Btu/hr/ft^2/ft/F.

The statement "It is common *knowledge* that the thermal conductivity of gray iron is lower than that of carbon steel, . . ." should more correctly have been worded "It is common *opinion* that the thermal conductivity of gray iron is lower than that of carbon steel, . . ." When tests are made at the same time by the same technique, low-carbon steels have higher thermal conductivity than cast irons, and high-carbon steels have about the same thermal conductivity as some cast irons. The thermal conductivities of cast iron and steel appear to be about equal for equal amounts of combined carbon. Published values are, however, quite conflicting. On this point, the authors wish to point out that it was not their intention to compare the thermal conductivity of gray iron with that of steel. The intention was to show that cast iron can have a variable thermal conductivity.

The term "acicular ferrite" was apparently used originally by D. J. Reese and R. A. Flinn in an AFS publication to describe a particular structure obtained in some cast irons containing nickel and molybdenum. The authors continued the use of this terminology for the same structure. The term "acicular structure" is probably better, but neither term appears to be entirely satisfactory.

The authors do have examples relating thermal conductivity with the performance of gray iron castings in two specific applications involving rolls and brakes. These data support the statements made in the paper. The specific examples, however, have not been released for publication.

HEAT TREATMENT OF GRAY-CAST IRON

By

A. H. Rauch* and J. B. Peck**

ABSTRACT

Gray iron, like steel, can be strengthened and hardened by quenching from above the upper critical temperature followed by tempering. Hardenability of both alloyed and unalloyed irons in the strength range of ASTM Class 25 to Class 35 irons was determined by means of the end-quench hardenability test. Effect on hardenability of chromium, nickel and molybdenum singly and in combination was explored.

Conventional Rockwell C and brinell tests made on heat treated gray iron are an integration of the hardness of the graphite and metal matrix. Graphite crushes under the indenting load and thus true hardness of the metal matrix is not revealed. Microhardness tests were made on as-quenched gray iron samples differing in Rockwell C hardness and on samples tempered at several temperatures. It was shown that the matrix hardness of irons differing in Rockwell C hardness was essentially the same. Response of the hardened metal matrix to tempering was determined.

Effect of tempering on the tensile strength, flexural strength, impact resistance, modulus of elasticity, and hardness was explored.

Introduction

Gray-cast iron, like steel, can be heat treated by quenching and tempering to obtain increased strength and hardness. This has been demonstrated by other investigators.† It is the practice, however, of the foundryman to obtain increased strength by addition of alloying elements and control of carbon and silicon content. There are probably many applications where strength can be increased more economically by heat treatment.

The main application of heat treatment of gray iron to date has been to increase wear resistance — chiefly by selective or localized hardening. Furnace hardening or through hardening is often avoided because of possible distortion and also because of concern over loss in toughness. This investigation was undertaken to determine the effect of tempering temperature on strength, impact resistance, matrix hardness, and stress-strain relationships of quenched gray iron and to compare these properties with those of the as-cast material.

In any discussion of heat treatment the question of hardenability of the material becomes inportant. The end-quench hardenability test used for steel was employed in this investigation of gray iron. An estimate was made of the maximum section size of unalloyed iron that can be hardened to 50 per cent martensite at the center in an oil quench. Effect of alloying elements on hardenability was explored to a limited extent.

Hardness of quenched and tempered gray iron, determined with the conventional Rockwell or brinell tests is the resultant or integration of hardness of the metal matrix and the graphite flakes. Hardness of the metal matrix is more indicative of wear resistance of the material than the composite or integrated hardness. This paper includes an investigation of the matrix hardness of quenched and tempered gray iron determined with microhardness tests.

Gray iron has been called steel plus graphite. The metal matrix is essentially a high-silicon, medium-carbon steel. Like steel, an upper critical temperature must be exceeded in order to secure an uniformly hardened structure. Unlike steel, however, the carbon content of the metal matrix is dependent upon time and temperature employed in heating for hardening. This is due to presence of excess carbon in the form of graphite. In order to consistently harden satisfactorily the full range of compositions of the Class 25 to Class 35 irons investigated, all samples were heated to 1600 F and soaked at temperature for 30 minutes.

No paper on the heat treatment of gray iron would be complete without mention of austempering. Cast iron can be austempered, i.e., quenched into a bath maintained at about 450 F to 900 F and held until transformation is complete. Bartholemew* has shown substantial improvement in strength and wear resistance over that of the as-cast iron by quenching into a bath at 510 F. This treatment also minimizes distortion. Some preliminary

*Senior Metallurgist and **Metallurgist, Materials Engineering Department, Deere & Co., Moline, Ill.

†See selected bibliography at the end of the paper.

*"Gray Cast Iron — A New Heat Treatment," E. L. Bartholemew, *Iron Age*, vol. 146, no. 5, August 1, 1940, pp. 52-54.

work by the authors has indicated that the impact resistance may be superior to that of the quenched and tempered irons reported in this investigation.

Hardenability

Hardenability tests were made on 26 bars of unalloyed Class 25 through Class 35 iron and 15 bars of Class 30 iron alloyed with chromium, nickel, molybdenum and copper, both singly and in combination.

The standard end-quench test, used for determining the hardenability of steel, was employed. All bars were austenitized for 30 minutes at 1600 F, then end-quenched in a standard fixture for 15 minutes. Bars containing 0.6 per cent to 0.7 per cent chromium were also austenitized for 1 hr. at 1700 F and end-quenched in the same manner. After quenching 0.015 in. was ground from the surface of the bar at two locations 180 degrees apart. A Rockwell C hardness survey at $\frac{1}{16}$-in. intervals was made on these flats starting at $\frac{1}{16}$ in. from the quenched end and extending to a distance of $4\frac{9}{16}$. The hardenability test results are summarized in Table 1.

Seven bars were metallographically examined on a polished flat to determine the distance to 50 per cent martensite.

The range of $\frac{6}{16}$ in. between the maximum and minimum distance, whether to C40 or C30 of the unalloyed irons is narrower than the specified bands for alloy steel. This is true in spite of the wide range in carbon, silicon, manganese and residual alloys of these irons. This hardenability corresponds to that of a medium-carbon, low alloy steel. Based on a correlation with published curves,† the lowest hardenability unalloyed cast irons would quench out in oil to 50 per cent martensite or Rockwell C30 at the center of a $1\frac{1}{2}$ in. round bar while the highest hardenability iron would quench out in oil to 50 per cent martensite at the center of a $2\frac{1}{4}$-in. round.

Hardenability of gray-cast iron is not significantly affected by chromium in amounts of 0.20 per cent to 0.75 per cent whether used singly or in combination with other elements. Copper and nickel have little effect on hardenability when used singly but nickel appears to accentuate or reinforce the effect of molybdenum and the same would be expected of copper. Molybdenum by itself has only a moderate effect but, combined with nickel, has a potent effect on increasing the hardenability of gray-cast iron.

Failure of chromium to increase the hardenability is thought to be due to the fact that it combines with carbon in the cementite associated with iron phosphide in the ternary type of steadite. Austenitizing at 1700 F for 1 hr was tried on the chromium-bearing irons in an attempt to break down the chromium carbides. A slight increase in hardenability was noted. The results are shown in Table 2.

†"A Correlation of End-Quenched Test Bars and Rounds in Terms of Hardness and Cooling Characteristics," E. W. Weinman, R. F. Thomson and A. L. Boegehold, *ASM Trans.*, vol. 44, 1952, pp. 803-834.

TABLE 1 — HARDENABILITY OF IRONS AUSTENITIZED AT 1600 F

Type of Iron	Rockwell Hardness $\frac{1}{8}$ in. from Quenched End	*Distance to C40	*Distance to C30
**Unalloyed	C44-52	7-13	9-15
**Alloyed			
Cr 0.2-0.4%	C49-51	7-9	9-12
Cr 0.6-0.75%	C51-53	8-9	12-14
Ni 0.55%	C45	10	14
Cu 0.58%	C48	8	11
Mo 0.48%	C50	15	20
Cr 0.20%, Mo 0.18%	C52	12	14
Cr 0.72%, Mo 0.42%	C53	18	20
Cr 0.75%, Ni 0.55%	C50	11	13
Ni 0.52%, Mo 0.61%	C51	26	36
Cr 0.72%, Ni 0.56%, Mo 0.52%	C51	24	28

*Distance to C40 and C30 is in sixteenths of an inch from the quenched end of the bar. Metallographic examination showed 50% martensite, 50% fine pearlite to coincide with Rockwell C30.

**Composition Range

Unalloyed Irons

TC	Si	Mn	P	S
3.25-3.55	2.05-2.75	0.50-0.75	0.15-0.50	0.10-0.15

Ni	Cr	Mo	Cu
0.01-0.09	0.03-0.30	Nil-0.06	0.09-0.53

Alloyed Irons

TC	Si	Mn	P	S
3.15-3.55	2.25-2.60	0.40-0.60	0.30-0.50	0.11-0.16

Composite Hardness Versus Matrix Hardness for As-Quenched and Quenched and Tempered Gray-Cast Iron

End-quench tests showed a wide range in maximum hardness between bars having approximately the same composition. In order to determine if there was a definite cause for this difference in hardness and also to determine the effect of tempering on hardness, matrix hardness determinations were made on two unalloyed bars with low Rockwell C hardness and two unalloyed bars with high Rockwell C hardness at $\frac{1}{8}$ in. from the quenched end. The samples were oil quenched from 1600 F and both matrix and composite hardness were determined in the as-quenched condition and after tempering for 45 minutes at 500 F, 700 F and 900 F. Matrix hardness was determined with a 200-gram load and the Vickers square pyramid diamond.

The hardness of as-quenched, fully-hardened, gray-

TABLE 2 — EFFECT OF AUSTENITIZING TEMPERATURE AND SOAK TIME ON CHROMIUM ALLOYED IRONS

Type of Iron	Distance to C 40		Distance to C 30	
	30 min at 1600 F	1 hr at 1700 F	30 min at 1600 F	1 hr at 1700 F
Cr 0.62 - 0.75%	8-9	12-14	10	15
Cr 0.72%, Mo 0.42%	18	20	19	24
Cr 0.75%, Ni 0.55%	11	13	12	16
Cr 0.72%, Ni 0.56%, Mo 0.52%	24	28	28	36

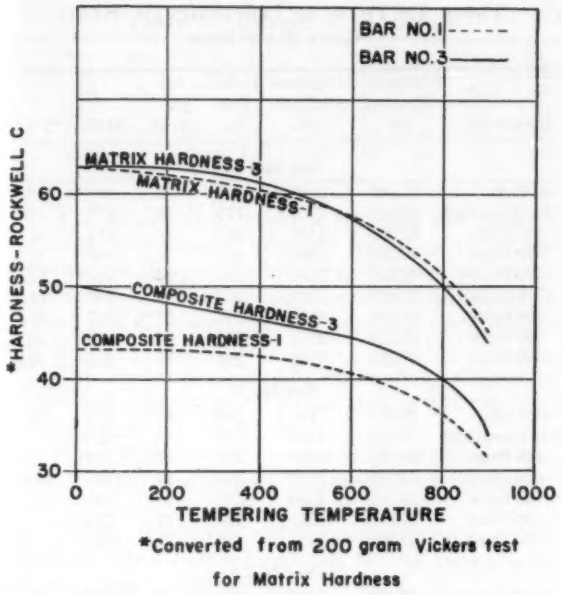


Fig. 1 — Effect of tempering temperature on composite and matrix hardness.

cast iron can vary from Rockwell C40 to Rockwell C53. The hardness determined with the Rockwell C test represents the composite hardness of the graphite plus the metal matrix. Hardness of the metal matrix of representative as-quenched gray-cast irons determined with microhardness tests is equivalent to Rockwell C61-63 regardless of the actual Rockwell C test value. Spalling occurs around the Rockwell impressions at high hardness levels which may account for the erratic results obtained. Cause for this spalling was not determined but graphite flake size and possibly arrangement and graphitic carbon content may be factors.

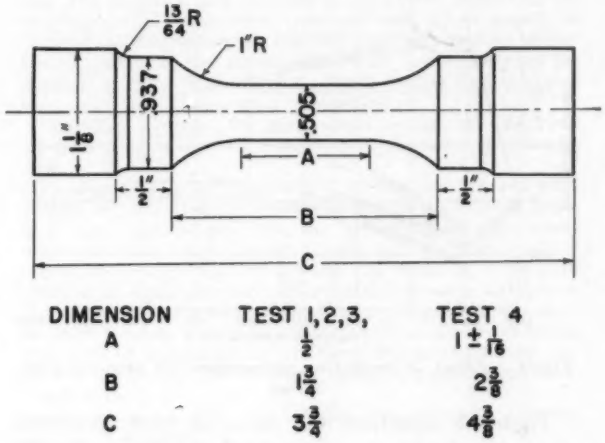
The matrix hardness produced by tempering is about the same as that of a 2 per cent silicon, 0.50 per cent to 0.60 per cent carbon steel. The matrix hardness of tempered gray iron is about 10 to 20 points Rockwell C higher than the actual Rockwell C test value, depending upon the tempering temperature.

Although there is a large variation in composite hardness from bar to bar, the matrix hardness is almost identical as shown in Fig. 1 and Table 3.

Physical Properties of Heat Treated Gray-Cast Iron

Physical properties of four sets of quenched and tempered bars, each poured from the same ladle, were investigated. Physical properties investigated were transverse bending, impact resistance, ultimate tensile strength and modulus of elasticity.

For sets No. 1 and No. 2 whole 1.2 in. arbitration bars were austenitized at 1600 F for 30 minutes at heat, then oil quenched. The bars were then tempered at the desired temperature for 1 hr total time. Transverse bending tests were made on the bars. The broken halves were cut to 8-in. lengths and centerless ground to 1.125-in. diameter and impact tested without further heat treatment. The broken halves of the bars were then tempered at 1200 F for 2 hr at temperature to facilitate machining and machined to a 0.505-in. gage diameter, shoulder-end tensile specimen as shown in Fig. 2. The tensile bars were then heat treated in the same manner as the arbitration bars and tested. The same procedure was used for sets No. 3 and No. 4 except that the tensile bars were machined from as-cast duplicate bars.



DIMENSION	TEST 1,2,3,	TEST 4
A	$\frac{1}{2}$	$1 \pm \frac{1}{16}$
B	$1\frac{3}{4}$	$2\frac{3}{8}$
C	$3\frac{3}{4}$	$4\frac{3}{8}$

Fig. 2 — Tensile test bar specimen.

The modulus of elasticity was determined on sets No. 1 and No. 3 by placing SR-4 strain gages 180 degrees apart on one of the tensile specimens at each tempering temperature. The average strain was used in plotting the stress-strain curves. The modulus of elasticity was computed at one-fourth the breaking load.

Tensile strength and impact resistance tests were made in duplicate. Transverse bending strength and modulus of elasticity were determined on a single specimen.

Table 3 — Comparison of Rockwell C and Microhardness on As-Quenched and Quenched and Tempered Gray Iron

Bar No.	1/8 in. on Hardenability Bar	Actual or Equivalent Rockwell C			
		As Quenched		500 F Temper	
		Actual	DPH*	Actual	DPH*
1	45	43	63	43-40	59
2	46	43	63	43-40	60
3	51	50	63	45	60
4	50	49	61	45	60
		700 F Temper		900 F Temper	
		Actual	DPH*	Actual	DPH*
1		38-40	55	30-33	45
2		35-40	53	28-30	43
3		42	54	33-34	44
4		43	52	34-35	47

*Rockwell C equivalent of hardness determined with 200-gram Vickers test on Tukon tester.

All actual Rockwell C values except for the hardness at 1/8 in. from the end-quench bars represent the range of five determinations. All microhardness values are the average of ten determinations.

Tests were made according to appropriate ASTM specification:

Transverse Bending – ASTM A48-48. 1.2-in. bars broken on 18-in. supports.

Tension Testing – ASTM A48-48. Specimen shown in Fig. 2.

Impact Testing – ASTM A327-54. Charpy type impact test 1.125-in. diam bar, 8 in. long, broken on 6-in. supports.

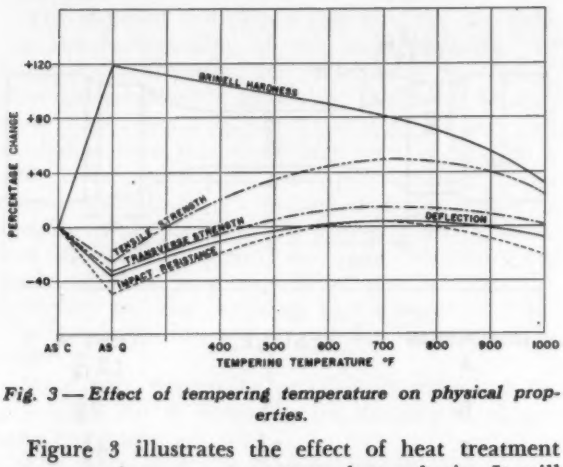


Fig. 3 — Effect of tempering temperature on physical properties.

Figure 3 illustrates the effect of heat treatment on properties on a percentage change basis. It will be noted that all strength properties in the as-quenched condition are below the corresponding values in the as-cast condition. Tempering at 400 F restored the tensile strength to that of the as-cast material while tempering at about 600 F was required to restore the transverse strength, deflection, and impact resistance. Tempering at 700 F appears to show a peak in all strength properties.

The percentage values shown in Fig. 3 represent the average change values of the four sets of bars shown in Table 4.

The higher strength as-cast irons generally show a greater percentage increase in tensile properties than the lower strength irons. Cast iron is a heterogeneous material. In addition to the metal matrix, graphite flakes and steadite colonies are also present. The structure and properties of the matrix can be changed by heat treatment but the graphite and steadite are not significantly affected by the temperature ordinarily employed in this type of heat treatment. Therefore, an unfavorable amount of distribution of these constituents cannot be altered by heat treatment and any effect they have on the physical properties remains unchanged.

The modulus of elasticity of the iron is apparently decreased by heat treatment, but in the heat treated state it is fairly constant regardless of tempering temperature. The stress-strain curves shown in Fig. 4 approach a straight line in quenched and tempered irons.

Conclusions

1. Gray-cast iron will respond uniformly and satisfactorily to heat treatment within the limits of a suitable range of section size commensurate with its hardenability.

TABLE 4 – PHYSICAL PROPERTIES OF HEATED GRAY CAST IRON

Treatment	Ultimate Tensile, psi	Transverse Strength, lb	Deflection, in.	Impact, ft-lb	BHN	*E×10⁶
			Set No. 1			
As-Cast	31,900	2465	.254	32	217	13.6
As-Quenched	25,125	1730	.175	18	477	11.7
400 Draw	42,000	2190	.202	27	444	12.1
500 Draw	42,750	2350	.223	31	429	11.1
600 Draw	47,325	2800	.255	28	401	12.2
700 Draw	47,625	2920	.292	36	388	11.0
800 Draw	45,900	2670	.244	33	341	11.2
900 Draw	43,975	2560	.246	30	293	10.9
1000 Draw	41,250	2400	.234	25	269	10.6
			Set No. 2			
As-Cast	26,875	2165	.246	29	192	
As-Quenched	19,185	1240	.144	14	415	
400 Draw	29,400	2070	.207	21	388	
600 Draw	35,500	2450	.253	24	375	
700 Draw	35,300	2580	.257	31	363	
800 Draw	37,000	2350	.237	27	331	
1000 Draw	32,200	2140	.225	25	269	
			Set No. 3			
As-Cast	38,775	2610	.269	40	217	15.6
As-Quenched	30,450	1870	.184	17	477	13.5
400 Draw	43,490	2590	.250	30	444	14.5
500 Draw	52,575	2620	.238	32	429	13.6
600 Draw	50,675	2680	.247	33	415	14.3
700 Draw	57,965	2830	.256	37	401	14.0
800 Draw	55,265	2630	.229	31	363	14.4
900 Draw	54,750	3050	.270	36	331	14.2
1000 Draw	49,100	2740	.254	34	285	13.4
			Set No. 4			
As-Cast	35,690	2500	.259	32	217	
As-Quenched	25,810	1740	.180	18	477	
400 Draw	42,375	2630	.257	27	429	
600 Draw	48,750	2870	.268	35	415	
700 Draw	61,125	2730	.248	37	401	
800 Draw	57,065	2870	.263	31	352	
1000 Draw	42,875	2550	.243	29	285	

*E = Modulus of elasticity determined at one-fourth of the ultimate strength.

2. Chromium, nickel and copper when used singly do not affect the hardenability appreciably. Molybdenum by itself raises hardenability only moderately but when used in combination with nickel is very effective in increasing hardenability.

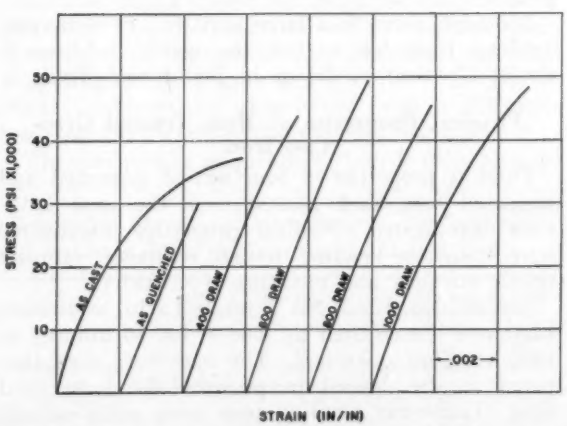


Fig. 4 — Effect of tempering temperature on stress-strain curve (Set 3).

TABLE 5 — CHEMICAL ANALYSIS

Set	TC	Mn	P	S	Si	Ni	Cr	Mo	Cu
1	3.38	.60	.24	.14	2.32	.07	.15	Nil	.16
2	3.59	.45	.44	.15	2.15	.05	.12	Nil	.12
3	3.21	.54	.23	.13	2.12	.08	.06	Nil	.13
4	3.19	.57	.21	.12	2.42	.06	.04	Nil	.07

3. The Rockwell C hardness test does not indicate the actual hardness of the metal matrix in either the quenched or tempered condition. Samples given the same heat treatment may show appreciable differences in Rockwell C values but will have the same matrix hardness.

4. The as-quenched matrix responds to tempering in the same manner as a 0.50 per cent to 0.60 per cent carbon, 2 per cent silicon steel.

5. Gray iron in the as-quenched condition has lower strength and impact resistance than the as-cast iron.

6. Impact resistance and deflection values in the transverse bend test are restored to the as-cast level by tempering at 600-800 F.

7. The transverse bending strength is restored to as-cast by tempering at a minimum of 500 F. After tempering at 600-900 F these values are on an average of 10 per cent above the as-cast values.

8. On tempering at 400 F the tensile strength is appreciably above the as-cast strength. A peak in tensile is reached on tempering at 600-800 F with values from 38 per cent to 72 per cent above the as-cast values.

9. The secant modulus of elasticity is decreased slightly by heat treatment.

Acknowledgment

The authors wish to thank F. T. McGuire, Manager; C. W. Ray, Metallurgical Supervisor, and C. A Davis, Mechanical Supervisor, of the Materials Engineering Department, Deere & Company, for their counsel and guidance.

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DISCUSSION

Chairman: H. W. LOWNIE, JR., Battelle Memorial Institute, Columbus, Ohio.

Co-Chairman: G. A. TIMMONS, Climax Molybdenum Co., Detroit, Mich.

R. A. CLARK[1]: I wish to congratulate the authors for demonstrating the fact that the matrix hardness of heat treated irons is harder than indicated by the conventional Rockwell test. The reverse is also true and it is common practice in the manufacture of automotive camshafts and valve tappets to balance the composition so that the structure contains a controlled amount of free carbide. This hard carbide increases the hardness as tested by Rockwell C. It also confers high resistance to abrasion and high pressures in operation.

CHAIRMAN LOWNIE: Do the authors regard the Rockwell test as a good method for measuring the hardness of gray cast iron?

MR. PECK: While we do not regard the Rockwell test as a good method for measuring the hardness of gray iron, we do not know of a more practical one. The use of the Brinell test, which might be better because of the larger area of indentation, is limited in its application by the size and thickness of the part being tested. Other tests have other limitations which make them impractical. We have found as much as a $\pm 2\frac{1}{2}$ points spread in Rockwell C values on a given sample which we attributed to the small area of indentation in relation to the microstructure and also the tendency which we found, of spalling around Rockwell C indentations.

R. C. SHNAY[2]: Is there a correlation between matrix hardness and tensile strength?

MR. RAUCH: We found no correlation between matrix hardness and strength. We would not expect any because of the effect of the amount, size, and distribution of graphite on strength.

NILES KITCHEN[3]: What precautions would the authors recommend in hardening thin sections along with heavy sections simultaneously in order to prevent quench cracking?

MR. RAUCH: We have no trouble with quench cracking on the parts we are now hardening; however, we have no parts with abrupt changes in section size. The best thing to do would be to avoid such designs. If this is not possible, marquenching may be helpful.

N. M. LAZAR[4] and A. M. MARS[5] *(Written Discussion):* The authors have treated very well a rather pertinent topic as heat treated gray cast iron has some remarkable possibilities.

We have studied the heat treatment of gray cast iron and have obtained results which are similar to those reported in this paper. We have also observed that the microhardness of the matrix of two irons may be the same but the average or composite hardness may be appreciably different. We have been able to correlate this with graphite distribution. Irons with type A graphite distribution gave higher composite hardness than irons with other graphite patterns.

Also we found that the wear resistance correlates better with graphite distribution than with hardness both in the as cast and heat treated conditions.

We believe that a systematic study of the effect of graphite distribution would be of great value in understanding the heat treatment of cast iron.

MR. RAUCH: We did not make any serious investigation to determine the cause for differences in composite hardness in test pieces having the same matrix hardness. It is interesting to note that Messrs. Lazar and Mars have found that graphite distribution has an effect. Lindsay and Hoke* expressed the opinion that the variation in composite hardness was due to cracking around the impressions or crumbling under the indentor. Data reported by these authors indicate that graphite flake size is an important factor.

We have no information on the relationship of graphite distribution and wear. Reports of other investigators seem to parallel the findings by the discussors, insofar as scuffing is concerned on hardened cylinder liners.

*"Effect of Graphite Flake Size on Tensile and Fatigue Properties of Gray Cast Iron," AFS TRANSACTIONS, vol. 62, pp. 321-328 (1954).

1. Electro Metallurgical Co., Detroit.
2. Dept. of Mines & Technical Surveys, Ottawa, Canada.
3. Steel Heddle Mfg. Co., Greenville, S. C.
4. Wayne University, Detroit.
5. Vickers Incorporated, Detroit.

PRACTICAL FOUNDRY APPLICATION OF STATISTICAL QUALITY CONTROL

By

Ross Martin, Jr.*

Statistical Quality Control is a method or tool whereby we are enabled to establish a yardstick to measure the variability of a process or part. Everything occurring in Nature, or made by main or machine varies. No two items are exactly alike if we use sufficiently precise measuring instruments. Once this concept can be grasped the barrier to the use of Statistical Quality Control methods is broken.

We all prefer to think in abstract terms. We like to think that 2 plus 2 is exactly 4 and never approximately 4. We may have difficulty in accepting the fact that all things vary. We rely on a 6-ft rule being exactly 6 ft long but if measured precisely enough none would be exactly 6 ft in length. Once this concept of variability is accepted then we find a need for some way to measure the variation. Once we can measure variation we will be in a position to control it.

Every characteristic of an item will have a normal or expected variation inherent in the item and other abnormal or unexpected variation caused by some extraordinary cause or event. Statistical Quality Control gives us a means of separating the usual or normal variations from the unusual or abnormal variations. It will enable us to determine when a process is operating as well as can be expected and when it is not.

These normal variations are referred to as having been caused by chance or luck. Shooting dice or playing a slot machine might be considered games of chance or luck to the individual participating but not to the gambling house. The difference is that the individual does not know how to measure the expected variations or odds while the professional gambler does. Too often decisions are based on hunches rather than facts. In other words, we are gambling with our business or with our jobs without knowing the odds.

By proper application of Statistical Quality Control we can determine the odds and can be on the "house" side. It is all tied up in the trick of knowing how much variation to expect. Then we will not take corrective action until this normal or expected variation has been exceeded. Instead of growing ulcers trying to win a hopeless battle we can devote our attention and effort to only those variations over which we can exercise some measure of control.

Details of application of the particular formula or chart will not be discussed. This information is available in several books—an excellent one being the AFS Statistical Quality Control for Foundries. Reference to the tables and formula could detract from the emphasis of the practical applications.

For a possible application, let us consider the cupola. We know that our raw material charged into the stack varies in analysis. Depending upon the source and the charging practice this variation may be great or relatively small. We also know that we will have variations in the operation of the cupola which will cause variations in the resulting analysis of the metal at the spout. Take total carbon for an example. Over a control period during which time we maintain what we consider good practice for our shop, we can determine the variation between estimated or calculated analysis and the actual analysis. We may find that the average variation may be 5 or 6 points but that it may, at times, go as high as 10 or 12 points.

Establishing Standard Variation in Process

By referring to a handbook on Statistical Quality Control we can easily establish what we can consider to be normal variations in the total carbon. In other words, if the cupola is operated in accordance with our shop practice, we will expect a certain variation in total carbon. As long as we stay within these control limits we can safely assume that our shop practice is still as good as it was during the control period from which the standard variation or control limits were established. However, if the variation on a particular heat exceeds the standards we have established or if the control limits have been exceeded we can, with confidence, know that the practice is not the same and look for the trouble.

Let me emphasize that we should not make any changes when the total carbon (in this case) is within our control limits. The only time changes should be made is when the total carbon is outside the control limits.

*Plant Manager, Rich Mfg. Co. of Calif., Los Angeles.

If changes are made while the total carbon is still within the control limits we will make matters worse. If the total carbon should be 8 points lower than was expected but is still within the control limits, and we change our mix to compensate, our average total carbon will then be 8 points too high. Just as in throwing a pair of dice, the average should be 7 but we may get a 10. It does not necessarily mean that there is something wrong with the dice. If the control limits are too wide we must improve our cupola practice. We can never outguess and improve poor practice by changing the mix when the process is still within the control limits.

Over control can be as costly as under control. By making changes when we are still within the control limits, we will actually be creating a variation when we are seeking to eliminate.

Applying Control Charts

Once we apply control charts on the total carbon and work toward eliminating all points outside of the control limits we will be running the cupola instead of it running us.

Someone may have a question in mind at this stage such as "how close can I expect to control the total carbon?" There is no standard answer. It will all depend upon the shop practice—from the purchase of the raw materials right down to the method of analysis in the laboratory. Once standard practice is established, be it good or bad in comparison with other foundries, through Statistical Quality Control we will be able to evaluate results and determine if the standard practice is being followed.

Silicon, manganese, sulphur and all the other elements of our iron analysis may be controlled in a similar manner and with amazing results. We will find that we can achieve closer control and make fewer changes than we ever dreamed.

Weighing of raw materials which make up the melting charge is an excellent application for Statistical Quality Control. The human tendency may be for the charges to always give good measures. If this is the case we will usually find that the charges will, on the average, be heavy. If the charge is made up with a magnet and it is not possible to easily remove material from the charge bucket or hopper, the tendency toward heavier average weights will be even greater. By analysis of actual charge weights we can establish control limits, so that the stock weighman will know when the weight is "good enough." Exact weights would be impractical as the weight of the increment of individual component parts will usually be large.

Although cupolas in particular were used as an example the same method of analysis and control can be applied to nonferrous melting as well.

We in the foundry industry need to know when to let the job run and when to take action. Statistical Quality Control will give us the advantage of knowing that the odds are always in our favor when we decide to let the job run or decide to make a change based on the control limits of the process.

Control limits can be used on temperatures and physical properties as well. We will be able to tell when we are doing as well as can be expected with our accepted standard practice.

Scrap and rework are ideal subjects for Statistical Quality Control applications. By studying our process we can determine the range of the scrap which can be reasonably expected and devote the maximum effort to determining the cause and remedy for only those items which have exceeded the control limits. The control limits can be expressed either as a percentage or as a number of items, depending on the local practice and custom.

Certain castings may have surface defects which would not interfere with their use but which just "look bad" if the defect is present over a wide area. Quality evaluation is possible by Statistical Quality Control. Once a procedure is set up the quality level of the castings can be controlled. With this control, it will be possible to take corrective action before actual scrap is produced. Inspectors tend to become more severe with a good lot of castings and lax with a poor lot unless the quality evaluation can be tied to standards.

Spot checking or sampling inspections are usually more effective than checking each casting individually which is usually referred to as 100 per cent inspection. This will especially be true of minor defects which might not be objectional in small quantities, but would be objectional in large quantities. An inspector inspecting each item will not be on the average more than 80 per cent effective. Some good castings will be scrapped and some defective castings will be shipped. Sample inspection or spot checks are usually more effective, and of course more economical.

Controlling Casting Weight

Variation in casting weights is a constant problem. Even with the best foundry practice some variation is expected. This degree of variation can be decreased only by changing the method of molding or casting. By the application of Statistical Quality Control methods, the normal or expected variations can be determined and control limits established for the minimum and maximum castings weights. As long as the weights of the castings remain in control or between the minimum and maximum limits, the process is doing as well as can be expected.

If weights occur outside of the control limits, it is a signal that an investigation should be made and corrective action taken. Considerable potential savings may be realized from control of casting weights. At the risk of excessive emphasis, it must be remembered that control methods in themselves do not create better control or improve quality. They are only a means to an end and not an end in themselves. If the information developed from the control methods is not put to use, nothing can be gained. These methods signal when something is wrong or when the process in doing as well as can be expected.

Effectiveness of a safety or accident prevention program can be evaluated and controlled by proper application of control charts. If there is an increase in the number of accidents which exceeds the upper control limit, we know that some external forces are

at work or that the safety rules and procedures are not being followed as they should. On the other hand if there is a drop in the number of accidents below the lower control limits, the matter should be investigated to determine what factor may have contributed to a better safety situation than should be expected.

Absenteeism may be charted and controlled. Excessive absences may be due to a number of different things but when it exceeds the upper control limit, an investigation is called for. It could be caused by an epidemic of sickness or might be an indication of poor employee morale.

Not to be overlooked is the question of costs. The possibilities for Statistical Quality Control application in this area of cost control is enormous. Instead of spending time tracking down variations in costs which we cannot easily control, this time can be concentrated in the areas where the maximum results can be realized.

Molding Sand Control in Foundry

Sand control should be one of the first applications of Statistical Quality Control to be considered. It is a generally accepted fact that foundry sands vary and an established fact that this variation in the sand should be eliminated. Complete elimination of all variation is impossible but the magnitude of the variations can be reduced and controlled.

The spread of the moisture in molding sand for example will be influenced by a number of factors. The sand in the heap or in the system will not be absolutely uniform. A system sand with proper mulling or mixing facilities should be more uniform than a heap sand cut by hand, but either extreme can be charted and control limits established for moisture.

The quantity of water may be checked by feel or by a moisture control device but the results in any event may be controlled. The spread between the control limits in one case will be much greater than the other but the possibilities of control will be as great in one as the other.

Sand is an excellent subject for Statistical Quality Control. The "buck has been passed' on sand for so long that a good control chart application will find widespread acceptance and approval.

Permeability, grain size, clay content, strength and all the other characteristics of sand which we measure can be charted and control limits established. Foundry sands are likely the most over-controlled factor in most foundries. There is always the constant temptation to add a little more new sand, a little less bond, change the moisture, add wood flour, etc. Most changes are usually prompted by hunches or at best too little factual information. Unless control charts are used the foundryman will most likely resort to the "rat race" of over control in an effort to maintain the sand within some preconceived control limits which the chances are not realistic.

The methods and process must always be studied to see just how well they can be controlled before arbitrary control limits are set.

Tool wear or tool replacement can be controlled. The life of grinding wheels can be charted and controlled. This is a very simple application requiring little time and effort, and can result in the control of a cost factor which may have been disregarded or considered beyond control.

Control limits on flask closing pins and bushings will give timely notice when replacement or repairs should be made in advance of actual difficulties in the castings.

Through the use of Statistical Quality Control Methods many important and critical details may be delegated thus releasing time and thought for creative projects rather than concentration of time and effort on maintenance.

The examples mentioned are by no means the only areas of practical application of Statistical Quality Control. Control charts and control limits can be applied to everything which varies, and all things do vary. The limitation is only because of the lack of imagination. Some applications will not be economical but as one becomes familiar with the techniques the evaluation of their economical value is not difficult.

Statistical Quality Control is not something to consider if you can afford to use—the question is can you afford not to use it. It is a challenge to every foundryman; to investigate these control techniques with the vast possibilities that are available.

DISCUSSION

Chairman: M. T. SELL, Sterling Foundry Co., Wellington, Ohio.

Co-Chairman: M. E. ANNICH, American Brake Shoe Co., Mahwah, N. J.

D. C. EKEY[1] *(Written Discussion):* I would like to congratulate Mr. Martin for his excellent presentation of a most difficult subject. Mr. Martin has unselfishly explained a highly theoretical subject in terms which the operating man can grasp and appreciate. There are several points which I should like to clarify without detracting from the quality of his presentation.

The author indicates that once the concept of variation is grasped the barrier to the use of Statistical Quality Control methods is broken. Knowledge of inherent variation of properties is very important, but I do not feel that such an understanding in itself will sell Statistical Quality Control. One must recognize that the mathematics of this tool can seem most confusing to a great majority of people in plant operations, and they all too frequently seek other simpler and less reliable tools to implement controls of variables which most persons do recognize.

It appears that the important point in breaking the barrier to Statistical Quality Control is to show that the variations can be defined as rather precise functions (i.e., normal, binomial and poisson) and that the quality of decisions based on such variation can be greatly improved by calculation the probabilities associated with a given universe of variation.

Another point of question concerns the author's statement that "Once we can measure variation we will be in a position to control it." We should recognize that here are numerous measurable forms of variation which we do not have sufficient technological understanding to control. In fact many statistical tools (e.g., the analysis of variance) are used by research groups to study the unknown causes of variables which we can measure but not control at the present time.

On this same point one must realize that there are certain sources of variances subject to extreme error (i.e., lack of control) where statistical controls cannot be used. The frequency of the occurrence for a given variable and the stability of its environment are prime factors which determine the suitability of statistical controls.

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1. Technical Director, Lebanon Steel Foundry, Lebanon, Pa.

There is one rather serious point of error in the author's statement "The only time changes should be made is when the total carbon is outside the control limits." The interpretation of statistical controls on this basis seriously limits the value of the tool.

Statisticians use the probability theory of "runs" to show that a variable can be out of control even though no points fall outside the control limits. This method is very important in the anticipation and detection of *trouble* even before points do fall out of control and trouble is encountered. Several points in a row, dependent upon the sigma distance from the mean, can be just as significant an indicator of trouble as one point outside the control limits.

The author's discussion on sampling inspection is very good and one additional comment seems in order. Sampling inspection is not only effective and economical but in the case of destructive testing it is essential. The statement "An inspector inspecting each item will not be, on the average, more than 80 per cent effective", should not be accepted as a general statement. This statistic of 80 per cent will vary tremendously with the type of inspection (e.g., visual, mechanical or chemical) and the type of work being performed.

Mr. Martin recommends the AFS STATISTICAL QUALITY CONTROL FOR FOUNDRIES as an excellent source of information. It is agreed that the AFS text is an excellent reference on the application of statistical control to foundry work, but the theoretical treatment of this reference is inadequate to the extent that one might arrive at erroneous conclusions and/or apply the statistic tools incorrectly.

I would like to recommend two additional texts for anyone who is not thoroughly grounded in Statistical Quality Control. These texts are:

1. *Statistical Quality Control* by E. L. Grant (McGraw-Hill Book Co.)
2. *Quality Control and Industrial Statistics* by A. J. Duncan (Irwin Publishers)

It is sincerely hoped that my observations will be accepted in the light of constructive suggestions, and they are not intended in anyway to detract from the fine work of Mr. Martin.

MR. MARTIN *(Reply to Dr. Ekey):* Dr. Ekey's remarks probably express the sentiments of most foundrymen using Statistical Quality Control. However, this paper was primarily for the majority of the foundrymen who are not familiar with Statistical Quality Control techniques and to a measure not particularly interested.

The results of this paper may have been an over-simplification of a technical subject but the objective has been creating interest and illustrating that results may be obtained without extensive technical training. This by no means is intended to indicate that technical training is not only advisable but essential if complete utilization is made of the various Statistical Quality Control techniques which are being developed in this field daily.

Some variation, it is true, we will not control because of economical reasons or because of lack of technological understanding: but through the application of these Statistical Quality Control techniques we will be in a position to decide which we will attempt to control as opposed to those we will just "live with".

The theory of runs offers closer control but the beginner in Statistical Quality Control needs to learn the primary concepts before advancing to the more complex applications. If the paper has aroused interest in Statistical Quality Control in the mind of the average foundryman it's purpose has been satisfied. This is strictly a "nontechnical" technical paper.

In regard to texts, those suggested by Dr. Ekey are excellent. The International Correspondence Schools of Scranton, Pennsylvania offers an excellent correspondence course based on E. L. Grant's text *Statistical Quality Control*. There are also numerous short courses offered throughout the nation under joint sponsorship of the local sections of the American Society for Quality Control and various universities. These short courses not only give the necessary technical training but also afford an opportunity to discuss practical applications with those confronted with similiar problems.

The author sincerely appreciates the points raised by Dr. Ekey and hopes that more foundrymen may pursue the study of these Statistical Quality Control techniques.

MACHINABILITY OF ALUMINUM DIE CASTINGS

By

Donald L. Colwell and Oldrich Tichy*

ABSTRACT

With the tremendous growth in the use of aluminum die castings, machining has become an important contributor to their success. The common difficulties in the machining of the typical die casting alloy SC84A are considered.

The composition of the typical alloy is given with the reasons for control of each element, with particular reference to the effects of each element on machinability. Emphasis is placed on the increasing use of zinc in aluminum die castings for its improved machinability.

Porosity can cause trouble particularly in areas where holes are to be drilled or tapped. Porosity is always due to trapped air or gases in the die and can be eliminated by proper gating and venting and by judicious use of die lubrication.

Excessive tool wear can come from occluded oxides in the metal and these can be minimized or removed by proper metal melting and by proper fluxing. Sludge inclusions in a finely dispersed form can also cause tool wear.

Hard spots causing tool breakage and nicking may come from several sources, including undissolved silicon, and larger pieces of the oxides or of the sludge mentioned above.

Sludge which can cause excessive tool wear and hard spots is a function of composition and of temperature. Within the composition limits required for successful castability, sludging is most apt to occur when cold metal is charged into the holding pot at the die casting machine. The reasons for this sludge precipitation are given.

Proper melting and metal handling practice is described, with particular reference to the necessity for low temperatures in the holding furnaces of the cold chamber machines.

Machining of die castings, particularly in the drilling and tapping of holes and the facing of certain surfaces, is of much greater importance now that die castings have achieved a prominent place as an engineering material. This discussion is limited to some of the difficulties occurring in machining operations for which the die casting itself is responsible, and methods for the prevention of these difficulties will be presented.

The composition of the aluminum alloy used has a bearing on the machinability of a die casting. The most commonly used alloy contains 9 percent silicon and 3½ percent copper basically, although the copper may be lower and the silicon higher in certain cases. A typical ingot composition, for this alloy ASTM SC84A, is shown on Table 1.

TABLE 1 — TYPICAL COMPOSITION OF SC84A

Mg	Zn	Mn	Si	Cu	Fe	Ni	Others	Al
0.06	0.80	0.40	9.00	3.50	0.90	0.20	0.40	Bal.

Silicon is added for castability, as many of the intricate shapes which are common today could not be cast without it. Unfortunately, silicon adversely affects machinability.

Copper increases hardness and strength and particularly increases yield strength. Copper also improves machinability.

Other elements which have an effect on machinability are iron, magnesium, manganese and zinc. An iron content of about 1 percent and a manganese content of about 0.4 percent are highly desirable from a castability standpoint. They greatly reduce the tendency of the alloy to solder or weld to the die and they improve castability. In these quantities, they tend to improve machinability by providing higher hardness, although if the metal is not handled properly, they have a secondary effect which will be discussed below.

Magnesium in the silicon copper alloys is present only as an impurity and because of its age-hardening effect, should be limited to less than 0.10 percent. It is also undesirable from a casting standpoint, although larger quantities of magnesium would help machinability.

Zinc is now recognized as a desirable alloying element. The greatest beneficial effect of zinc in the alloy is improved machinability, and in many instances, a difficult machining problem has been eliminated by the addition of 2 or 2½ percent zinc to this type of alloy rather than by limiting it to the 1 percent more commonly specified.[1]

Because of the beneficial effects of zinc on machinability, tests of the American Society for Testing Materials now in progress indicate that the corrosion resistance of SC84A is not impaired by higher zinc contents. Exposure of test bars of this alloy with varying zinc contents at three locations, New York City, Kure

*Director of Laboratories and Chief Metallurgist, respectively, Apex Smelting Co., Cleveland.

Beach, N.C., 800 feet from the ocean and Kure Beach, N.C., 80 feet from the ocean, have proved that after one year's time, zinc contents up to 2 percent and higher had no effect.

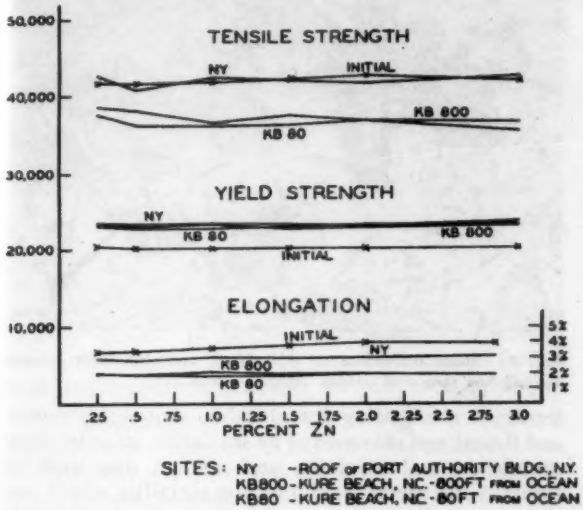


Fig. 1 — Effect of zinc on corrosion of SC84A, 1 year at site (As reported by ASTM Committee B6).

These results are summarized in Fig. 1, where it can be seen that the tensile strength is not affected by outdoor exposure in New York but is reduced by both Kure Beach exposures and by approximately the same amounts with both low and high zinc contents. The same is true of the elongation for all zinc contents, with a slight hardening effect at New York. The yield strength at any of these zinc contents has not been affected by corrosion at these locations.

Serious porosity of die castings has been largely eliminated as a result of the newer types of machine and the increased skill of the die caster in gating and venting. If porosity does occur, however, at an area to be machined, it can cause difficulty with tool wear and with breakage of drills and taps. The latter is particularly troublesome when a lug or boss has to be drilled and tapped in a small size and the drills run into porous areas with resulting bending and breaking.

Proper gating and venting of the die to eliminate trapped air can often clean up troublesome porous areas. Gases other than air can cause the same difficulty, and it seems that gases due to the volatilization of the lubricant used in the die are more often at fault than air. In order to obtain solid castings, a die should be operated with a minimum of lubrication. Castings showing porous areas using radiographic examination, will often be contaminated with oil. Sometimes the amount of oil or lubricant trapped in the casting is so great as to be plainly visable when the casting is sectioned. Such an area is shown in Fig. 2, where the porosity is not excessive but oil exudation is plainly visible at the upper left. In this particular casting, the amount of lubricant used was so great that the surface of the casting was discolored in places.

One of the more frequent troubles with machinability of die castings and strangely enough, the one which is probably the easiest to control, is one due to



Fig. 2 — Seepage of lubricating oil from interior of diecast section. Mag. 25X.

the presence of non-metallic inclusions or "hard spots." These inclusions, when small and finely divided, act as an abrasive and cause excessive tool wear. In the worst cases, sometimes, chunks plainly visible to the naked eye are occluded in the castings and are hard enough to nick and break the cutting tool. The black areas of the micrograph shown in Fig. 3 at 75X are examples of this type of inclusion.

Aluminum, either solid or molten, is covered with an oxide skin, and this skin forms so rapidly in the molten condition, that it builds up into a surface dross. This buildup is faster if the metal is agitated by new charges, by stirring or by dipping out. The oxides are fairly soft when newly formed, but when they collect on the sides of the furnaces or pots and are exposed to high heat, and in many instances, direct flames, they are converted into an extremely hard form of aluminum oxide and are comparable to the fused aluminum oxide abrasive known as "alundum."

Some of these oxides have about the same density

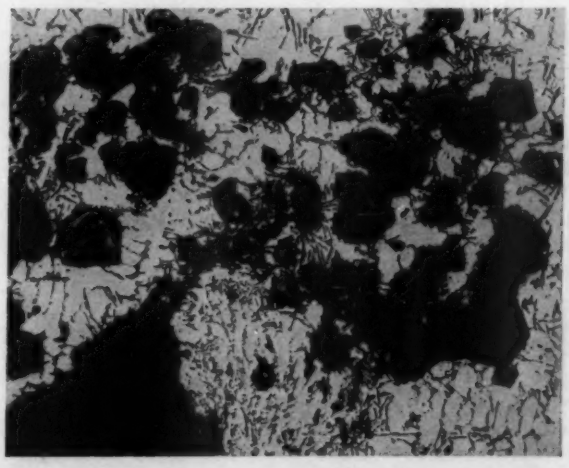


Fig. 3 — Fused aluminum oxide particles (black) and plates of AlFeSiMn complex (gray) in normal cast structure of SC84A. Mag. — 75X.



Fig. 4 — Section of sludge; it settles in layers. Mag. 40X.

as the molten aluminum, and consequently, they are often suspended in the aluminum in more or less finely divided form and are transferred to the casting. The result, therefore, is a soft aluminum casting carrying imbedded particles of an abrasive, and this acts as a grinding wheel in destroying the tool. Furthermore, if a chunk of the fused oxides which has formed on the wall of the furnace or the pot is chipped off and gets into the casting, it can be exposed at a surface to be machined and can nick and break steel, tungsten carbide or diamond tools.

The aluminum oxide type of hard spot can be avoided by good housekeeping. Furnaces, pots, ladles and skimmers must all be kept clean. Slag buildups on the furnace walls and on the sides of the pot must be scraped off periodically and removed. It is poor practice to charge scrap directly into the holding pot, as the large surface area from the scrap has a tendency to make more oxide and thus greatly increases the chances of such oxide inclusion.

Both scrap and ingot should be melted in a preliminary melting furnace and heated to about 1300 to 1400F, depending on the temperature drop in transferring to the holding pot, and then thoroughly mixed and fluxed and skimmed. The skimming must be done completely and carefully, and enough flux used to separate the metal from the non-metallics which are then removed as a dry powder. After the skimming is complete, care should be taken not to dislodge any adhering non-metallics from the sides of the furnace or pot.



Fig. 6 — Hard inclusions of AlFeMnSi complex after proper fluxing has removed oxides. Mag. — 250X.

An entirely different type of "hard spot" or inclusion and one which often causes the same type of difficulty, is the heavy undissolved metallic segregation which settles in the molten aluminum bath as a sludge and is often carried into the casting. In the alloy referred to above, such sludges are mainly hard metallic compounds of iron and aluminum; iron, silicon and aluminum; and iron manganese, silicon and aluminum. A section of such a sludge removed from the bottom of a holding furnace is shown in Fig. 4 magnified 40X. The layers of AlFeMnSi complex are clearly shown and some typical aluminum alloy is occluded in these layers in the lower left hand corner.

If this sludge is carried into the casting, the hard inclusions of the AlFeMnSi compounds are entrapped

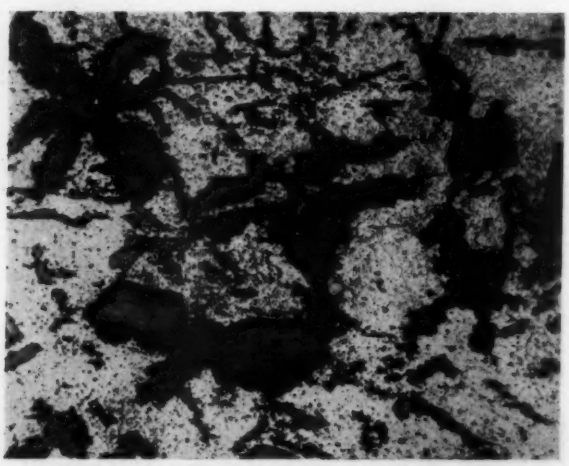


Fig. 5 — Hard Inclusions of AlFeMnSi complex with oxides in casting. Mag. — 250X.

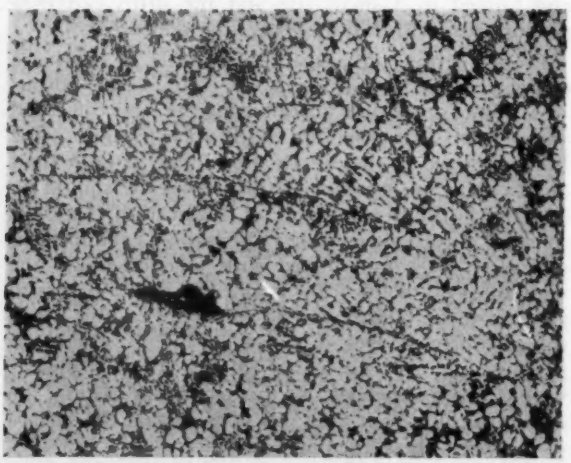


Fig. 7 — Normal SC84A cast structure. Lines are thin oxide layers or "cold shuts." Plates are AlFeSiMn complex. Script formation is AlFeSi complex. Mag. — 150X.

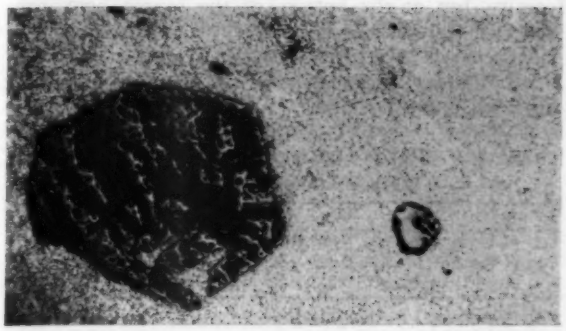


Fig. 8 — Piece of hard complex trapped in casting. Mag.—40X.

as shown in the photomicrographs at 250X in Fig. 5 and Fig. 6, the latter after good fluxing has cleaned out the oxides. It happens that these two examples are of chill-cast sections, but they illustrate the formation and relative sizes of the metallic compounds.

The photomicrograph shown in Fig. 7 is that of a typical die-cast structure of SC84A alloy at 150X. Note the uniform structure and the finely divided constituents present. In contrast to this structure that of Fig. 8 was made at 40X. Attention is directed to the large agglomerate of the undissolved AlFeMnSi complex which was transferred to this casting from an improperly melted bath. Castings from this lot resulted in serious machining difficulties.

When sludge accumulates in the holding pot at the die casting machine, it collects on the bottom as a sandy substance and can be dipped out with a skimmer. Its accumulation is most often due to lack of agitation, too low a temperature or both. Sludge accumulation, therefore, can be prevented by adequate mixing and proper temperature control. Its formation can be explained by reference to an over-simplified consideration of the binary system aluminum-iron. The aluminum end of the binary diagram is illustrated in Fig. 9.

In this diagram, iron content is plotted against temperature, and it can be seen that below about 1.8

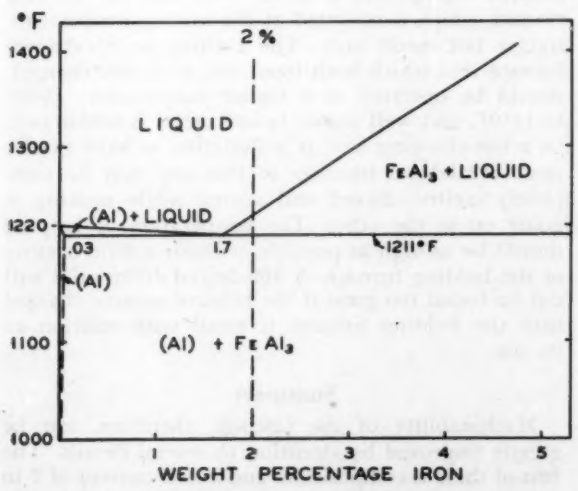


Fig. 9 — Aluminum end of aluminum-iron system.

percent iron and above 1220F, the metal is all liquid. With higher iron percentages, the temperature of initial freezing (or liquidus) rises along the slanting line to the right, and between this liquidus and the eutectic temperature of 1200F, the bath is a mixture of the iron compound $FeAl_3$ and the aluminum bath.

With 5 percent iron, for instance, at 1300F, the bath would consist of solid $FeAl_3$ in the liquid aluminum, and if allowed to stand at this temperature, the $FeAl_3$ would settle and form a sludge. At 2 percent iron, illustrated by the dotted line, the metal is completely liquid above about 1230F, and completely solid below 1211F. Between these two temperatures, solid $FeAl_3$ can form and settle out as a sludge. This explanation is over-simplified as the sludge forming elements are more complex with the presence of silicon, manganese, and sometimes copper, and consequently, the temperature range of sludge formation is wider. The same principle, however, applies.

A frequent practice in the die casting industry often promotes the accumulation of such sludge and is illustrated in Fig. 10. When a molten SC84A alloy is



Fig. 10 — Cold ingot charged in molten aluminum alloy induces sludging.

held at about 1200F and a solid ingot is charged into it, the metal immediately freezes around the ingot and then this and the main body of the ingot slowly melt away. During this process, the temperature range is wide and the number of temperatures infinite, so that all of the high melting point compounds observed in previous figures are freed at sometime during the cycle, and consequently settle to the bottom of the pot as a sludge.

The temperature of the holding pot is never raised high enough nor is sufficient time allowed for complete re-solution to be effected. Preheating of the ingot to 400 or 500F would have little effect on this phenomenon. The only preheating which would help is preheating to a temperature high enough to have all constituents completely dissolved, and this should be done in the melting furnace.

An example of aluminum and iron pot alloying is illustrated in Fig. 11, a section of aluminum knocked off of the interior of an iron pot. The iron-rich alloy is at the right, and a plate of $FeAl_3$ is apparent in the left layer of aluminum alloy at the center of the photo. Neither of these will easily dissolve in aluminum at the usual temperatures of the holding pot, and if carried into a casting would cause troubles. Normally,

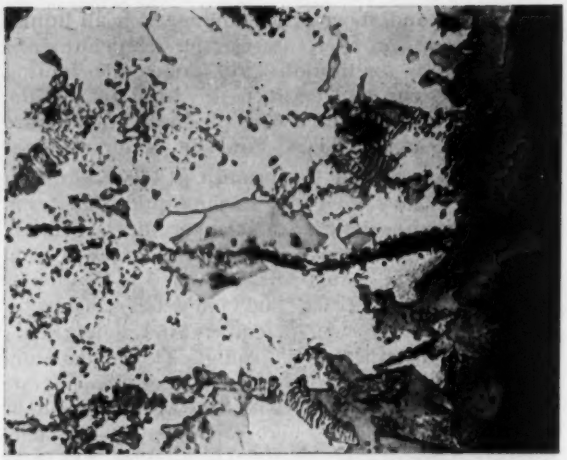


Fig. 11 — Aluminum and AlFe alloy formed inside iron pot. Center plate is AlFeSi complex, iron-rich alloy at right. Mag.— 200X.

they will settle to the bottom of the pot and accumulate as sludge.

Some illustrations of segregation for sludging of an aluminum bath at certain temperatures will be interesting. In one die casting operation, ingot of the composition shown in Table 1 was charged into a breakdown furnace which was operated at 1250F. Scrap was charged into a breakdown furnace along with this ingot and the surface of the metal was skimmed occasionally but no attempt was made to stir regularly. After several hours' operation, analyses of samples from the top of the bath and from the bottom of the bath differed considerably, as shown in Table 2 (A). The top composition, therefore, had been transferred to the holding furnace at the machine and the accumulation of the heavy elements in the bottom of the furnace continued.

The practice was then modified so that the furnace was operated at 1300F, the metal thoroughly mixed and fluxed with a neutral flux and skimmed. Samples from the top and bottom of the bath analyzed as before gave excellent uniformity as shown in Table 2 (B). This metal, when charged to the holding furnace operated at a lower temperature, did not segregate nor sludge even after days of continuous operation.

TABLE 2 — CHEMICAL ANALYSES

	Mg	Zn	Mn	Si	Cu	Fe	Ni
		A					
Top of Bath.....	0.20	0.96	0.35	9.35	3.45	0.95	0.19
Bottom of Bath.....	0.07	0.61	2.0	10.10	2.96	5.0	0.64
		B					
Top of Bath.....	0.04	0.87	0.32	8.75	3.38	0.90	0.22
Bottom of Bath.....	0.04	0.83	0.33	8.50	3.33	0.87	0.23

Segregation of AlFeSiMn complex can be prevented by adequate mixing at adequate temperature.

In another instance, metal was charged into a breakdown furnace operated at about 1350F. The charge consisted mostly of ingot with a few gates, and immediately after reaching 1350F temperature, the metal was skimmed and poured to a heel and another charge melted. Samples taken from the top of the first heat and the top of the fourth and last heat showed uniformity, but at the bottom of the fourth heat, considerable segregation of the heavy element complexes had occurred. The three analyses are given in Table 3 and attention is called to the increase in silicon, iron and manganese.

TABLE 3 — SEGREGATION AT ADEQUATE TEMPERATURE WITH INADEQUATE MIXING

Sample	Mg	Zn	Mn	Si	Cu	Fe	Ni
Top of 1st Heat.....	0.09	0.33	0.33	5.79	3.51	0.34	0.13
Top of last heat.....	0.09	0.30	0.31	5.79	3.45	0.44	0.14
Bottom of Last Heat...	0.07	0.20	1.44	6.69	2.91	3.18	0.02

In this instance, the temperature was satisfactory, but mixing was insufficient. Proper agitation and cleaning would have prevented this segregation. In still another instance, one furnace full of ingot was melted and maintained at 1380F all day. When no mechanical stirring was done, there again was a gradual increase in silicon, iron and manganese at the bottom of the bath. When the metal was stirred and cleaned regularly, similar analyses of samples from the beginning, middle and end of the day showed excellent uniformity. The two sets of analyses are shown in Table 4.

TABLE 4 — SEGREGATION CAN BE PREVENTED BY ADEQUATE MIXING AT ADEQUATE TEMPERATURE

Samples	Mg	Zn	Mn	Si	Cu	Fe	Ni
Poor (No Stirring)							
Beginning	0.03	1.26	0.33	2.57	6.79	0.88	0.22
Middle	0.03	1.28	0.30	2.52	6.89	1.02	0.21
End	0.03	1.20	0.57	3.04	6.42	2.02	0.25
Good (Stirred Well)							
Beginning	0.07	1.40	0.31	2.43	6.80	1.00	0.31
Middle	0.08	1.36	0.32	2.57	6.99	1.02	0.30
End	0.08	1.36	0.39	2.62	6.99	0.99	0.31

In the first instance, even though the temperature was sufficiently high, the lack of agitation resulted in incomplete solution and allowed the higher melting point complexes to settle to the bottom as sludge. In the second instance, proper mixing resulted in a homogeneous bath.

In order to eliminate the type of hard spots due to metallic segregation, it is essential that the holding furnace which is operated at the casting temperature receive hot metal only. The melting or breakdown furnace into which both ingot and scrap are charged, should be operated at a higher temperature, 1300F to 1350F, and well mixed before metal is withdrawn. In a hot charging line, it is desirable to have two or more breakdown furnaces so that one may be completely melted, fluxed and stirred while melting is going on in the other. The temperature of tapping should be as high as possible, without undue heating of the holding furnace. A 100-degree differential will not be found too great if the relative volume charged into the holding furnace is small with relation to its size.

Summary

Machinability of die castings, therefore, can be greatly improved by attention to several details. The first of these is composition, and a zinc content of 2 to 2.5 percent in SC84A is suggested. Porosity due to

entrapped air can be minimized by proper gating and venting, but porosity due to excess lubricant in the die is more commonly found and can lead to trouble in machining. Die lubrication, therefore, should be kept to an absolute minimum and the use of lubricants with suitable solids in suspension is usually more satisfactory than the indiscriminate application of oil over the face of the die.

Oxides and other non-metallic inclusions can be eliminated by good housekeeping and by thorough fluxing and mixing operations in the melting furnace. In certain cases, it might be desirable to flux the holding furnace at the machine occasionally. If the metal turnover is small, there can be an oxide buildup during the holding operation.

Metallic inclusions due to sludging or segregation can be eliminated by avoiding the charging of cold metal into the holding furnace and by operation of the melting furnace at a sufficiently high temperature and with thorough mixing and stirring.

If all of these precautions are observed, castings suitable for machining can be consistently produced. The tooling used in machining is another problem outside of the scope of this presentation. It has been discussed previously.[4]

References

1. D. L. Colwell, "Zinc in Aluminum Casting Alloys," AFS TRANSACTIONS, vol. 60, p. 517 (1952).

2. ASTM Report of Committee B-6, 1955. Tables 2, 3, 4.

3. ASM *Metals Handbook*, 1948 edition, p. 1161.

4. J. J. Stobie, Jr., Machining Cast Aluminum Alloys," *Modern Metals*, Nov. 1947.

DISCUSSION

Chairman: B. L. MEREDITH, Federated Metals Div., American Smelting & Refining Co., South Plainfield, N.J.

Co-Chairman: W. BABINGTON, Bell Telephone Laboratory, Murray Hill, N.J.

H. E. TRIMBLE[1]: Does iron-manganese pickup cause the hard spots or is it oxide, or both?

MR. COLWELL: Hard spots are caused both by an iron-silicon-manganese compound or other metallic compound and by oxides. There is no pick up of iron and manganese. Iron, manganese and other elements are caused to settle out by improper temperatures and improper stirring and can be prevented by the recommended procedure of separate melting and adequate temperatures and hot charging.

W. A. ECKHART[2]: What is the minimum requirement of zinc content in SC84A alloy to effect noticeable improvement in machinability?

MR. COLWELL: There is no hard and fast minimum for a zinc content. Improvement in machinability becomes noticeable at about 1 per cent and at about 2 per cent or more good machinability is almost guaranteed if the foreign inclusions described are not present.

P. R. BABYAK[3]: In induction melting furnaces, is the inherent stirring sufficient to prevent accumulation of sludge?

MR. COLWELL: The induction melting furnace will prevent the accumulation of sludge insofar as the agitation factor is concerned. We try to point out however, that this is also a function of temperature and that improper operation of an induction furnace along the lines shown in Fig. 10 can be just as harmful as in any other kind of furnace. Temperatures must be adequate, chilling must not take place by cold charging, and only after this is done does stirring become important.

FRANK TURK[4]: In fluxing operations to separate oxides, is the flux stirred throughout the molten metal or merely rabbled in the surface of the melt?

MR. COLWELL: Mr. Turk asked about stirring the flux throughout the metal, and, in this we are firmly of the opinion that this is desirable provided the stirring is not vigorous enough to cause undue agitation. This is contrary to certain statements that have been made previously, but in our opinion it is illogical to expect fluxes to clean metal if they are not put in contact with the metal itself. It is desirable to avoid excessive agitation however as that increases oxidation.

CHAIRMAN MEREDITH: For separation of oxides in a melt, do you recommend a holding time after fluxing; if so, how long?

MR. COLWELL: We recommend a holding time after this fluxing operation of approximately 5 to 10 minutes for the best cleaning. This, of course, can be reduced if the fluxing action required is only mild.

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1. Torrington Co., South Bend Div., South Bend, Ind.
2. Power Products Co., Grafton, Wis.
3. Westinghouse Electric Co., Cleveland.
4. Joy Mfg. Co., New Philadelphia, Ohio.

MELTING MALLEABLE IRON WITH PULVERIZED COAL AND OIL AS FUEL

By

E. H. Nielsen*

ABSTRACT

The purpose of the combined application of pulverized coal and oil as a fuel was to reduce oxidation during the melting down, heating up and tapping of the metal.

The apparatus used was a batch melting reverberatory furnace; pulverized coal and oil equipment was installed on duplexing malleable iron furnace. This furnace was fired long enough to find that the combined fuel will give better control of oxidation than firing with coal or oil separately. (This duplexing furnace has not yet been in continuous production due to lack of tonnage where it is installed.)

The batch melting furnace has been in operation for a period of one year. Tables show oxidation before combined fuel was applied, and after the application of combined pulverized coal and oil.

Conclusions show the savings in lower oxidation, higher fluidity, more uniform metal and lower cost of charging material.

In melting iron in a reverberatory furnace, or in any type furnace, using oil as a fuel there is a tendency toward oxidation. Prior to October, 1919 when the first malleable iron was melted successfully with pulverized coal, practically all malleable iron was melted with handfired furnaces. Melting loss by oxidation was high and temperature was low compared with temperatures required today, using a charge of 50 to 60 per cent pig iron and the rest re-melt (sprue).

Some attempts were made in the early twenties to melt malleable iron with oil as a fuel. The results were little better than in handfired furnaces insofar as oxidation loss of elements were concerned. However, they obtained some higher temperature when tapping over handfired furnaces – no more than the brick would allow. At that time a temperature of 2700 F at the tapping spout was considered hot.

The success and greatest benefit of pulverized coal-fired reverberatory melting furnaces was control of the fuel and air; the fuel being carbon, it produced controllable atmosphere that created the high temperature and decreased oxidation to a normal point that helped increase fluidity. This same controlled atmosphere that produced high temperature and lower

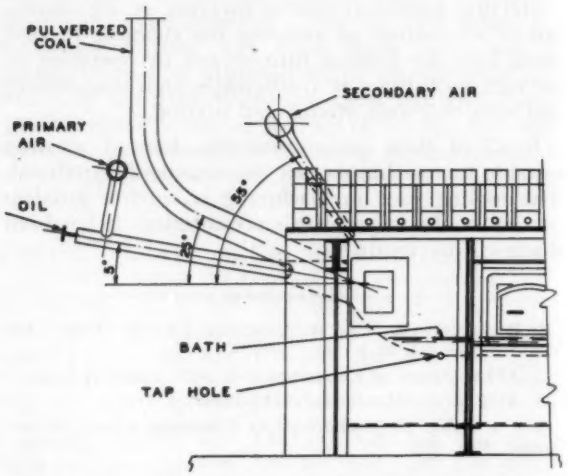


Fig. 1A — Showing side view of batch-type malleable melting furnace, front end showing slope of burners, both oil and coal ducts, and also top blast (secondary air).

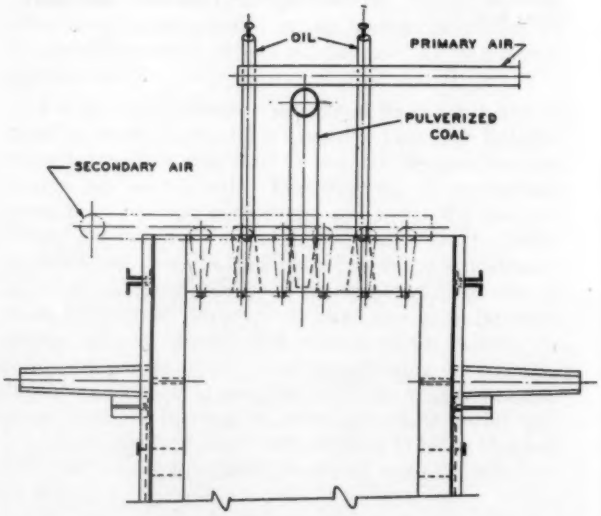


Fig. 1B — Top view of same furnace as in Fig. 1A. Note the six outlet nozzles equally spaced from the manifold secondary air blast pipe. Also, note oil burners inserted parallel to each other.

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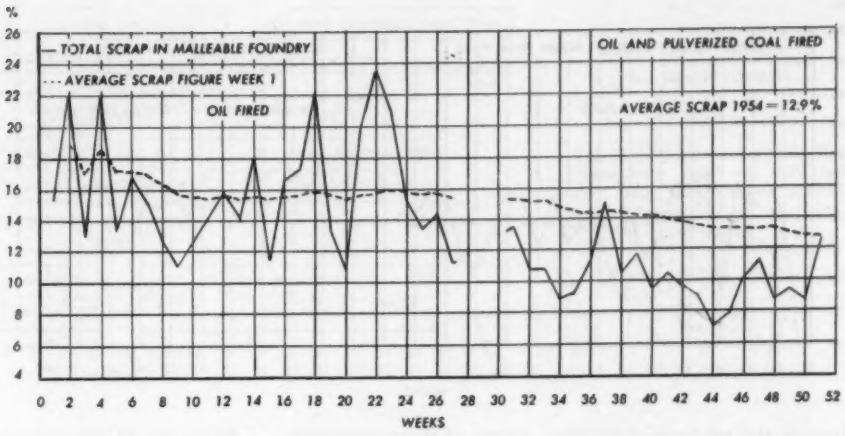


Fig. 2 — Total and average scrap figures for first 6 months of 1954 when firing with oil only. On the right is total and average scrap figures for the last 6 months of 1954 when melting with combined fuel of oil and coal.

oxidation proved successful in melting material for chill and grain rolls in roll foundries.

In 1954 the writer was in Sweden to check on oil-burning equipment to be installed on a duplex furnace. In this foundry they were melting 10 to 12 tons of malleable iron in a reverberatory furnace fired with oil. (A few years previous to that time it was hand-fired.) The molten iron from this oil-fired furnace was not satisfactory, and the scrap was high. The carbon content of the iron was 2.90 to 3.00 per cent; silicon was below 1 per cent, which is necessary for this high carbon. The temperature was low; if fired much longer to get higher temperature, there would not have been any elements left.

To overcome high oxidation they were using not over 40 per cent sprue; the balance was standard pig iron and some special pig iron that increased cost of the charge.

This foundry for some time had been considering combining pulverized coal and oil to get better iron at lower cost, reducing oxidation, increasing temperature, and using more sprue and eliminating use of special pig iron.

Before we could apply the pulverized coal combined with oil in that same furnace some changes had to be made. One change was to enlarge the underground flue from the furnace to the stack which was located outside of the building. This flue was approximately 75 ft long. There is no doubt that if we had enlarged the flue only and continued to melt with oil we would have improved melting conditions. (The reason for mentioning this is that through the many years of rebuilding furnaces for batch melting iron and rolls that the size of a flue, or stack, has always met with resistance to the idea of enlarging flues and stacks, and there was no different opinion in Sweden. Their engineers considered it unnecessary to enlarge the flue. However, the flue was enlarged and it resulted in better operating conditions.)

When building a reverberatory furnace for melting malleable iron or metal for rolls it has been proven that it is better to have too large a stack, and tunnels leading to the stack, for good operating conditions than it is to have a stack that is too small. An inadequate stack is a handicap in many ways.

The original oil installation consisted of two low-pressure luminous flame oil burners set on an angle of five degrees. Height of the burners was such that the flame impinged upon the top of the molten metal in line with the taphole. The taphole was located close to the front bridge wall.

Part of the secondary air went through the oil burner; the remainder came from a wind bung. This wind bung is similar to the one used on the hand-fired furnaces. The only difference was that the nozzles were enlarged to get additional secondary air to burn more fuel, both coal and oil, to increase melting capacity in less time. This was made possible by increasing the size of the tunnel from the furnace to the stack to handle larger volume of gases and reduce back pressure on the furnace allowing high temperature flame to flow uniformly through the furnace.

Application of pulverized coal together with oil was as follows: We left the oil burners and the secondary air from the wind bung as originally, except the six nozzles were enlarged. We used pulverizer equipment with a capacity of 250 to 300 lb of coal per hour at 2150 rpm.

We used a high and low velocity burner, a type commonly used in malleable foundries.

After a few trial runs we found it necessary to water-cool the nozzles. This was done to eliminate fire in the coal burner nozzle when firing at low fuel capacity. We used both the high and low chambers wide open.

The pulverized coal burner was located between the two oil burners. The centerline of the pulverized coal burner nozzle was 4 in. higher than the center-

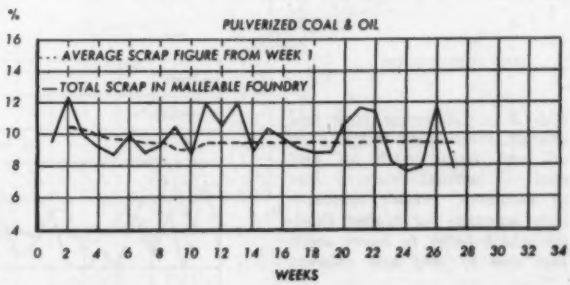


Fig. 3 — Showing first 6 months of 1955 total and average scrap figure. There is marked improvement for more and better castings.

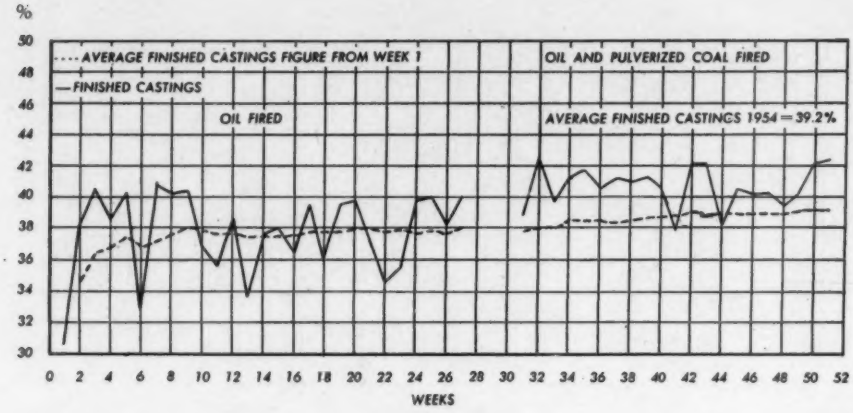


Fig. 4 — Finished castings and average finished castings figures for the first 6 months of 1954 melting with oil, and the last 6 months of 1954 melting with coal and oil. The carbon is up indicating higher yield of good castings.

line of the oil burner nozzles. Slope of the pulverized coal burner was 25 degrees so that the stream of the pulverized coal passed through the oil stream impinging upon the bath at the same point as the oil burner.

Figure 1 shows a side view of batch-type melting furnace, front end showing slope of burners, both oil and coal ducts, and also top blast or secondary air. See also Fig. 1B.

The top blast pipe is 6 in. in diameter extending across the top width of the furnace. From this 6-in. manifold secondary air blast pipe there are six outlet nozzles equally spaced on furnace with set at an angle of 55 degrees and extending through the brick bung. The combined area of the six is less than the area of the 6-in. manifold pipe. The reason for careful calculation of top air nozzle area is to get uniform volume of air through each nozzle across the width of the furnace.

Note location of tapholes in Fig. 1A. (There is one on each side.) Also, note location of front bridge wall in relation to taphole. For good melting results the hottest spot in the furnace should be where the taphole is, rather the taphole should be located where the hottest flame temperature is in the furnace. In this application using a top blast the highest flame temperature is close to the front bridge wall, where the taphole is located.

The oil burners are set at an angle of five degrees; the pulverized coal, 25 degrees; and the top blast secondary air nozzles, 55 degrees. The three different angles of fuel and air invert into each other above the bath in front of the taphole. This produces a uniform mixture of fuel and air creating a high temperature flame as it impinges on the bath of molten metal where the taphole is located.

The slope of the bottom from the back bridgewall to the taphole is $\frac{3}{16}$ in. per foot. Height of back bridgewall was the same as in good practice when batch melting only with pulverized coal as fuel (4 to 6 in. above bath of molten iron).

Fuel required per hour to melt 12 to 15 tons of malleable iron, chill roll or grain roll heats in a batch melting reverberatory furnace was approximately 1650 lb of good metallurgical coal having a heat value of 14,500 Btu. We used 85 per cent oil and 15 per cent coal combined (equal to 24 million Btu) to accomplish the reduction of oxidation during the melting down, heating up and tapping of the metal.

Normal oxidation loss in a malleable batch melting furnace from start of heat to tapping time is 45 to 50 points of carbon, 20 to 25 points of Si and 15 to 20 points Mn. It is necessary to calculate the cold charge properly, weigh the charge correctly and fire the furnace carefully. This should result in uniform oxidation loss, high fluidity, proper temperature, and a high yield of good castings when firing with pulverized coal. Firing with oil and pulverized coal we obtained the same good melting results, normal oxidation, and hot iron.

Figure 2 shows total and average scrap figure the first six months of 1954 when melting with oil only. On the right is total and average scrap figure for the last six months of 1954 when melting with combined fuel oil and coal.

Figure 3 shows first six months of 1955 total and average scrap figure. There is marked improvement for more and better castings when fired with combined

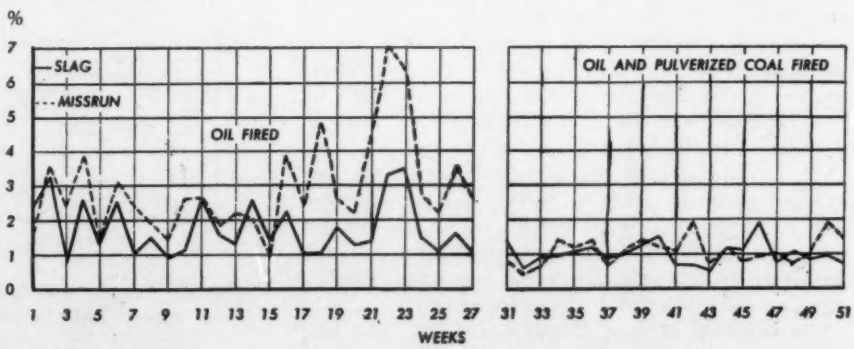


Fig. 5 — Showing slag and misrun scrap loss firing with oil and firing with oil and coal — normal uniform loss of elements; proper temperature accounts for higher fluidity which helps to lower scrap loss due to slag and misrun.

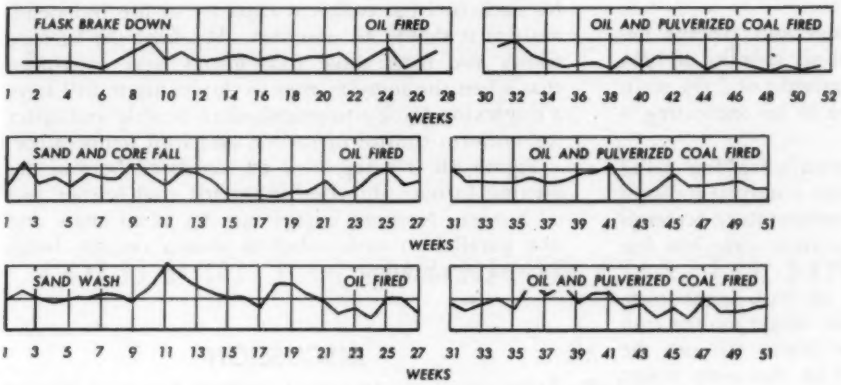


Fig. 6 — Scrap reports on flask breakdown, sand and core fall, and sand wash. Good molten iron is important to keep scrap loss down.

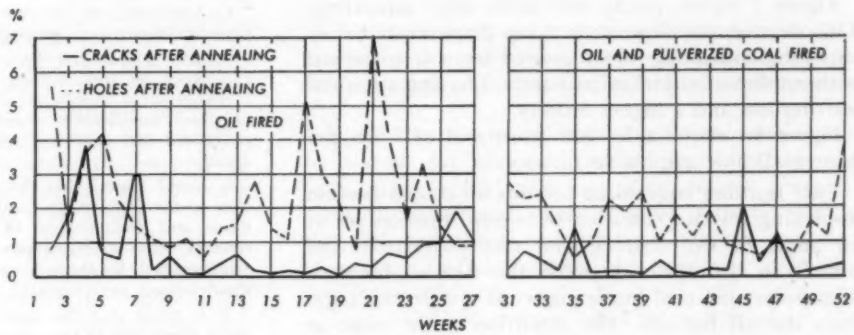


Fig. 7 — Cracks and holes after annealing. This chart shows improvement when castings were poured from iron having a normal melting loss and higher fluidity.

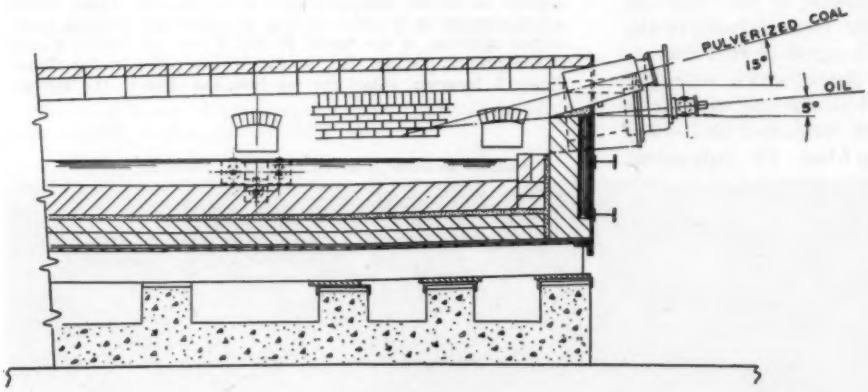


Fig. 8A — Showing side view front end of 5-ton hourly rated malleable duplexing furnace showing slope of burners, both oil and coal. Note location of taphole in relation to front wall when not using top blast secondary air.

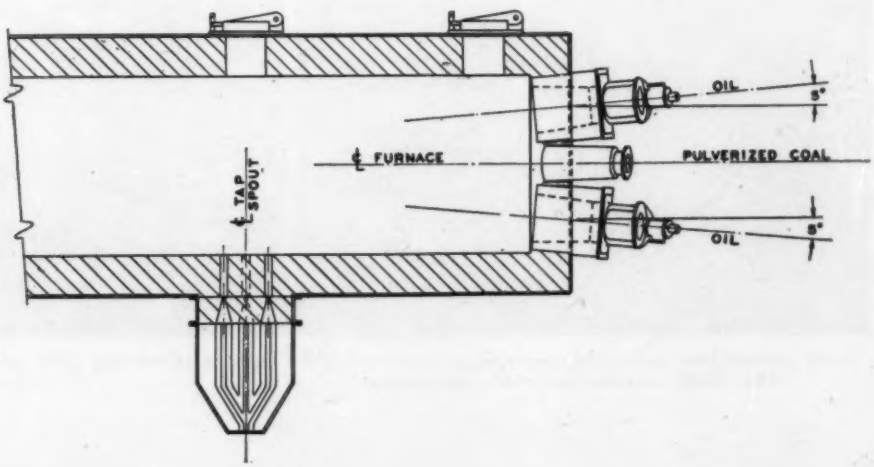


Fig. 8B — Top view of same ventilated bottom duplexing furnace showing pulverized coal burner and oil burner. Note oil burners are set at an angle and not parallel to each other as shown on the batch melting application.

fuel, oil and pulverized coal.

Figure 4 shows finished castings and average finished castings figures for the first six months of 1954 firing with oil; and the last six months of 1954 melting with oil and coal the curve is up indicating a higher yield of good castings.

Figure 5 shows slag and misrun scrap loss firing with oil and firing with oil and coal – normal uniform melting loss of elements; proper temperature accounts for a higher fluidity that helps to lower scrap loss due to slag and misrun.

Figure 6 shows scrap reports on flask breakdown, sand and core fall, and sand wash. Good molten iron is important to keep scrap loss down, but not the answer to all problems, as shown on this scrap report chart.

Figure 7 shows cracks and holes after annealing. This chart shows decrease in scrap from cracked castings after annealing when poured from iron melted with combination fuel of coal and oil having a normal melting loss and a higher fluidity.

Figure 8A shows side view front end of 5-ton per hour malleable duplexing furnace.

Fuel burning equipment for this air-cooled bottom duplexing furnace consisted of two oil burners set at an angle of five degrees, and one pulverized coal burner in the center between the two oil burners. The pulverized coal burner is set at a different angle than the oil burner. The intention is the same as the application of oil and coal on the batch melting furnace. The two different angles of fuel and air invert into each other above the bath in front of the taphole producing a uniform mixture of fuel and air creating a high temperature flame which impinges on the bath of molten metal where the taphole is located. On this application of combined fuel – coal and oil – we eliminated the top blast. The equipment for each fuel has sufficient capacity to fire independently if it should be necessary. We fired the furnace empty and from those trial firings were convinced that when the foundry goes to duplexing it will have a duplexing firing equipment more flexible and faster for uniform control of carbon and iron temperature.

Figure 8B is a top view of ventilated bottom duplexing furnace showing pulverized coal burner and oil burner. Note oil burners are set at an angle and not parallel to each other as shown on the batch melting application.

DISCUSSION

Chairman: JAMES THOMSON, Continental Foundry & Machine Div., Blaw-Knox Co., East Chicago, Ind.

Co-Chairman: H. C. WEIMER, Beardsley-Piper Div., Pettibone-Mulliken Corp., Chicago, Ill.

MILTON TILLEY[1] *(Written Discussion):* Not having extended experience on cold melting malleable iron in an air furnace I cannot speak authoritatively on detailed operation, but having had considerable experience on duplexing, using both pulverized coal and oil, I may be able to contribute some comments.

It would appear that the author has arrived at a good solution to the problem at hand. Oil has the advantage of hotter flame and disadvantages of 100 per cent higher cost in this country, more oxidizing products of combustion and more difficult control. By the mixing with pulverized coal some of these disadvantages are alleviated, particularly the oxidation.

It is noted that the angle of the oil burner is set at 5 degrees which checks our experience with either oil or coal. The angle of 25 degrees for cold melt and 15 degrees for duplexing appears to be for adequate mixing of the two fuels, which would seem to be in order so long as it does not produce a too violent agitation of the batch. In Fig. 8 the tap hole is placed at the proper position, i.e., at the hottest point in the flame. Figure 1, however, shows the tap hole too close to the burner.

1. National Malleable & Steel Castings Co., Cleveland.

COLD PROCESS FOR RESIN COATED FOUNDRY SANDS

By

J. E. Bolt*

The newest and fastest growing technique today in the shell molding field is in the resin coating of shell molding sands at room temperature. Resin coated sand, as such, has been recognized for some time as a notable improvement over the conventional dry sand-resin mixture from the aspects of resin economy, less dust, and the ability to blow shells and cores without resin segregation.

Widespread concentrated activity is currently underway by resin producers, equipment manufacturers, and progressive foundrymen throughout the country to further improve sand coating resins and the methods for their application. The objective is to make possible a simple, rapid, and thorough processing of shell molding sand by the foundryman in his own shop, using existing equipment to the maximum, to produce resin-coated sands that make for a cleaner and more economical foundry operation.

With the recent advent of suitable resins and techniques for the cold-coating process, there has been much speculation regarding future potential usage of coated sands in the foundry industry. Within the next few years, resin-coated sands are expected to replace almost entirely the presently used dry sand-resin mixtures.[1]

The trend is toward blowing rather than the dumping of shell molds, cores, and stack molding segments, all requiring coated sands when blown. Complete adoption of resin-coated sands by a foundry could allow for a single centralized sand coating station with pneumatic conveyance of the resin coated sand to the various shell and core blowing stations without danger of resin segregation.

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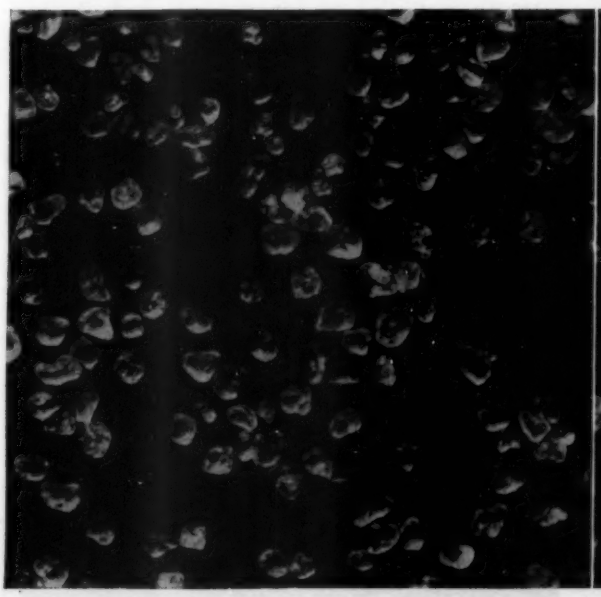


Fig. 1 — Round grain silica sand AFS 120 before coating. Mag. — 25x.

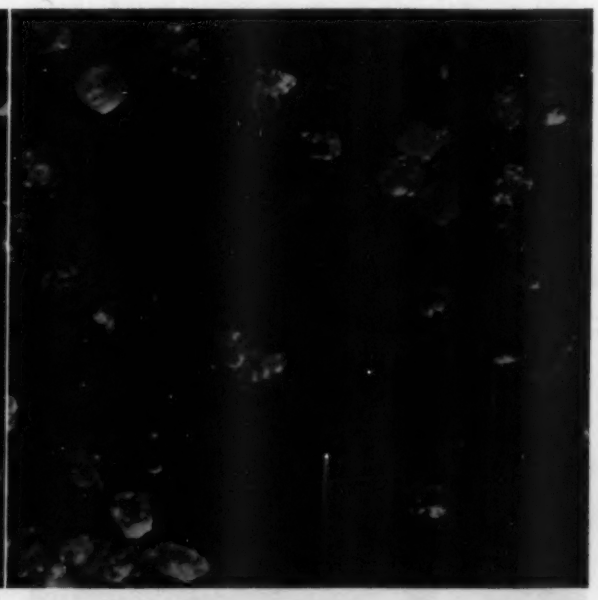


Fig. 2 — Resin-coated round grain silica sand showing glossy appearance. Note agglomerates. Mag. — 25x.

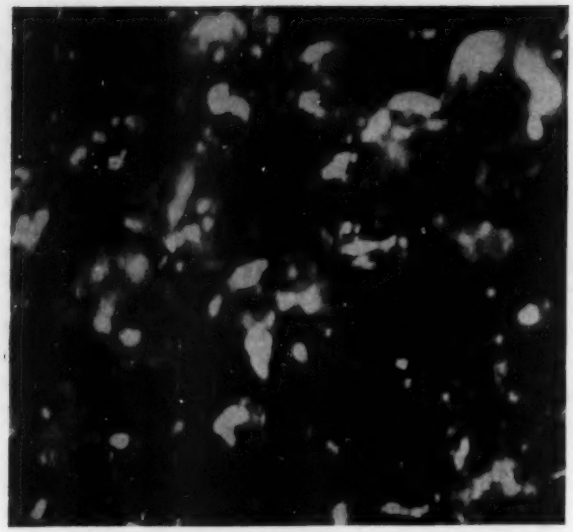


Fig. 3 — Section of shell mold made from resin coated round grain silica sand. Mag. — 25x.



Fig. 4 — Sub-angular sand AFS 110. Mag. — 25x.

Figures 1, 2, and 3 illustrate respectively uncoated washed round-grain silica sand, resin coated sand, and a section of coated sand when cured. Compare these with Figs. 4, 5, and 6 which show respectively uncoated sand, sand and powdered resin mixture, and a section of the cured sand-resin mixture. In the mixture, Fig. 5, the small light particles are resin and the larger particles are sand grains. A dark Juniata sand of approximately AFS 110 fineness was used for color contrast with particle size of the resin in the order of 300 mesh and smaller. The resin coated sand, Fig. 2, shows little difference, except for a glossy appearance, as compared to the satin finish of the uncoated round grain washed silica of Fig. 1. In both cured sections, Figs. 3 and 6, the individual grains exhibit a glossy curing.

Phenomenal growth potential is seen in the use of blown shell cores for conventional green sand practice. With currently available equipment, resins, and processes, cores may be blown directly into heated boxes, with cure effected and cores completed, ready for use in a matter of seconds. Resultant cores possess all the advantages of the shell molding process, such as surface smoothness and dimensional accuracy, plus greatly increased venting in the hollow cores. The need for core ovens and driers is eliminated in such operations.

While cost comparison with cores made by conventional practice is necessarily based on individual

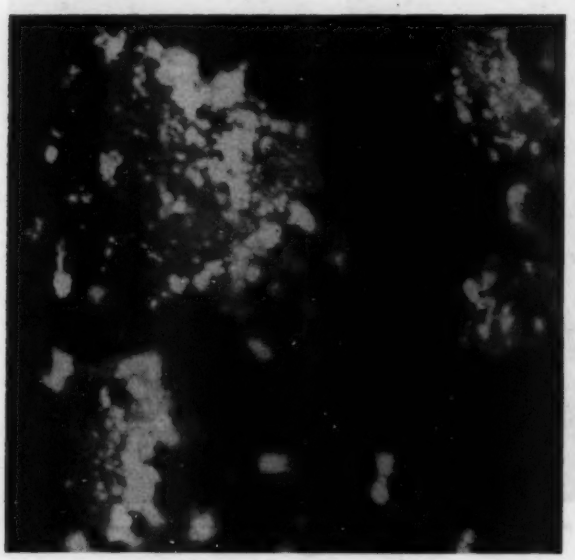


Fig. 5 — Dry sand-resin mixture with small resin particles adhering to sand grains. Mag. — 25x.

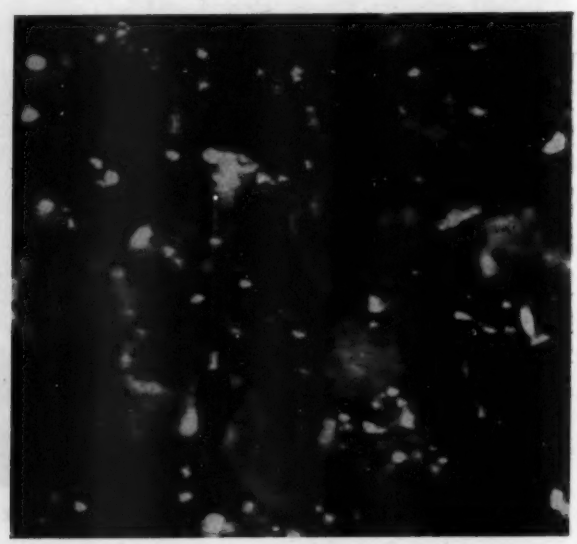


Fig. 6 — Section of shell mold made from sand-resin mixture. Mag. — 25x.

applications, production reports indicate that suitable shell cores may be produced at a cost approximately one-third the amount of conventional cores.[2] Considering the excellent quality and the potential cost savings possible through the use of blown shell cores made of resin-coated sands, it is very likely that in the near future resin consumption for shell cores, both for shell and green sand molding applications, will exceed that for shell molds.

At the time of this writing, the cold coating of foundry sands by phenolic resins is in its infancy, with materials, equipment, and technique improvements being made constantly and rapidly. Hence, any write-up will be necessarily in the form of a status report.

Cold coating of sands at room temperature by phenolic resins may be accomplished by one of two methods. A liquid resin is added to and distributed over dry sand particles; or a dry powdered resin is first mixed with dry sand then dispersed by the use of a solvent. Adequate mixing and drying must be accomplished. With either method this is done without further advancement or reaction of the resin.

Liquid resins are normally one stage, needing no catalyst addition. After adding the resin to the sand, mulling is continued until the mixture becomes dry and free-flowing.

The greatest usage at present appears to be in the use of solid resins, which may consist of ground one-, or two-stage resins. Normally, with the latter the resin (both novolak* and hexa**) is added to the sand and distributed by a short mull. A solvent composed of alcohol and water is then added to the mixture which immediately converts the resin into a viscous fluid. Further mulling distributes the liquified resin over the sand grains, coating them. Following this, the solvent is made to evaporate leaving a free-flowing dry resin coated sand.

Phenol-formaldehyde resin is readily soluble in alcohol and either denatured ethyl or isopropyl alcohol may be used. Methyl alcohol is not suggested for use because of its toxic properties. Water added to the solvent serves two purposes. It controls the rate of solvent evaporation and effects dispersion of the hexa equal to that of the resin. The solubility of hexa is approximately 10 times greater in water than in denatured alcohol. Ratios of alcohol to water currently in use vary from 2.5/1 to 3.5/1 with 3/1 by weight being the most common.

This technique of using an alcohol and water solvent system, along with more recently developed sand coating resins, has been largely accountable for the great progress made in the cold coating process recently. Earlier trials of coating sand with a solution of resin in alcohol were not fully successful.

The amount of resin to be used is governed by the application and the sand being coated. Once the percentage of resin is determined, the amount of solvent required may be expressed as a ratio of resin to solvent by weight. Normally these resin solvent ratios range from around 3/1 to 5/1 and are dependent upon several factors including:

1. **Grain size:** Small grains (particularly those retained on screen 200 and finer) require not only more resin but more solvent (a lower resin/solvent ratio) than do large grains. Some of the solvent obviously goes initially to wetting the sand grains as well as the resin.
2. **Percentage resin:** Low resin content (below 3 per cent) requires an increased proportion of solvent (a lower resin/solvent ratio) in order to wet the mixture. Using higher percentages of resin (over 5 per cent) a lower proportion of solvent (a higher resin/solvent ratio) is permissible because the basic requirement for wetting the sand grains is met at a lower solvent quantity level.

A good indication of proper solvent quantity is shown by the buildup of a good green strength or plasticity of the mixture during mulling plus the maintenance of this plasticity for at least two minutes. Excess solvent prolongs this period of plasticity, but also prolongs the drying cycle as well.

The sand to be coated has a tremendous effect on the amount of resin to be used and on performance of the resultant product. While round grain sands, because of their low surface area to volume relationship, permit high tensile strength values with resin economies, they are not always best in application performance, usually because of a lack of resistance to thermal cracking when made into a shell mold or shelf core. For this reason, in many cases, a sub-angular grain sand is required with a good screen distribution using a higher percentage resin.

Another consideration is the screen size or AFS number of the sand. While we usually think of shell molding applications at grain finenesses of AFS 100 or finer, many core applications are permissible using sands of grain fineness in the order of AFS 60 to 100. Certainly larger grains will reduce the amount of resin required, and when compared to the conventional cores they replace, the blown shell cores will, in most cases, perform an equally satisfactory or a superior function in producing a smooth, accurate, internal casting surface.

The choice of grain size for shell molds must be made regarding resin economy versus the desired surface finish of the casting. While premium priced washed and dried silica sands are preferable from the aspect of resin economy per se, many shops are finding it worthwhile to investigate local sands. If they are not too high in extraneous matter and clay content, it is possible that suitable coating techniques may be worked out. The choice of sand then must be based on a cost and performance comparison of the local sand plus usually increased

*Novolak is the thermoplastic component of a two-stage resin to which the **hexamethylene-tetramine is added. When heated, the "hexa" breaks down into ammonia and formaldehyde, thus effecting cure of the resin.

resin content versus the cost of premium sands plus a lesser percentage of resin.

Either decreasing the particle size of the sand or increasing the fine particle additives will require more resin. It has been observed that sand high in fines, particularly those having particle sizes retained on screens 270 and pan tend to cause lumps which are still present at the end of the mulling cycle. It is believed that these small lumps consist of agglomerates of very small particles each coated with resin, which tend to cling together and retard solvent drying. Mulling action is often insufficient to break up these agglomerates. Analyses have shown that these lumps are not areas of high resin concentration. After thorough drying, they break up quite readily. Additives such as iron oxide, may be added either before or after the coating, with the latter being preferable. Such additives behave in a manner similar to the fine particles referred to above.

Two typical mixtures, suitable for blowing either shell molds or cores, are listed below. One is for use with round grain, and the other with sub-angular grain sand.

	Round Grain	*Sub-Angular Grain*
Sand (AFS 100)	100 lb	100 lb
Resin	2.5 lb (2.44%)	4 lb (3.85%)
Alcohol	.54 lb (305 cc)	.75 lb (425 cc)
Water	.18 lb (80 cc)	.25 lb (115 cc)
Total Solvent	.72 lb*	1.0 lb**

*Resin/solvent ratio of 3.5/1 by weight
**Resin/solvent ratio of 4/1 by weight
Alcohol/water ratio is 3/1 by weight in all cases.

The use of a lubricant is desirable. It may be added "built-in" with the resin or may be added as a separate additive. If the latter is the case, such addition in the order of 2 oz per 100 lb of sand should be made when the mixture begins to break down showing signs of dryness. Commonly used lubricants are zinc, aluminum, and calcium stearates, and certain high grade waxes, added as cold powders and mulled with the mixture during the room temperature coating process.

The above mixtures are suggested starting points only. With the many variables of equipment, sands, and resins, the optimum procedure and technique must be arrived at through experiment in each shop. While resins made by various manufacturers are of the same general family, there are characteristics peculiar to each. It is suggested, therefore, that the resin manufacturer's instructions be followed as closely as possible for best results.

In calculating the mixture, to summarize, there are three factors to be estimated after choosing the proper sand. These are, the percentage of resin, the ratio of resin/solvent, and the ratio of alcohol/water making up the solvent. Note in the typical mixtures that lower resin/solvent ratio was used for the round grain sand mixture than for the sub-angular. This is due to the fact that only 2.5 lb resin is used per 100 lb round grain sand vs 4.0 lb resin for the sub-angular sand. For a constant resin content, the sub-angular grain sand mixture would require more solvent (lower resin/solvent ratio) because of its greater surface area per volume than would round grain sand. Experience will soon clarify these relationships.

The resin coating of sands may be accomplished in either the high-speed or conventional low-speed type mullers or other mixing devices. However, procedure varies considerably between the two types of mullers. Using either, it is best not to interrupt the cycle once it has begun. For employing the low-speed muller, the following procedure is suggested:

1. Clean muller and adjust to the minimum plow clearance.
2. Add sand and resin; mull dry 1-2 minutes.
3. Add solvent—taking 10-15 seconds (See Fig. 7).





Fig. 7 (Left) — Solvent is added to the dry sand-resin mixture. Fig. 8 (Right) — Soon after solvent is added, the mixture develops green strength, or a state of plasticity. It is desirable that this condition exist for at least two minutes.

The mixture will develop green strength* within one or two minutes and should remain plastic for a period of at least two minutes. See Fig. 8.

4. If lubricant is used, add it when mixture begins to show signs of dryness after large lumps begin to break up (approximately 5-9 minutes after starting, depending on conditions). Figure 9 illustrates this stage.
5. Continue mulling until the solvent has evaporated. This should take a total of 10-20 minutes, depending on sand and resin content. Test for residual green strength by squeezing a handful. No lumping should occur in a completely dry mix. A completed mix is illustrated in Fig. 10.
6. Screen mixture if desire. In some mixtures, particularly high in fines, and with mullers having poor plow clearance, a few small lumps are not uncommon. Typical lumps from a sand containing fines are shown in Fig. 11.
7. AVOID OVERMULLING. Once the solvent has dried and the mixture is free-flowing and free of green strength, further mulling abrades the resin coating and reduces tensile strength.

Low speed mullers are currently available which have been modified especially for cold-coating operation. Modifications include the introduction of an air stream, (in some cases heated), supplied to the muller, and in the addition of vertically positioned lengths of angle iron fastened inside the vertical shell of the muller, allowing wheel and plow clearance. Three or four equally spaced angle irons so positioned will prevent the "riding around" of a mix on the plows after having developed green strength or plasticity. The drying of the solvent is an essential function of the resin coating process, hence the muller construction, particularly the cover, should allow for a maximum rate of evaporation.

It must be kept in mind that the alcohol used in the solvent system is inflammable. Proper precautions must be taken to prevent accumulation or ignition of the solvent fumes.

In considering the high speed muller, it is important to remember that, while this type equipment provides intensive mixing on a rapid cycle, its use requires careful timing of the mulling cycle. The following procedure is suggested:

1. Clean muller and adjust wheel spacing to approximately $\frac{1}{8}$ in. The air supply valve should be regulated to avoid blowing out the fine sand particles.
2. Add sand and resin, mull dry 15 seconds.
3. Add solvent taking 10-15 seconds. The mixture will develop green strength within 30 seconds or less and should remain plastic for at least one additional minute.
4. Turn on air blower at two minutes from starting.
5. If lubricant is used, it should be added when a breakdown of the mass occurs. It is necessary to turn off the air but not the muller to add lubricant.
6. Continue mulling. Test for dryness at four minutes after starting. (This is a complete cycle for a round grain sand and $2\frac{1}{2}$ parts resin.) If further drying is needed, test after mulling every 30 or 60 seconds. AVOID OVERMULLING.

One phenomenon experienced in the early days using either type of equipment was in the lack of drying out of some mixtures regardless of the processing time. The cause was finally traced to an occurrence somewhat similar to case-hardening. When the solvent is evaporated too rapidly from around the sand grains, there is left underneath the hardened and dried resin surface, a solvent-rich resin layer which softens and causes caking on further standing. Such resin coated sands, dried by introducing the air too rapidly, tend never to dry out.

*Mixes of approximately one-half to one-third muller capacity are suggested. Overload will cause the entire mixture to ride around on the plows, resulting in poor resin distribution, poor coating, and a generally weak mixture.



Fig. 9 (Left) — After maximum green strength, the mixture will begin to break up into lumps, decreasing in size with continued mulling.

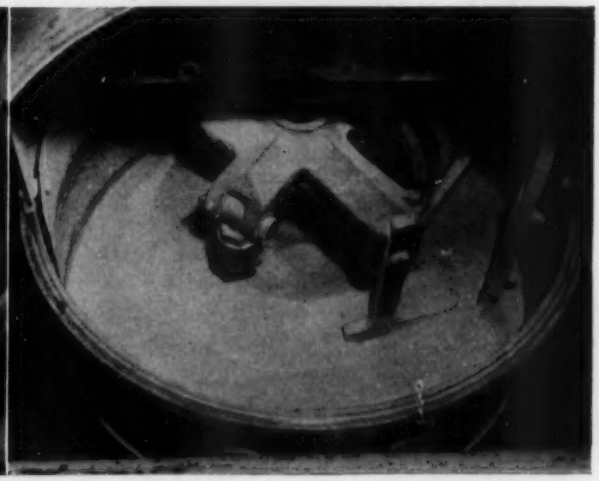


Fig. 10 (Right) — Completed cycle — Resin coated sand.



Fig. 11 — Lumps retained on No. 16 sieve from coated sand containing a high percentage of fine.

Hence, a certain period of time should be allowed for the initial drying of the resin itself before the air is introduced. Failure to remove all solvent will cause soft caking to occur in the coated sand on standing. Properly coated and dried sands have been stored for approximately one year without caking. Periodic examinations showed no loss of tensile strength during this storage.

Using either type mulling equipment mentioned above, the cycle time is dependent upon several factors including:

1. Resin content: more resin, greater cycle time.
2. Resin/solvent ratio: more solvent, greater cycle time.
3. Sand grains: smaller size, or angular shape, greater cycle time.
4. Humidity: high relative humidity in the atmosphere, greater cycle time.
5. Equipment: less availability of drying air or less intense mulling action, greater cycle time.

It appears, particularly with low speed mulling equipment, that the muller alone is not adequate equipment with which to accomplish the rapid and large volume coating of sand with phenolic resins at room temperature. The process requires first the mixing of the sand and liquid or liquified phenolic resin with its distribution over each grain. After this distribution, the solvent is forced to evaporate leaving a dry resin coating on the sand grains. The process also requires the breaking up of small lumps which develop by the agglomeration of either small or large resin-coated particles still possessing some tackiness.

It becomes obvious that a muller has performed its original design function after the first three to five minutes of operation during which time the resin is distributed over the sand grains. It is possible that future equipment development will be along lines to allow performance of this function only in equipment similar to a muller. Subsequent processing in auxiliary drying equipment would permit more intimate contact of the resin-solvent coating with the drying medium, thus shortening the coating cycle. Once fully dried, the coated sand offers little resistance to disintegration of lumps.

Experiments have been made consisting of taking out portions of the mixture following this initial resin distribution and allowing air drying to take place. Subsequent testing proved that the tensile strength of such portions is equal to that of the mixture which was processed to completion in the muller. This is entirely different from the condition in which excessive green strength causes the mixture to "ride up" on the muller plows allowing a drying action to take place. In this latter occurrence, the resin in its liquid or liquified form is never distributed over the sand grains. Such a mix, if air dried without first properly distributing the resin over the sand gains, will result in a weak mixture and consequently poor results in operation.

As in conventional shell molding work, the tensile strength test is about the best indication available for resin-coated sands to determine the probability of successful application. While the tensile strength test is useful, it should not be depended upon too heavily as an indication of ultimate performance characteristics. For example, one coated sand may register a very high tensile strength and yet when blown into shell molds or cores and used to produce castings, the results may be unsatisfactory because of the poor resistance to thermal cracking of the cured mold or core. On the other hand, a coated sand having relatively low tensile strength, when blown into a shell mold or core, may perform quite satisfactorily by virtue of good sand properties and proper additives.

Recommendations of AFS Committee 8-N

Recommended tensile strength testing procedure in shell molding work recently released[3] by the Shell Molding Materials Testing Committee (8-N) of the American Foundrymen's Society Sand Division, involves the dumping of the dry sand-resin mixture onto a heated pattern for the production of tensile specimens. Since coated sands are used primarily in blowing applications, it is the opinion of the writer that in order to be representative of the coated sand at the time of usage, tensile specimens should be prepared by blowing, similar to the blowing of shell cores and shell molds. The following is one procedure for preparation of tensile strength specimens which has been worked out.

Oven temperature is controlled at 700 F. Tensile specimen pattern temperature, read before making specimen at 450 F $\pm$ 10 F.

1. Place coated sand in the cartridge. Position a flat plate under the cartridge and position under the coreblower, apply air pressure for several seconds. This pressure forces the coated sand close together inside the cartridge and prevents leakage of the coated sand through the blow holes.

2. After obtaining correct pattern temperature, locate the cartridge over the tensile specimen pattern and position under the coreblower. See Fig. 12. The cartridge has been modified to allow for blowing of the coated sand and venting to the outside.
3. Apply air pressure at 30 psi* for two seconds.** Avoid prolonged contact between the heated pattern and cartridge since heat will cause the coated sand to stick to the cartridge.
4. Reverse position of the tensile specimen pattern, blowing into the other end repeating steps 3 and 4 above. See Fig. 13.
5. Place the blown, uncured tensile specimens and pattern into the oven and bake at 700 F for two minutes. Lower temperatures may be used. However, baking time must be adjusted to assure complete cure.

Blown specimens average approximately 10 to 25 per cent higher in tensile strength values than those made by AFS Committee 8-N procedure using the same coated sand time-temperature relationship.

This is but one expedient method devised to provide a consistent testing procedure with existing equipment. Others are possible including the use of a specially designed cartridge to blow three tensile specimens simultaneously, and the addition of heating elements to the specimen pattern to eliminate oven curing.

It is obvious from the foregoing that the cold process for the resin coating foundry sands has a multitude of variables. Like any process in its early stages, it is subject to rapid and continuing improvement. Whatever specific paths of progress this process shall take, there is little doubt that we will first hear about, see, and then use, successfully and economically, resin-coated foundry sands, processed in the foundry.

*30 psi was highest feasible pressure allowable to blow coated sand using this blowing arrangement. 35 and 40 psi caused sticking to the tensile specimen pattern face.

**The two seconds referred to was arrived at by experiment noting that a controlled time of air application resulted in more uniform results in various tensile bars under any given set of conditions. Additional time of air application beyond two seconds was not beneficial.

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2. R. L. Olson, "Shell Molding Is Easier," *Foundry*, January, 1955, p. 150.
3. "Foundry Facts,' MODERN CASTINGS AND AMERICAN FOUNDRYMAN, December, 1955, p. 69.

DISCUSSION

Chairman: J. O. OCHSNER, Crouse-Hinds Co., Syracuse, N. Y.

Co-Chairman: E. C. ZIRZOW, Werner G. Smith, Inc., Cleveland.

Secretary: D. J. JONES, New Jersey Silica Sand Co., Millville, N. J.

L. J. PEDICINI[1] *(Written Discussion):* The author has done a commendable job of presenting the techniques of cold coating sand with resin. However, he implies that this technique is one of the well established shell molding processes whereby it is only now coming into the picture in most foundries.

1. Foundry Supervisor, General Motors Corp., Process Development Section, Technical Center, Detroit.

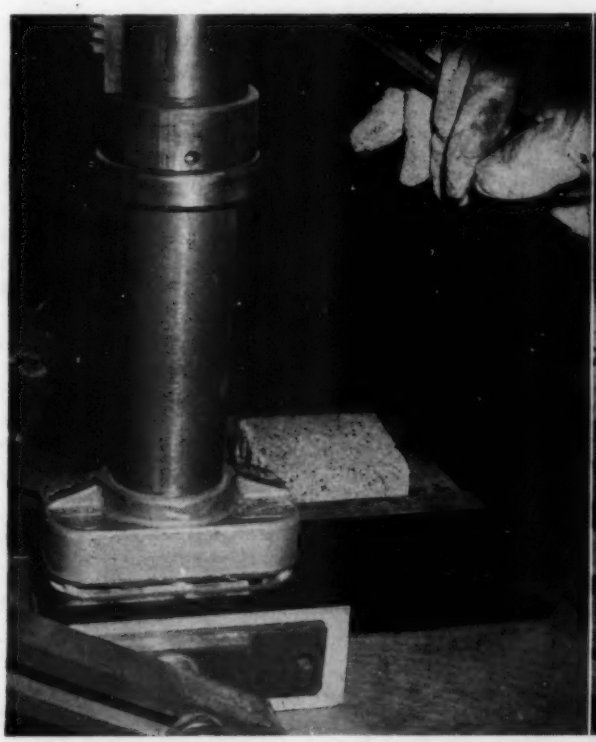


Fig. 12 — Modified cartridge is positioned on tensile specimen pattern.



Fig. 13 — Removing cartridge after blowing second tensile specimen.

The author's reference to another article quoting tremendous savings due to the use of shell cores is somewhat misleading. I am sure that there are a very limited number of applications wherein a shell core will replace an oil-bonded core at one-third cost. I think such a cost reduction may be potentially possible if the cost of the resin can be made competitive with the cost of core oil. In the many applications studied in the automotive industry only those cores which require contour driers, core wash, and other extra processing could be replaced with shell cores at a reduction in cost.

MR. BOLT *(Reply To Written Discussion By Mr. Pedicini)*: There was no intent to show this coating process is in widespread use. It is something quite new. It depends upon each shop's applications as to whether or not the coating process is accepted. They may prefer to use the dry mix method or conventional core mix in production. This cold coating process has followed the same general line of development as did the shell process from its inception. The consumption of resins is now climbing based on the work done several years ago on the dry mix method. We can expect more widespread application of coated sand in the near future based on present development effort.

As to the one-third cost of the conventional core, the right application must be there for this to be possible. Generally speaking, to be comparable in material costs alone, a shell core must not exceed one-third to one-fourth the weight of a solid core made from conventional core mix. You also have other potential savings which are, elimination of coatings, core ovens and driers, and the quality of the cavity produced by this core. The advantages are commensurate with the shell molding practice, as such. It depends upon how much one can utilize the process.

Coating costs money. However, it is in this field that the changeover to shell cores offers an attractive saving.

M. H. HORTON[2]: It is possible to use less resin on the round-grained sand than on the angular. Would it be possible to increase the use of solvent and use the same amount of resin used with the round-grained sand? Could you increase your drying time to compensate for the increased solvent? We found that tensiles are related to the rate of solvent removal.

MR. BOLT: In answer to your first question, on strictly a comparable tensile strength basis, yes, less resin is required for round grained sands. However, they are not always best in performance. Increasing the solvent will increase tensile strength only if insufficient solvent has been used formerly, as illustrated on the left side of the curves shown earlier. Increased drying or total cycle time is required for the increased solvent; however, this extra time is usually relatively short. All of our data are based on the complete removal of the solvent. You can evaporate too fast and come up with a condition similar to case-hardening.

It is my opinion that tensiles differ because of the area of the grain to be covered. At the same time, one school of thought maintains that the thinner the coating, the higher the bond strength. The tensile strength test is not the ultimate. I wish we had a better test available. I agree with your statement that there is a lot of confusion existing about the cold coating of sand.

W. H. BUELL[3]: Other conditions being the same, is thermal cracking of phenolic-bonded shells greater or less with round-grained sand than it is with subangular-grained sand?

MR. BOLT: Round grain sands do tend to crack more than the sub-angular. This is probably explained by more interlocking of the sub-angular sand grains. At the same time, some resins tend to show more cracking than others. The resin has to be distributed well over the sand grains in order to get the most out of the percentage used.

The Shell Molding Materials Testing Committee (8-N) is actively engaged in trying to correlate results of such testing but, to date, there has not been too much correlation. Frankly, there is much work yet to be done before we can say conclusively that one sand is definitely superior to another providing certain factors are taken into consideration.

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2. Materials Engineering Div., Deere & Co., Moline, Ill.
3. Vice President, Aristo Corp., Detroit.

NEW ALUMINUM PERMANENT MOLD CASTING ALLOYS C355 AND A356

By

R. C. Lemon* and H. Y. Hunsicker**

ABSTRACT

This paper describes two new aluminum alloys that provide high levels of strength and ductility for permanent mold castings. Designated C355 and A356, these higher purity variations of alloys 355 and 356 offer the design engineer a wide range of mechanical properties.

For applications requiring maximum strength, alloy C355 may be used in the -T62 temper, which provides in separately cast test bars a typical tensile strength of 52,000 psi, yield strength of 44,000 psi and 4 per cent elongation. The higher ductility afforded by the -T61 temper of this alloy matches that property in 356-T6 but is accompanied by appreciably higher tensile and yield strengths.

Alloy A356-T61 has about double the amount of ductility of 356-T6 and also tensile and yield strengths that are somewhat higher. In the -T6 temper, alloy A356, with a typical elongation value of 15 per cent, is suited for applications requiring especially high ductility.

Introduction

For many years, aluminum alloy castings have been used for stressed structural parts in aircraft, automotive and machine equipment. The demands for lighter and more efficient equipment are ever present, however, and to help meet this need significant progress has been made in improving processes to produce castings of higher and more uniform quality. Among established casting methods, the permanent mold process has remained attractive for medium to high volume production, and advances have been made in the size and complexity of parts that may be produced by this method. The mechanical properties of well designed permanent mold castings produced with adequate control of foundry operations are generally higher than those afforded by castings produced by methods which cause less rapid solidification.[1]

Further progress toward higher strengths and load carrying capacity in such parts appeared possible through changes in alloy composition, and with this as an objective, an investigation was initiated. Among the different alloy compositions available for producing permanent mold castings for structural use, the aluminum-silicon-copper-magnesium alloy 355 and the aluminum-silicon-magnesium alloy 356 have long been favorites both in the foundry and in service. Their popularity stems from a combination of desirable properties that include excellent casting characteristics, high mechanical properties and outstanding resistance to corrosion. Other types of alloys are known that excel in one or another of these qualities, but none of the others has exhibited as favorable a balance of characteristics.

The two alloys, nominal compositions of which are shown in Table 1, offer several combinations of strength and ductility in the various heat-treated tempers available, the ductility generally decreasing with increasing strength. The maximum strength, for example, is provided by alloy 355-T62, but no minimum elongation is guaranteed for this material.

TABLE 1 — NOMINAL COMPOSITIONS OF ALLOYS 355 AND 356

Alloy	Cu, %	Si, %	Mg, %
355	1.3	5.0	0.5
356	—	7.0	0.3

Despite the necessity of designing to a lower stress level to use alloy 356-T6, a preference for it has developed because of its superior ductility. The added ductility provides opportunity for sufficient deformation to permit redistribution of stresses in the event of overloading. For aircraft applications, an elongation value of 3 per cent has frequently been quoted as a desirable minimum value.

In the publication by Archer and Kempf[2] describing developmental work leading to the establishment of alloys 355 and 356, information was included concerning effects of composition variations on the tensile properties, particularly those of sand castings tested in the -T4 temper, that is, after quenching from the solution heat treatment. These data, though not directly applicable to permanent mold casting, gave strong indications that improved

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combinations of strength and ductility might be attained by adjustments in the compositions of the alloys and justified a further examination of the effects of composition variations.

Experimental Procedure

In the investigation, all alloy variations were prepared and handled according to established foundry practices. Standard cast-to-size tensile specimens were cast in a permanent mold. Care was exercised to assure uniformity of mold operating practices. In heat treating the tensile specimens, precautions were taken to control the interval of room temperature aging after quenching from the solution heat treatment, and before artificial aging.[3]

Effect of Chemistry on Tensile Properties

Silicon

Archer and Kempf presented data showing that the ductility of aluminum-silicon-magnesuim and aluminum-silicon-magnesium-copper alloy sand castings changed little with variation in silicon content between 5 and 10 per cent, but improved with decreasing silicon below 5 per cent.

To determine whether a similar relationship could be expected for permanent mold casting, several lots of tensile specimens were cast in the four alloy compositions listed in Table 2. As shown by the average tensile properties listed, no advantage was found for the permanent mold cast alloys containing 3 per cent silicon compared with those at the higher silicon content of 5 per cent. Since other experience indicated that less favorable casting characteristics were encountered at the lower silicon level, and no mechanical property advantage was apparent, the lower silicon content alloys were given no further consideration.

TABLE 2 — EFFECT OF SILICON CONTENT ON TENSILE PROPERTIES OF HEAT TREATED* PERMANENT MOLD CAST SPECIMENS

	Composition, %				Tensile Strength, psi	Tensile Yield Strength, psi**	Elong., % in 2 in.
	Cu	Fe	Si	Mg			
A.	1.25	0.15	3.0	0.5	49,700	31,700	12.8
B.	1.25	0.15	5.0	0.5	51,100	34,000	11.0
C.	0.00	0.15	3.0	0.5	43,300	32,000	10.0
D.	0.00	0.15	5.0	0.5	44,400	33,800	10.5

*Heated 8 hr at 980 F (A and B) or 1000 F (C and D), quenched in boiling water, aged 4 hr at 310 F.
**Offset = 0.2%

Magnesium and Copper

Increasing magnesium contents up to 0.5 per cent produce progressively higher tensile and yield strengths and lower ductility in the aluminum-silicon alloys. On the other hand, copper additions up to 1.25 per cent have been used to raise the strength properties with only slight apparent effects on ductility.

With the object of developing an alloy with high ductility, permanent mold tensile specimens were prepared from alloys containing 7 per cent silicon and 0.15 per cent iron with magnesium contents

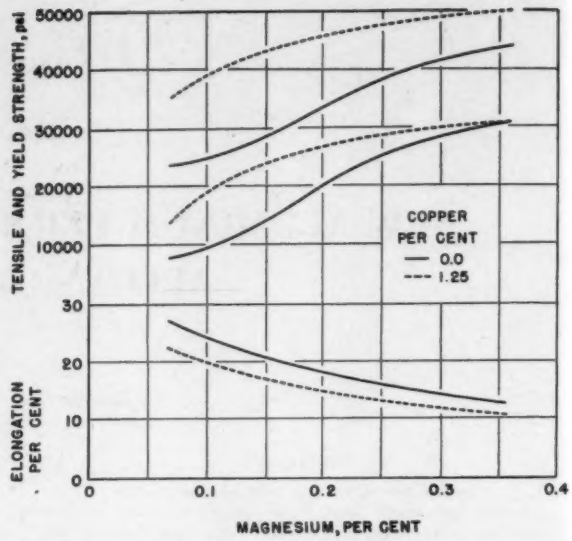


Fig. 1 — Effect of magnesium on tensile properties of some permanent mold cast alloys containing 7% Si, 0.15% Fe with 0.0 and 1.25% Cu, given a -T6 type heat treatment.

from 0.05 to 0.3 per cent and copper contents of 0.0 and 1.25 per cent. The tensile properties of specimens given a -T6 type heat treatment are illustrated in Fig. 1. These data demonstrate the improvement in tensile and yield strengths afforded by the additions of magnesium and copper. The elongation values were decreased by the magnesium and copper additions. Figure 2 shows graphically the relationship between the elongation values and the tensile and yield strengths of these alloys. It also shows that the yield strength was always reduced when the elongation was increased. However, it was demonstrated that the addition of a small copper content produced a higher tensile strength without sacrificing ductility.

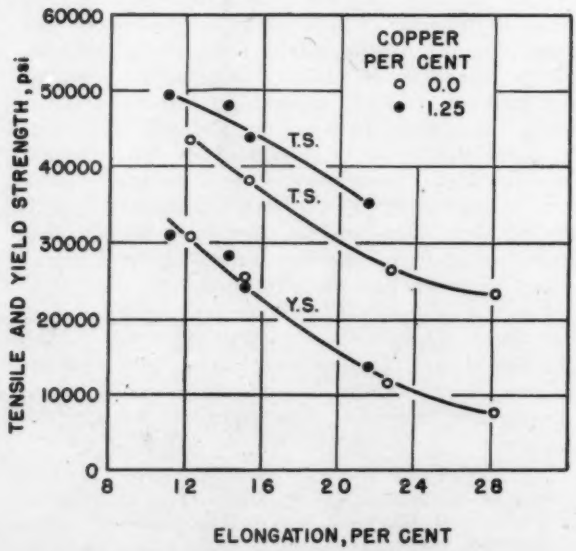


Fig. 2 — Relationship of tensile and yield strengths to elongation of alloys shown in Fig. 1.

As was indicated previously, varying the silicon content in the range of 5 to 10 per cent did not appreciably influence the tensile properties. It is reasonable to conclude, therefore, that alloys containing 5 per cent silicon would provide essentially the same properties as were obtained with the 7 per cent silicon alloys.

Iron

Iron has long been known to have a potent influence on the ductility of aluminum alloys. To obtain data on permanent mold cast 355-T6 and 356-T6, tensile specimens were prepared and tested in which the iron content was varied within the ranges permitted by the specifications for these alloys. The results, shown in Fig. 3, corroborate the conclusions drawn by Archer and Kempf with respect to the deleterious effect of iron on the ductility, but their observations that tensile strength also decreased substantially with increasing iron was not demonstrated by the alloys cast in permanent molds.

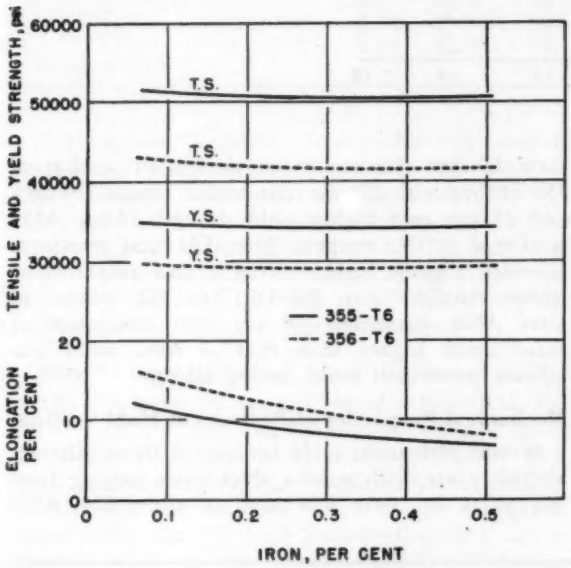


Fig. 3 — Effect of iron on tensile properties of permanent mold cast 355-T6 and 356-T6.

Selected Chemical Compositions

Consideration of the data reported briefly above demonstrated that iron in decreasing quantity increased ductility without sacrifice of either tensile or yield strength. Furthermore, other experimental work has indicated that the variations in tensile properties produced by changing the conditions of heat treatment are interrelated in a manner that necessitates sacrificing some strength to obtain higher ductility. Consequently, alloys were selected having the same nominal chemical compositions as 355 and 356, but with the maximum iron contents reduced to 0.20 per cent in place of 0.6 per cent and 0.5 per cent, respectively. Maximum limits for other impurity elements were also lowered slightly in setting up the complete specifications. The high purity variations have recently been designated C355 and A356.

Heat Treatment and Mechanical Properties

The aging curves of alloys C355 and A356 shown in Fig. 4 and 5, aided in the selection of heat-treating procedures that would provide combinations of tensile properties utilizing the superior ductility to best advantage. Also, several hundred permanent mold tensile specimens were given selected heat treatments and tested to establish typical and guaranteed minimum tensile properties for specific tempers of the two alloys. The properties established as a result of this survey are listed in Table 3. A comparison of the typical tensile properties with those of alloys 356-T6 and 355-T62 is shown in Fig. 6, which also illustrates the wide range of properties available in the selected tempers of the new alloys.

Alloy C355-T61 provides the same 3 per cent minimum elongation as that of 356-T6 but with 21 per cent higher tensile strength and 36 per cent higher yield strength. Alloy C355-T62 provides the

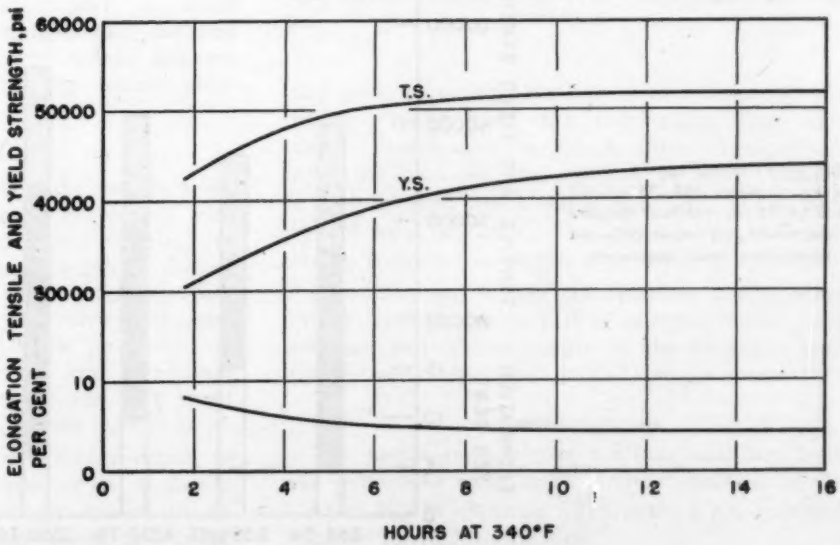


Fig. 4 — Aging curves of alloy C355 at 340 F. Solution heat treated 8 hr at 980 F, quenched in boiling water.

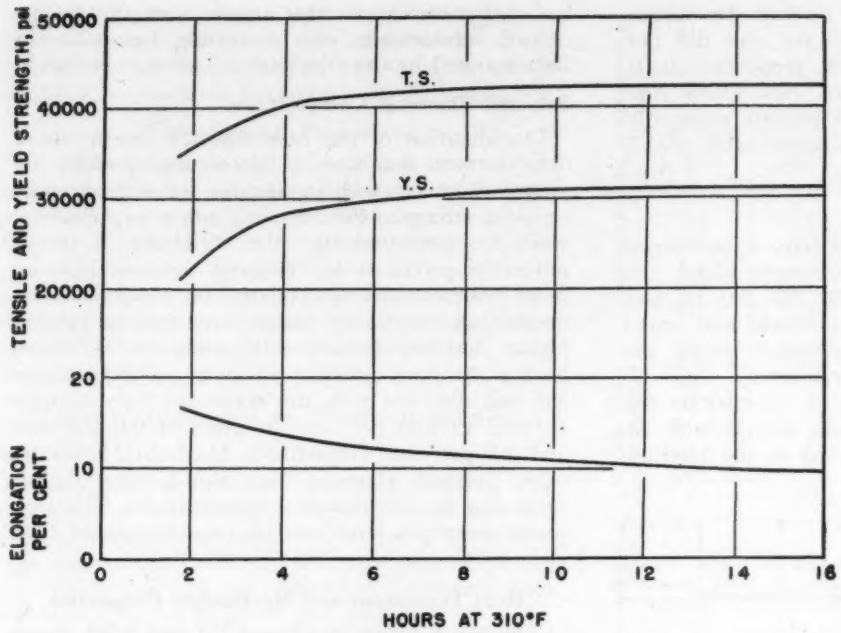


Fig. 5 — Aging curves of alloy A356 at 310 F. Solution heat treated 8 hr at 1000 F, quenched in boiling water.

TABLE 3 — MECHANICAL PROPERTIES OF PERMANENT MOLD CASTING ALLOYS C355 AND A356 IN VARIOUS TEMPERS*

	Minimum Values for Specifications			Typical Values (Not Guaranteed)		
Alloy	Tensile Strength, psi	Yield Strength,** psi	Elong., % in 2 in.	Tensile Strength, psi	Yield Strength,** psi	Elong., % in 2 in.
A356-T6	32,000	18,000	10.0	37,000	22,000	15.0
A356-T61	37,000	26,000	5.0	41,000	30,000	10.0
C355-T61	40,000	30,000	3.0	46,000	34,000	6.0
C355-T62	47,000	40,000	1.5	52,000	44,000	4.0

*Tests of separately cast permanent mold tensile specimens.
**Offset = 0.2%

same 1.5 per cent minimum elongation as that of 355-T6 but with 27 per cent higher tensile strength and 43 per cent higher yield strength. Alloy A356 is offered in two tempers. The -T61 heat treatment provides slightly higher strength and substantially higher ductility than 356-T6. The -T6 temper of alloy A356 guarantees 10 per cent elongation, a value much higher than that of most other aluminum permanent mold casting alloys.

Mechanical Properties of Permanent Mold Castings

Several permanent mold castings of three different aircraft parts, with section thicknesses varying from $\frac{3}{16}$ to $\frac{3}{8}$ in., were fabricated of alloys 356, A356

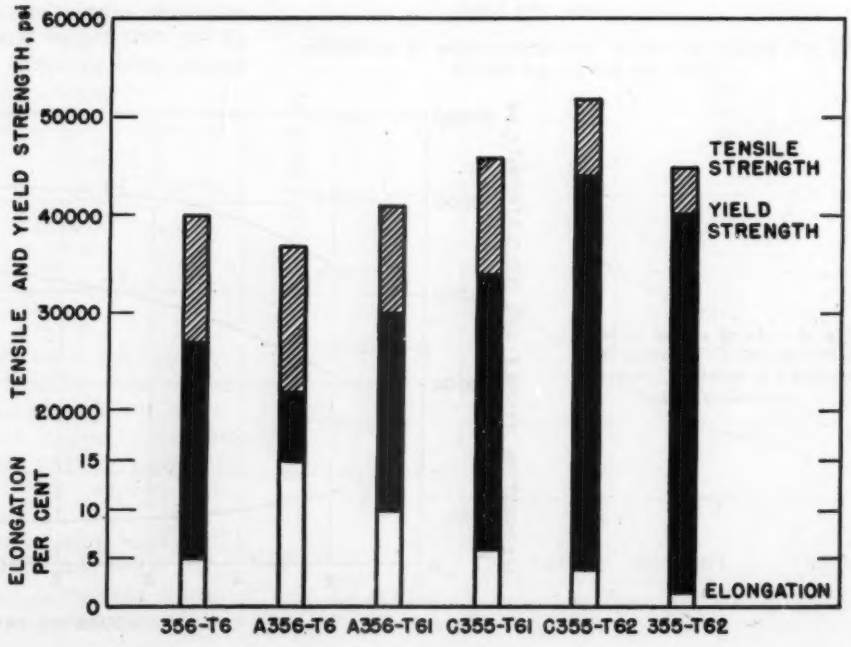


Fig. 6 — Typical tensile properties of alloys 355, 356, C355 and A356 in various tempers determined on separately-cast permanent mold specimens.

and C355. The castings, showing outlined locations from which test specimens were taken, are illustrated in Fig. 7. The typical tensile properties obtained are shown in Table 4.

TABLE 4 — AVERAGE TENSILE PROPERTIES OF SPECIMENS MACHINED FROM PERMANENT MOLD CASTINGS SHOWN IN FIG. 7

Cstg.	Weight, lb	Section Thickness, in.	Alloy and Temper	Tensile Strength, psi	Tensile Yield Strength*, psi	Elong., % in. 2 in.
A	¾	3/16-¼	356-T6	32,000	24,250	4.6
			A356-T61	40,400	28,500	10.0
			C355-T62	47,150	41,650	2.8
B	1	3/16-5/16	356-T6	33,100	28,550	2.6
			A356-T61	37,700	29,000	6.7
			C355-T62	50,900	44,900	2.1
C	1½	¼-⅜	356-T6	33,350	26,200	3.6
			A356-T61	39,450	29,050	9.5
			C355-T62	49,100	42,700	2.3

*offset = 0.2%

The tensile specimens from C355-T62 castings possessed a wide margin of strength over those of 356-T6 without substantial sacrifice in ductility. On the other hand, alloy A356-T61 also with higher tensile and yield strengths than 356-T6, exhibited large improvement in ductility. These advantages offer improvements from the structural standpoint that are useful to the design engineer.

Other Characteristics of Alloys C355 and A356

The similarity in composition of alloys C355 and A356 with 355 and 356, respectively, should make them also similar in such characteristics as electrical and thermal conductivity, thermal expansivity, density, machinability and castability. It is expected also that the high purity alloys will be at least as good as the standard alloys in fatigue strength and resistance to corrosion; however, work is in progress to assess these properties of the new alloys. Mechanical characteristics such as shear strength, compressive yield strength and elevated temperature mechanical properties should conform to the usual relationships with the tensile properties of these alloys.

Applications

These two high purity alloys in the various available tempers present to the design engineer a wide range of tensile properties from which to select for stressed parts. Where maximum strength is desired and ductility requirements are moderate, alloy C355-T62 may be used. The higher load-carrying capacity of this alloy indicated by the tensile properties has been verified by simulated service and breakdown tests of such parts as textile beam heads, high speed rotating components and aircraft and missile items.

Alloy A356-T6 exhibits a much higher degree of ductility than most cast aluminum alloys. It should be advantageous for parts that are subject to accidental overloading, particularly by impact loads, and parts requiring that only deformation and not fracture result from such overloading. One commercial application in which these characteristics were the basis for the selection of alloy A356-T6 is a heavy-duty truck suspension part. This alloy should also be useful if a high degree of malleability is needed to permit some forming operations.

The alloy and temper combinations that produce intermediate properties will be of equal interest, and selection will depend chiefly on the designer's preference with respect to ductility requirements.

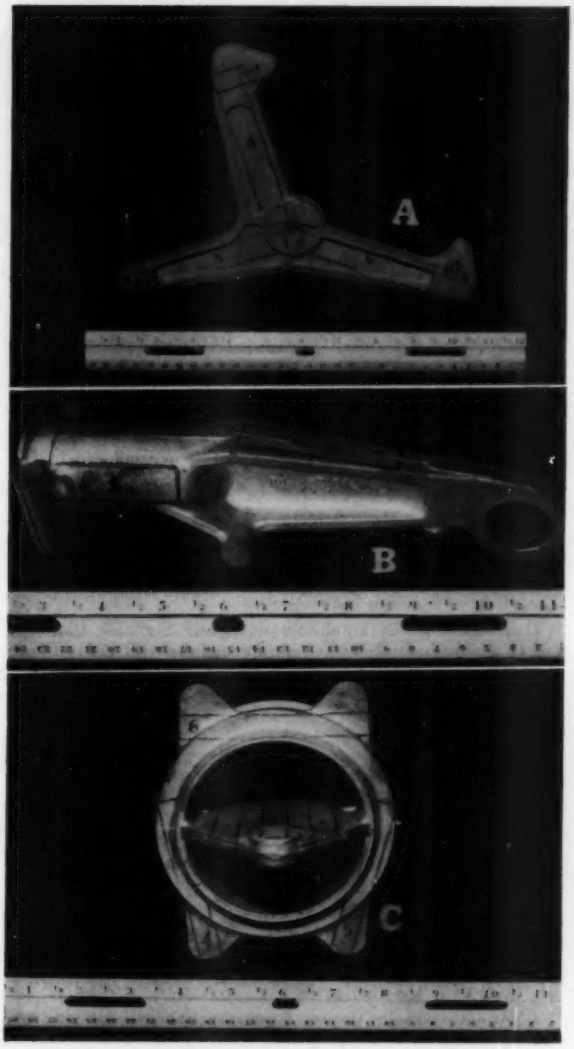


Fig. 7 — Permanent mold cast airframe parts produced in alloys 356-T6, A356-T61 and C355-T62. Markings on castings designate locations from which tensile specimens were machined.

Acknowledgment

The tensile properties of test bars machined from the permanent mold castings were obtained at the Vernon Works of Alcoa. The authors are indebted to them for this information.

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DISCUSSION

Chairman: A. G. EALY, Littlestown Hardware & Foundry Co., Littlestown, Pa.

Co-Chairman: C. SZEGO, Mold Cast Products, Newark, N.J.

WALTER BONSACK[1] *(Written Discussion)*: Congratulations to the authors for their fine reporting on the higher purity variation of 355 and 356 alloy. As pointed out in Fig. 3 and by the writer's publications many years ago, iron is one of the elements which rapidly reduces ductility in silicon alloys. Another element is calcium.

With the event of prime aluminum production with iron as low as 0.10-0.20 per cent and a low calcium content, and silicon production with a low calcium content at a reasonable price a new field of endeavor has been opened up.

This higher purity aluminum was made available from two other producers and tests running parallel to the authors' report have been produced by a number of foundries. It is gratifying to see such close agreement. Figure A shows as a solid line 356 alloy with an average of 0.30-0.50 per cent iron and the same heat treatments. The results* are comparable to each other and to the authors' results. The tensile and yield strengths are nearly as high as with present day 356, but the elongation is practically 100 per cent higher, close to 20 per cent at T6 conditions. This of course opens up better ways of (1) alloying and (2) heat treating.

1a) Adding alloying elements which were interfered with by iron to give better properties such as manganese, chromium and magnesium.

1b) Adding larger amounts of grain refiners such as titanium, boron and zirconium to obtain finer grain. Metals of higher purity exhibit larger grain and more grain refiners are necessary to obtain the fine grain.

2) As indicated by the solid line, by judicious alloying and heat treating desirable properties can be obtained. Sure, the elongation drops rapidly, but taking the hatched line of elongation, we find that the elongation is doubled by dropping the iron content just about 50 per cent, i.e., 0.35 per cent to 0.20 per cent or a little lower. The low iron-aluminum can only be supplied by our prime producers of aluminum.

The same conditions of course prevail in sand casting. Just for the record** alloy 355 with 0.14 per cent iron has been tested in T61 condition at 35000 psi yield strength, 42000 tensile strength and 3-5 per cent elongation depending on the amount of grain refiners. Alloy 356 of equal purity sand cast, provided properties of 31-35000 psi yield strength, 39-42000 psi tensile strength and 3.5-6.5 per cent elongation. In this case again the iron content was far below the normal limit and allowed more grain refiners and higher known beneficial alloying elements to achieve these good desired properties.

Experience with semi-permanent mold reveal the same advances.* Coupons cut from an 8-lb. casting in 356 T62 alloy gave 32000-34500 psi yield strength, 40800-43700 psi tensile strength and 4.5-6.0 per cent elongation. The accompanying standard testbars produced 32500-34600 psi yield strength, 44500-45200 psi tensile strength and 9-12.5 per cent elongation.

*Courtesy of Century Metalcraft Corp., Los Angeles, Calif.
**Data supplied by Wm. F. Jobbins, Inc., Aurora, Ill.

These experiences recorded above and by the authors should be enjoyable to every foundryman, since they indicate that production of higher purity aluminum at reasonable cost gives him a chance to produce in turn higher property castings which are so widely in demand.

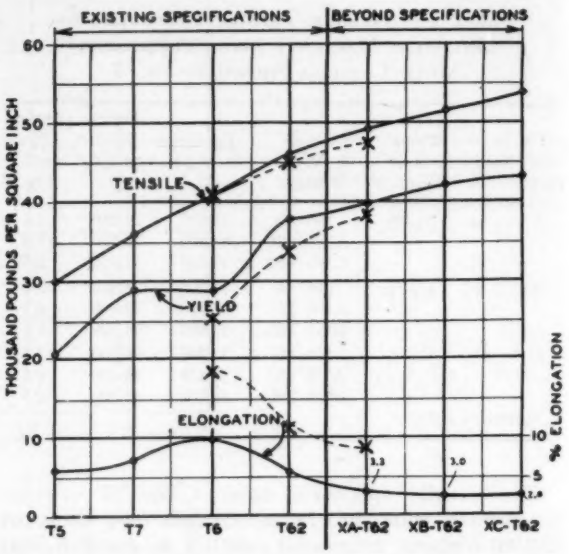


Fig. A — Average of test results on 356 type alloy (permanent mold cast).

MESSRS. LEMON and HUNSICKER: The authors greatly appreciate Mr. Bonsack's comments and the additional data which he has supplied. These supplementary test results show good general agreement with the authors' data as pointed out by Mr. Bonsack and demonstrates further the benefits of the new alloys.

K. E. NELSON[2]: What are the high-temperature properties of these alloys?

W. D. WALTHER[3]: What are the fatigue properties of these alloys?

MESSRS. LEMON and HUNSICKER: Although fatigue tests and elevated temperature tensile tests are underway on these new alloys, no data are yet available to answer the questions by Messrs. Nelson and Walther.

W. A. MADER[4]: What percentage of grain refiners was used in melting? What was your melting procedure, degassing, etc.?

MESSRS. LEMON and HUNSICKER: In reply to Mr. Mader's questions, practices regarding grain refining, melting, degassing, etc., are the same for the new alloys as for alloys 355 and 356.

D. L. COLWELL[5]: What is the effect of small amounts of impurities on the heat treated alloys? What was the exact heat treatment used on these alloys?

MESSRS. LEMON and HUNSICKER: The question raised by Mr. Colwell in regard to the effects of impurities undoubtedly refers to elements other than iron. In setting up specifications for alloys C355 and A356, maximum impurity limits for a few elements were reduced slightly from those in 355 and 356. The presence of small quantities of certain elements such as titanium has no harmful effect on tensile properties.

The heat treatments required for these alloys to produce the listed tensile properties were based on the artificial aging data given in the paper.

1. Consultant, North Hollywood, Calif.
2. Dow Chemical Co., Midland, Mich.
3. Ford Motor Co., Dearborn, Mich.
4. Oberdorfer Foundries, Inc., Syracuse, N. Y.
5. Apex Smelting Co., Cleveland.

NEW ALUMINUM CASTING ALLOY XA140 FOR ELEVATED TEMPERATURE APPLICATIONS

By

R. C. Lemon* and W. E. Sicha**

ABSTRACT

A new aluminum casting alloy designated XA140 has been developed for elevated temperature applications. This new alloy contains 8 per cent copper, 6 per cent magnesium, and 0.5 per cent each of manganese and nickel. At temperaures of 400 to 600 F, the tensile and fatigue strengths of alloy XA140 are superior to those of other commercial aluminum casting alloys, and its resistance to loss of high temperature strength upon prolonged heating is outstanding. Production to date, though limited, has demonstrated that alloy XA140 has good casting characteristics.

Introduction

Technological advancements in several fields have increased the need for improved performance of cast aluminum alloy structural components that function at temperatures in the range of 400 to 600 F. For many years, the aluminum alloys 142 (4 per cent copper, 1.5 per cent magnesium, 2 per cent nickel) and A355 (5 per cent silicon, 1.4 per cent copper, 0.5 per cent magnesium, 0.8 per cent each manganese and nickel) were recommended and used for castings expected to be operated at elevated temperatures. Superior properties at elevated temperatures were afforded by aluminum alloy 254 (6 per cent magnesium, 1.5 per cent nickel, 1 per cent manganese), but this alloy never gained commercial importance because of unsatisfactory casting characteristics.

This paper describes the recently developed aluminum casting alloy, XA140, which exhibits outstanding mechanical properties at temperatures of 400 to 600 F together with good casting characteristics. Alloy XA140 has the nominal composition: 8 per cent copper, 6 per cent magnesium, 0.5 per cent each of manganese and nickel, with small quantities of boron and titanium for grain refinement and beryllium for oxidation resistance.† The influence of some chemical composition variations is described and mechanical and physical properties are given.

*Research Metallurgist and **Chief, Cleveland Research Division, Aluminum Research Laboratories, Aluminum Company of America, Cleveland.

†Patent Pending.

Procedure

Uniform melting and handling practices were employed in preparing test specimens. Melts were heated to 1400 F and stirred well at that temperature to assure solution of high melting additions. Chlorine gas was passed through the melts for periods of from 5 to 15 minutes depending upon the size. All specimens were subjected to radiographic examination to assure uniformity of quality.

Since the mechanical properties of alloy XA140 at room or elevated temperatures are not improved substantially by heat treatment, all specimens were tested in the as-cast (-F) condition.

Procedures followed for conducting tensile tests,[1] creep tests[2] and fatigue tests[3] at elevated temperatures have been described previously.

The brinell hardness tests at elevated temperatures were conducted with a device adapted to the elevated temperature tensile testing machine. This equipment provided excellent temperature control of the specimen and the penetrator. The penetrator was a 10-mm tungsten carbide ball.

Chemical Composition

Moderate allowances for variation from the nominal copper and magnesium contents would be required in production of XA140 alloy. Examination of the effects of deviations from the nominal contents of these two elements, by ± 1 per cent copper and ± 0.5 per cent magnesium, disclosed an insignificant influence on the tensile properties at room temperature. Representative values were a tensile strength of 33,000 psi, a yield strength of 29,000 psi and 1 per cent elongation. The influence of these variations in composition upon tensile properties at 600 F is indicated graphically in Fig. 1. It is evident that the copper and magnesium contents can vary within commercial ranges without appreciably changing the tensile properties at 600 F.

The nominal contents of 0.5 per cent manganese and nickel in alloy XA140 substantially improve the tensile properties at 600 F over those provided by a similar alloy without these elements. However,

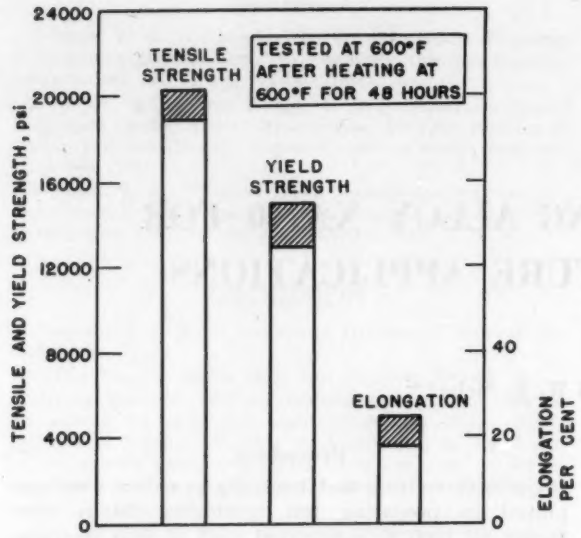


Fig. 1 — Effect on tensile properties at 600 F of variations within composition limits of alloy XA140-F shown by cross-hatched areas.

increasing the manganese or nickel to 1 or 1.5 per cent does not offer any substantial improvement in properties at 600 F. Therefore, variation of these elements within the small range permitted commercially should have no appreciable effect on properties.

Mechanical Properties

The superior tensile properties of alloy XA140-F are illustrated by a comparison with some other commercial aluminum casting alloys in Fig. 2. At 600 F, after 1000 hr heating at that temperature, this new alloy exhibits about double the tensile strength and more than double the yield strength

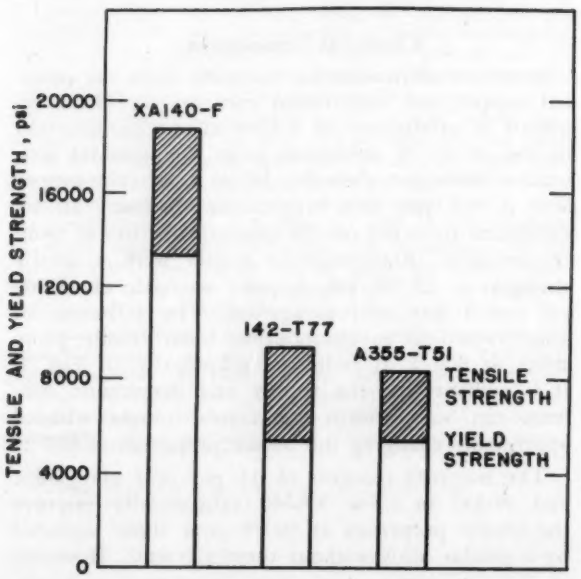


Fig. 2 — Tensile properties of some aluminum alloys at 600 F after heating 1000 hr at 600 F.

of alloy 142-T77. Limited tests indicate alloy XA140 exhibits creep and stress rupture properties that, in comparison with other aluminum casting alloys, are superior at 400 F but slightly lower at 600 F. Also, fatigue strength at 400 and 600 F has been indicated by limited data to be substantially higher than that of the other casting alloys.

Elevated temperature tensile properties of XA140 are very stable as shown in Fig. 3. Increasing the heating periods up to 10,000 hr at 400 F initially raised and then reduced the strength to about the levels provided by heating for ½ hr. At 600 F, extending the heating periods caused only small reductions in strength. The alloy may also experience short-time exposure to temperatures up to 700 F without noticeable effect upon the 600 F strength.

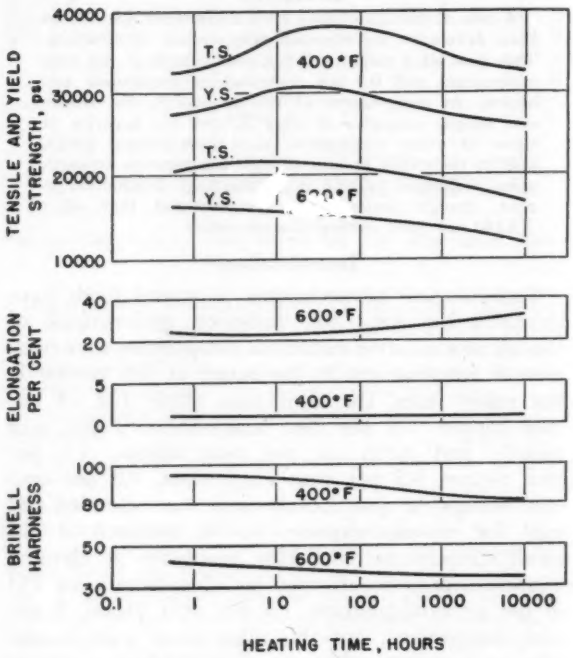


Fig. 3 — Effect of heating time on tensile properties of alloy XA140-F at 400 and 600 F.

Figure 3 also shows the brinell hardness of alloy XA140-F at 400 and 600 F. The alloy exhibits higher hardness than any other commercial aluminum alloy in this temperature range.

Physical Properties

Measurements to determine the growth characteristics of XA140-F have revealed that this alloy is extremely stable dimensionally, exhibiting no growth after prolonged heating. Other physical characteristics of the alloy are as follows:

Weight, lb per cu in.	0.101
Approximate Melting Range, F	950-1110
Electrical Conductivity, per cent I.A.C.S.	23
Thermal Conductivity at 25 C, c.g.s. units	0.23

Average Coefficient of Thermal Expansion per F

Range 68 to 212 F	12.3 x 10^{-6} in./in.
Range 68 to 392 F	13.0 x 10^{-6} in./in.
Range 68 to 572 F	13.5 x 10^{-6} in./in.

Casting Characteristics

Foundry experience has demonstrated alloy XA140 to have good casting characteristics. The alloy has an abundant quantity of eutectic to promote feeding and aid in the production of sound castings. Permanent mold castings of a few parts have been made successfully, but it may be necessary to compensate for a mild hot cracking tendency.

Applications

Alloy XA140 is designed primarily for applications that require good strength at elevated temperatures in the range of 400 to 600 F. The most promising field at the present time appears to be in the aircraft engine industry, particularly in gas turbine engines. Superior performance should be provided by using alloy XA140 for compressor castings that operate at the higher temperatures desired for improved engine efficiency.

This new alloy may be suitable for gas turbine accessory equipment powered by small reaction turbines operating on compressed gas at 400 to 600 F. Since certain airframe and guided missile parts may attain temperatures in the 400 or 600 range, possible applications for alloy XA140 should exist in these fields.

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DISCUSSION

Chairman: A. G. EALY, Littlestown Hardware & Foundry Co., Littlestown, Pa.

Co-Chairman: C. SZEGO, Mold Cast Products, Newark, N. J.

D. L. COLWELL[1]: Why is there a need for manganese and nickel together?

MESSRS. SICHA and LEMON: The question as to the need for both manganese and nickel in alloy XA140 is answered in the affirmative. Both of these additions contribute to the elevated temperature properties, although larger quantities than those specified by the composition limits are of no appreciable further benefit.

D. A. JOHNSON[2]: Were the castings which were made from these alloys pressure-tight?

MESSRS. SICHA and LEMON: This question cannot be answered directly since castings were not tested for pressure tightness. However, from observation of some exceptionally sound castings produced in alloy XA140 it would be expected that pressure tightness should present no appreciable difficulty.

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1. Apex Smelting Co., Cleveland
2. Bendix Products Div., South Bend, Ind.

CASTING TOLERANCES AS AFFECTED BY AUTOMATION IN THE MACHINE SHOP

By

E. L. Buchman*

The modern machine shop of today has become highly mechanized by the introduction of what is commonly known as "automation." One definition of automation is the automatic handling of parts between progressive production processes. However, this mechanization presents new and additional problems, especially for the foundry. It means that the foundry must have better planning, closer control of its processes and better than average inspection of its castings. It means that greater precision in making cores and molds, in melting, and in cleaning and inspection of castings has become the new order of the day.

Some of us, whose customers have mechanized machine shops, have already been confronted with these various problems. The others will become aware of what is ahead of them, and can possibly benefit by our actual experiences.

When the initital casting blueprint is received, the pattern engineers and the production department must meet to decide the best way to produce the casting, i.e., where the parting line will be, where cores are to be used, style of gating system, the kind of metal, etc. The machine shop must also be consulted so that consideration can be given to locating points or the amount of finish stock that can safely be removed during automatic machining.

After sample castings are made and checked in lay-out, they are put through the machine shop automation to observe contact points of the clamping and hold-down mechanism. Knowledge of clearances is required in the transfer and indexing part of the automation so that the foundry inspector knows exactly how much flash or parting fin can be tolerated.

Material Specifications Must Be Maintained

Now let's go back to the foundry—begin with the cores—and see what problems occur or what can happen if they are overlooked.

To produce consistently accurate cores, blueprint dimensions and material specifications must be maintained in the use of equipment and materials selected. A good core mixture must be established and rigidly adhered to, so that the cored area of the casting will be true to print and free of defects caused by inferior supplies or imperfect processing methods.

Therefore, constant checks must be made of incoming shipments of materials and the processes involved, as well as continual inspection for dimensional accuracy. In this day of machine shop automation, dimension looms most important because allowable tolerances have diminished to where the commercial casting of yesterday is not suitable today.

In some cases, all machining starts from a cored hole or a cored surface. Because automation is exactly what its name implies, strict dimensions and tolerances must be maintained, otherwise slight offlocations may not be discovered until the part is completely finished and the castings have become scrap.

Cores must be assembled accurately and in some cases gaged 100 per cent to make certain that they are within dimensional tolerances. On one cylinder block, for example, the location for the first machining operation is taken from two cored holes on the head deck of the block. These cored holes are a part of the jacket core. If the jacket core is shifted, it means that all subsequent machining operations will be incorrect. It could mean that some of the operations would be taking too heavy a cut and might therefore tear out a single or an entire bank of tools.

Casting Tolerances Must Be Maintained

In other cases, the casting may become jammed in the automation, because casting tolerances were not maintained. A wreck in an automatic machine shop can be a serious situation—could require the complete shutdown of production. To prevent such an occurrence, it is necessary to chamfer these cored locating holes, mentioned before, and to check them 100 per cent with a templet gage to make certain that the cored locating holes are in proper relationship to the outside contour of the casting so that the cylinder block will start correctly in the initial operation of the machine shop automation.

Core boxes, patterns, flasks and related equipment must be continually checked for wear and damage. A worn print in core box or pattern could mean an improperly set core. A worn flask bushing means a cope

*Manager, Quality Control Dept., Cleveland Foundry, Ford Motor Co., Berea, Ohio.

and drag shift and that also could be potential trouble. Flask bushings and pins should be checked daily with gages and those showing signs of wear should be replaced before trouble develops.

Because of close dimensional tolerances required by machine shop automation today, the control of molding sand properties has also become more important. You must have the proper control of physical properties to obtain the identical shape of the casting that the pattern designer built into the pattern equipment. Good flowability, close control of green strength and hot strength are important. Sufficient jolts, properly designed squeeze boards, constant air pressure to the molding machines, etc. are just some of the things that need to be watched very closely to make good molds. Periodic checks of the molding sand properties by the sand laboratory technician must be done and good records maintained in order that adverse trends in casting defects may be quickly observed, correlated and corrected. Swells or stickers, for example, cannot be tolerated. A swell or sticker missed by the inspector might mean a jam in the automation or possibly broken tools.

When a casting gets jammed in the automation, it means plenty of trouble. It means that either the casting or the transfer mechanism breaks. If it is the latter, then a serious delay follows until a replacement is installed or repairs are made. If it is the former, a piece of the broken casting may drop into the mechanism and time lost until it is retrieved. A casting jammed in the automation is usually a real problem to free because accessability may be limited.

To maintain accuracy in the molding department it is often necessary to set multiple cores with a fixture and then follow the core setting with a gage to make sure the cores are within acceptable tolerance in relation to the mold. With the help of a gage it is not uncommon to maintain core location to a plus or minus 0.015 in.

It has been found in some instances, that due to the design of a particular casting, warpage is found in the end product. If the warped surface is machined, it presents a problem due to the inconsistent finish stock. If the warped surface is not machined, it may present a problem on the assembly line or possibly the end product may not function properly.

In some cases it may be possible to design the patterns so that warpage in the casting is offset or at least minimized. Other times this may not be practical and 100 per cent gaging would be required to maintain acceptable tolerances.

Metal Control Mandatory

Good control of metal is very important. Cold metal promotes several kinds of defects—and, as stated before, with many operations in the machine shop tied together into what becomes a continuous operation, these defects can be costly because the whole system may be saturated by the time the scrap or repairable casting starts to show up at the finish end.

To overcome these defects related to cold metal, it is necessary to establish, set up and then rigidly enforce the correct pouring temperature ranges. Records of these readings should be maintained. Blow defects trends, for example, can usually be traced to low temperature trends. One of the best cures for many casting defects is consistently hot metal.

Metal chemistry is very important. After limits of carbon and silicon or any other elements have been established, they must be maintained and records kept.

If one or more of these elements goes out of control and soft or hard castings may be indicated, it is necessary to inform the inspection department. The inspection department having been alterted, the final inspector can intensify his brinell spot check as needed to maintain proper quality levels of the outgoing castings.

Raw Material Control

Inferior or lower-priced grades of scrap, pig, coke and other materials that go into the melting unit are false economies. Standards and specifications should be rigidly followed and all incoming materials thoroughly checked.

You all know what poor sized and low quality scrap and coke can do to your melting operations. How can you expect to make good iron if you use inferior materials,—for you get out of your cupola just what you put into it. Analysis and/or physical properties of all melting materials should be certified or checked before using. Consistency of melting materials and consistent operation will give you uniform iron and it is easier to build quality into your castings than it is to inspect defects out.

Good Machinability Required

Adequate means of determining machinability must also be used. It is much easier to put forth a little extra effort in the melting area than it is to sort out all of the castings that are too hard or too soft at the final inspection station.

In the days before automation and high speed machining, it was not nearly so harmful when a hard casting got into the machine shop. At the most it meant that a cutter might be ruined on one of several machines. Today, however, it could mean, for example, that a whole broach section goes down with hundreds of tools to be changed.

There are various ways of assuring good machining qualities. Samples of the iron can be observed as fractures and from previously determined limits, the metal inocculated as indicated by the fracture. Samples of metal must be taken at regular intervals and sent to the laboratory for analysis. Laboratory results must be reported to the melting department as soon as possible so that adverse trends can be corrected by an adjustment of the cupola charge.

In the cleaning room, it has become necessary to use automatic grinders, wherever possible, to eliminate or minimize the human element. When automatic grinders can be used it reduces dimension variance to a reasonable minimum. In some cases, a check for file hardness at critical areas of the casting may be necessary. Although a file check will only pick out the very hard to machine castings, a brinell check can be used to supplement this test and can tell accurately whether the machine shop is going to have normal tool life. In machine shop automation, short-lived and broken tools are very costly because it may be neces-

sary to stop the whole system or at least a section of it while replacing tools.

The last opportunity to keep the foundry out of trouble is in the inspection department. Layout inspection is continually in process so that trends can be observed in dimensional variations and corrections made before trouble is encountered. Castings are continually being sectioned to observe wall thicknesses, internal dimensions and to see that internal passageways are free of sand and metal penetration.

In some cases, a 100 per cent file check of critical areas on a casting is required. All castings are spot checked for hardness on brinell machines. Charts should be maintained on this type of inspection so that adverse trends can be corrected at once.

The repair of castings by welding, when permitted, must follow accepted standards and procedures, then be carefully inspected to insure a machinable weld. If utmost care has not been maintained then broken tools will result.

In the automatic transferring from station to station and the automatic positioning at various machining operations, the clearances are very close. The fins and flash in some cases must be completely removed. Even the surface finish becomes critical when the clamping mechanism will activate only by a build up of air pressure controlled by the smoothness of the area in contact with the machine locators.

The importance of good control of all foundry processes and especially final inspection cannot be over emphasized. Just consider the relationship it bears not only on the final quality of the product, but also to the proper trouble-free operation of high capacity, extensively automatic machine shop equipment in subsequent manufacturing processes.

This discussion has touched just some of the high spots but may give you some idea of how casting tolerances have become affected by automation in the machine shop.

DISCUSSION

Chairman: JAMES THOMSON, Continental Foundry & Machine Div., Blaw-Knox Co., East Chicago, Ind.

Co-Chairman: K. M. SMITH, Caterpillar Tractor Co., Peoria, Ill.

L. J. PEDICINI[1] *(Written Discussion):* The author has presented a picture of automotive foundry problems that should please the production people who have to fight this battle daily. I don't believe the problem has been at all exaggerated by Mr. Buchman. However, this presentation should indicate to automotive foundrymen that one solution of their problem in the foundry is automation. By this I do not mean automatic cycling of process equipment such as molding machines or core blowers but automatic cleaning, dipping, gaging and assembling of cores in the core room or in the foundry. The automatic grinding, chipping and gaging of castings in the finishing department and the automatic preparation of our molding sand to name a few examples.

Everyone must agree with Mr. Buchman's quotation that we cannot inspect quality into a product we must build it in. Therefore, steps must be taken to determine the limits of our foundry processes and if they are not satisfactory we must improve our processes to eliminate the maximum variation. A statistical quality control program can assist us tremendously in this effort, but it will not do more than indicate where our problem is and the correction of these problems usually requires some capital investment. We are going to have to spend sums of money equivalent to those sums spent in the machine shops for automation if we hope to keep our head above water.

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EVALUATING GREEN PROPERTIES OF CEREAL CORE BINDERS

By

E. C. Zuppann and H. Putz*

Choice of vendors for cereal core binders has been based primarily on transportation costs and secondarily on personal preference. Some laboratories have tried to make a quality evaluation based on density or green compressive strength. Neither of these two tests has agreed completely with, or satisfied, the core room personnel. The authors had long wondered how much of the lack of agreement was actual and how much imagined.

Green compressive strength has never been entirely satisfactory in evaluating one cereal as opposed to another. This test seems more adequate in comparing sand mixtures. A check of cereals over the past several years has not shown a significant difference when compared, weight for weight, using green compressive strength as a guide.

The present study uses the jolt overhang and jolt sag tests as the basis of comparison of the cereals. This test has correlated well with the coremakers' opinion of sand mixtures. Through the medium of the two jolt tests a method of evaluating cereal core binders has been evolved. The ranking of the cereals is near to that of the coreroom personnel. This work indicates that the man with his hand in the sand all day everyday has not been imagining differences in cereal binders.

The equipment and procedure used was fully described in March 1956 issue of MODERN CASTINGS. The original machine developed by the Harry W. Dietert Co. was used (Fig. 1). The specimens were blown with a bench type machine (Fig. 2) using specially designed corebox and vent pedestal (Fig. 3). The authors have standardized the procedures to give a minimum of error on successive tests.

A standard call for two samples of each regularly produced cereal was sent to ten suppliers. Six com-

*Foundry Engineer and Sand Technician, respectively, The Oliver Corp., South Bend, Ind.



Fig. 1 (Left) — Checking the Jolt overhang property.



Fig. 2 (Right) — Equipment for preparing blown jolt test specimens.

56-157

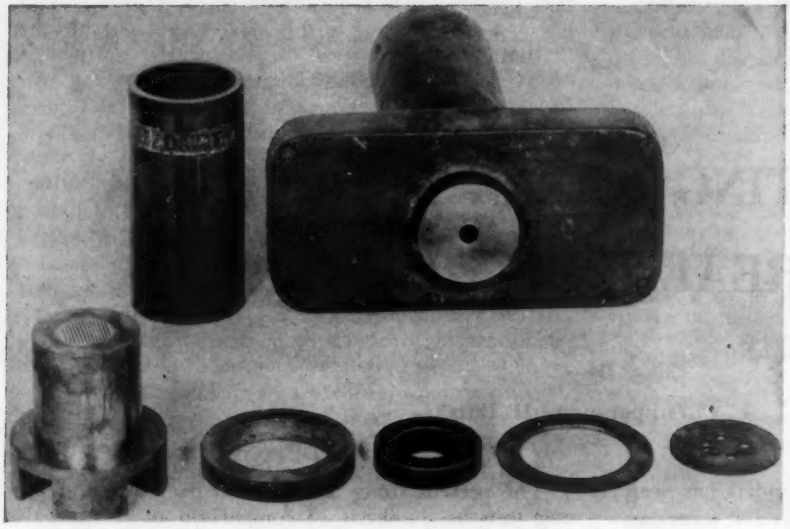


Fig. 3 — Photo of core box, cartridge and pedestal.

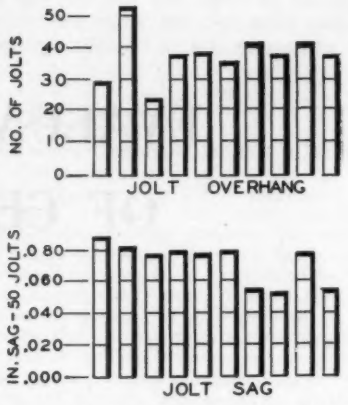


Fig. 4 (Above) — Comparison of jolt overhanging test results on ten core oils from seven producers. All sand mixtures used the same cereal binders.

panies responded with a total of 11 samples. Eight samples were heavy-weight and three light-weight cereals. Each of the two samples were tested separately and the final results averaged. Duplicate tests were run in the few cases where only one sample was submitted.

Tests on core oils have shown a marked effect upon the jolt properties. Figure 4 shows the results of jolt testing sand mixtures using the same cereal but ten different core oils from seven producers. Jolt overhang and jolt sag results are effected differently by the various core oils. The effect of three core oils on two cereals is shown in Fig. 5. Here we find that the effect of different core oils may even reverse the merits of the cereals under test. Considering the above, it was decided to conduct all tests in the presence of the core oil having jolt properties nearest the average of those shown in Fig. 4.

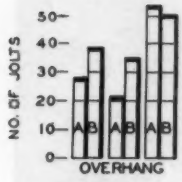


Fig. 5 (Left) — Comparison of jolt overhang results using three core oils. Each oil was tested with two cereal binders.

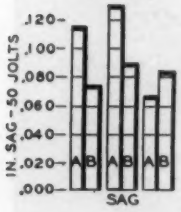


Fig. 7 (Right) — Comparison of jolt overhang test results showing the drop in test results with time. The variation in the rate of drop is shown for four core oils using the same cereal binder.



Fig. 6 — Comparison of jolt overhang test results of four laboratory mixed batches and one coreroom mixed batch. The steps indicate results of retesting the same mixture. This is only one of several such tests made.

The sand mixture selected is largely used in the Oliver core room for tractor transmission case cores. The laboratory size sand mixture batch contains:

1,000	gr	lake sand
15	gr	cereal
18 to 26	gr	water
6.6	gr	core oil

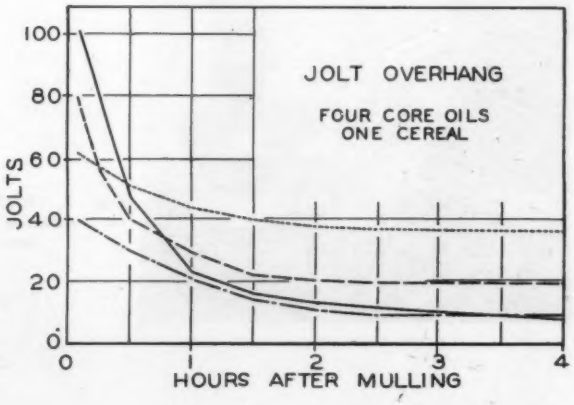
The mixing cycle was established previously to give test results that would duplicate batches from our production muller. The results of one of the final tests is shown in Fig. 6. The mixing cycle was:

One minute with sand and cereal. Add water.
Three minutes with sand, cereal and water. Add oil.
Three minutes with all ingredients.

The samples were placed in sealed cans and allowed to stand for two hours before testing.

Jolt tests run shortly after mulling give erratic results. This is due to the rapid change in properties effected by the core oil.

Figure 7 shows typical changes in overhang properties using the same cereal but different core oils. The flatter these curves the longer bench life the



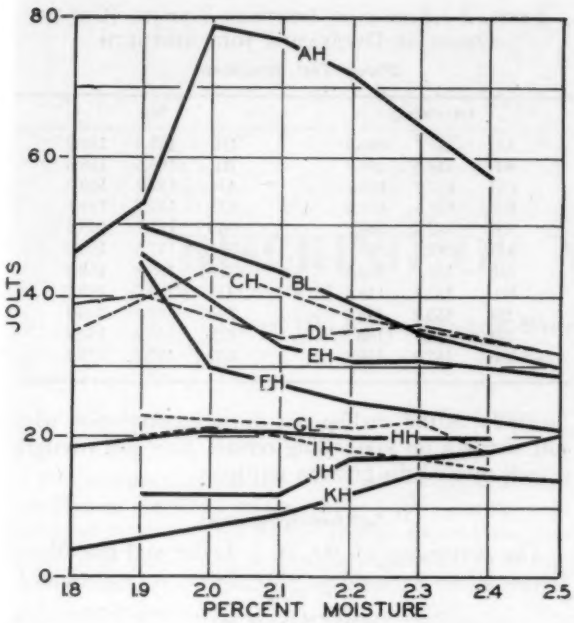


Fig. 8 — Comparison of results of the blown jolt overhang test of 11 cereal binders from six mills. The samples were tested over an extended moisture range.

mixture possesses. By waiting two hours before testing, the time and duration of testing was not nearly so critical.

The sand was blown in a cartridge-type blower. The air pressure was controlled between 97.5 and 100.0 psi. A receiver tank is just ahead of the blower to assure adequate and constant volume of air.

The jolt testing machine was placed on a very heavy steel table (used to hold the Brinell hardness testing machine.) This precaution did not appear necessary. The high cam was used in all tests. The sag test used consisted of measuring the sag in thousandths of an inch after 50 jolts.

Discussion

Results of the testing program are shown in Figs. 8 and 9. It can be seen that large differences in jolt properties were found. The various cereals are designated by a letter and an H or L for heavy weight (over 350 gr per qt) and light weight (under 350 gr per qt). All samples were designated as light weight or heavy weight by the supplier. From the two charts it can be seen that density as such is not closely related to jolt properties. However, if AHCH and GL samples are removed, those remaining show the light-weight cereals to be superior in both the jolt overhang and sag. The light-weight cereals appear more consistent among the three suppliers than do the heavy weights.

The big stumbling block is that the generalizations above do not cover AHCH and GL. In other words, one must test each cereal and not trust to generalized statements.

Of interest in Fig. 8 and Fig. 9 is that different cereals from the same mill can vary greatly. Heavy weight samples AH and FH are from the same mill as are HH and KH. Samples BL and IH; DL, HH and KH; GL and JH are from the same mills. Samples from carload lots of JH, AH and EH have been spot checked and have duplicated the results shown on the charts within 10 per cent. These checks indicate that the mills have good control of their process. However, many more such checks must be run to adequately establish the quality range of a particular mill.

Table 1 confirms an earlier finding that green compressive strength has no definite relationship to jolt properties on the same mixture. Table 2 shows that jolt properties are independent of blown density when specimens are blown in the same manner. The above statements are not to be confused with changes in density due to changes in air pressure, grain distribution of the sand, or in the oil or water.

Conclusions

1. There is a testable difference in cereals as claimed by coremakers.
2. Green compressive strength is inadequate to distinguish cereal quality.
3. The blown jolt overhang and sag tests appear to be an answer to the cereal testing problem.
4. Core oils have a marked effect on the jolt properties of cereals. Core oil must be used in testing cereals to check the degree of reaction.
5. Rapid drop in jolt properties during the first hour necessitates the waiting for two hours after mulling before testing.
6. Density of the cereal is not closely related to jolt properties.
7. Different cereals from the same mill may give radically different test results.
8. Green compressive strength is not related to jolt properties of cereal.

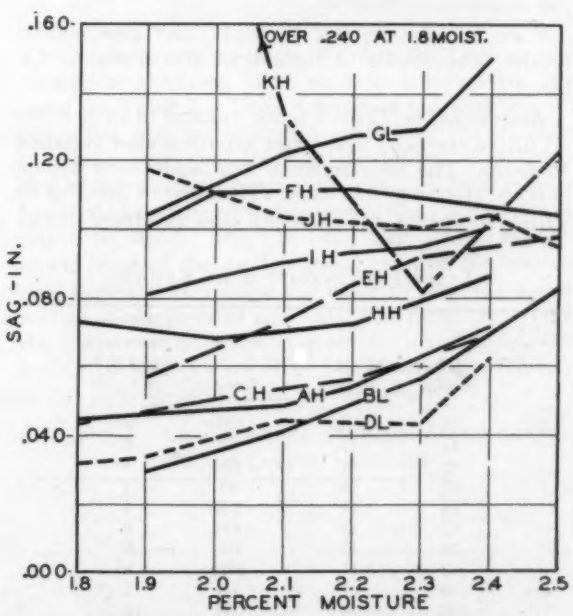


Fig. 9 — Comparison of results of the blown jolt sag test of the same cereals shown in Fig. 8.

TABLE 1 — AVERAGE GREEN COMPRESSIVE STRENGTH (PSI) IN ORDER OF DECREASING JOLT STRENGTH

Blown Test Specimens

Overhang		Sag, in./in.	
AH	0.75	DL	0.85
BL	0.77	BL	0.77
CH	0.85	AH	0.75
DL	0.85	CH	0.85
EH	0.80	HH	0.72
FH	0.60	EH	0.80
GL	0.73	IH	0.75
HH	0.72	JH	0.85
IH	0.75	FH	0.60
JH	0.85	GL	0.73
KH	0.73	KH	0.73

TABLE 2 — RANGE OF SPECIMEN WEIGHT (GR) IN ORDER OF DECREASING JOLT STRENGTH

Blown Test Specimens

Overhang			Sag		
AH	157.7	160.0	DL	157.5	157.9
BL	158.5	158.8	BL	158.5	158.8
CH	157.7	158.1	AH	157.7	160.0
DL	157.5	157.9	CH	157.7	158.1
EH	157.5	158.2	HH	157.7	158.4
FH	157.5	159.1	EH	157.5	158.2
GL	157.5	158.1	IH	158.2	158.9
HH	157.7	158.4	JH	157.7	158.3
IH	158.2	158.9	FH	157.5	159.1
JH	157.7	158.3	GL	157.5	158.1
KH	157.0	158.1	KH	157.0	158.1

9. The viscosity of a slurry of cereal and water seems to be closely related to jolt strength.
10. A 16-hour settling test showed that low jolt strength cereals settled the most.

Other Tests

Several chemical and physical tests were attempted to ascertain their degree of correlation with the jolt test results. Three simple tests showed high degrees of correlation.

One test was the Zahn viscosity of a slurry. A mixture of 20 grams of cereal and 200 milliliters of distilled water was mixed for 30 seconds in a large beaker with a high speed mixer. Bubbles were allowed to rise out for 20 minutes. The viscosity was checked with a Zahn viscosimeter using the number 3 cup. This instrument is not very accurate with mixtures having suspended materials and gave only a rough approximation of the viscosity. Stirring the mixtures and visually ranking the samples gave the same viscosity test results. More suitable equipment is needed for accurate determination of the viscosity-jolt strength relationship.

A second viscosity test 16 hours later gave similar results. The results of these tests are shown in Table 3.

Also shown in Table 3 is the amount of clear water on top of the flour and water mixtures after standing 16 hours. The measurement was made in a 250-ml beaker. Here one specimen, FH, is out of line due to jelling similar to AH from the same producer. It can be said that the settling test shows a correlation with jolt strength on non-jelling cereals. Low jolt strength cereals showed the greatest settling.

TABLE 3 — VISCOSITY TEST RESULTS

Zahn Sec, Vis. 20 Min	16 Hr	Sample	Thickness of water layer after 16 hr in 1/16 in.
15	solid	AH	0
14	17	BL	2
12	14	CH	4
12	15	DL	4
12	14	EH	4
12	12	FH	0
10	12	GL	6
9	11	HH	8
9	11	IH	7
8	9	JH	10
8	8	KH	11

Acknowledgment

The permission of Mr. M. J. Lefler and the Oliver Corporation to publish this paper is appreciated.

DISCUSSION

Chairman: CHARLES WILLIAMSON, Columbia Malleable Iron Co., Columbia, Pa.

Co-Chairman: R. H. OLMSTED, Whitehead Bros. Co., New York.

Secretary: D. C. WILLIAMS, Ohio State University, Columbus.

A. C. FERGUSON[1]: Was there a difference found in cereals after they had been standing in storage?

MR. ZUPPANN: Apparently not because the samples were used over a period of six months.

PAUL RENEGAR[2]: Core oils make considerable change in properties and the kind of oil must be considered especially the relationship of properties to oil viscosity.

MR. ZUPPANN: A thin oil gave best overhang properties. The drop off of the jolt is somewhat a function of viscosity of oil. Oil detracts from the overhang properties of a mixture.

V. M. ROWELL[3]: Did you vary the mixing cycle?

MR. ZUPPANN: Yes. and we used the Greco-Latin Square statistical procedure. Variation between cereals was more important then variation in mulling times. We recommend that foundries make use of statistical procedures.

H. W. DIETERT[4]: Foundries making cores with large overhang have difficulty which can be minimized by reducing the mineral spirit content of the oil.

O. J. MYERS[5]: Were all the cereal binders which were used in these tests products of corn?

MR. ZUPPANN: Yes, they were.

MR. MYERS: What was the ratio of water to cereal in the viscosity test?

MR. ZUPPANN: The ratio was 100 grams of cereal to 1000 grams of water and other quantities of cereal to the same amount of water.

MR. MYERS: You reported a correlation between viscosity and overhang. Was this expected with oil additions?

MR. ZUPPANN: No oil was added in viscosity test. Variations in jolt overhang were much greater between mixtures having oil or no oil than was found between mixtures containing different oils.

PAUL TEMPFLIN[6]: Why was the standard set at 50 jolts?

MR. ZUPPANN: We started using this test early when it was advocated for routine tests that 50 jolts be used.

MR. DIETERT: The jolt overhang test is the outgrowth of AFS Core Test Committee work.

1. Morris Bean Co., Yellow Springs, Ohio.
2. McKinley Iron Works, Fort Worth, Texas.
3. Harry W. Dietert Co., Detroit.
4. Harry W. Dietert Co., Detroit.
5. Archer-Daniels-Midland Co., Minneapolis.
6. J. R. Short Canadian Mills Ltd., Toronto, Ontario, Canada.

RISERING OF NODULAR IRONS*

Part III — The Effect of Pouring Temperature on Shrink Depth

By

R. C. Shnay** and S. L. Gertsman***

It is known that the size and depth of the shrinkage cavities in gray-iron castings vary with pouring temperature.[1,2,3] Because of the greater risering requirements and higher cost of inoculants, a similar pouring temperature effect in nodular irons is of even greater practical importance.

During the course of an earlier investigation,[4] the authors found that consistent results, using risers that were just adequate, could only be obtained by controlling the pouring temperature within relatively narrow limits. Any deviation from this range of pouring temperatures resulted in under-riser shrinkage. A further indication of such a pouring temperature effect appears in a paper by Flinn, Rees and Spindler,[5] who found that a certain range of pouring temperatures was "preferable for sound castings."

The primary purpose of this investigation was to determine how the shrink depth in nodular iron castings is affected by changes in pouring temperature. Tests were carried out in green-sand molds with and without riser insulation. Dry-sand molds were also used to determine the effect of mold rigidity with changes in pouring temperature. In addition, cooling rate curves and metallographic studies were carried out in an attempt to determine the underlying causes of the pouring temperature effect.

Experimental Methods

The melting and inoculation procedures used in this investigation were the same as those described in previous reports.[4,6] Briefly the procedure was to melt a charge of pig iron, wash metal, and steel scrap in a 500-lb acid-lined induction furnace. The metal was superheated to about 2730 F and allowed to cool to 2640 F before tapping. A ferrosilicon magnesium alloy was used as the nodularizing inoculant, followed by a post inoculation with ferrosilicon.

Five or more molds were poured at successively lower temperatures ranging from 2500 F to 2265 F. The variation in pouring temperature was obtained by pouring the molds at intervals of 2 to 4 minutes. In general, the difference in pouring temperature between molds was at least 35 F.

The molds each contained two cylindrical test castings 3½ in. in diameter and 3½ in. high. The gating and risering arrangement is shown in Fig. 1.

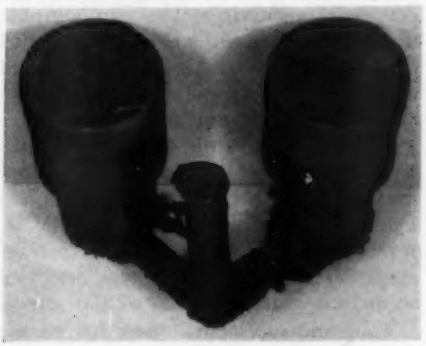


Fig. 1—Test castings; risers and gates still attached.

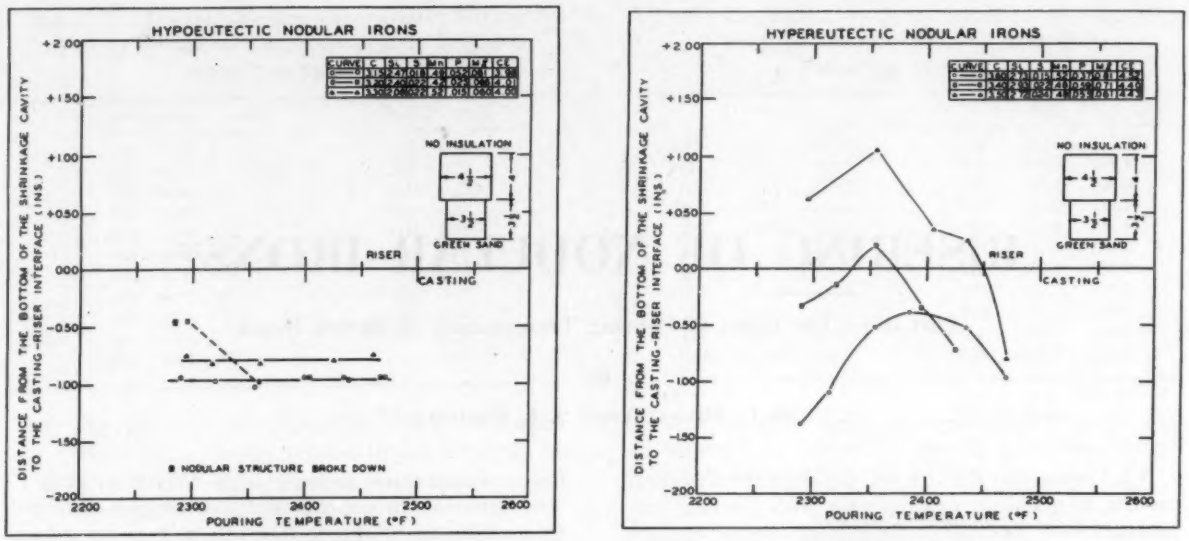
The various combinations of riser dimensions and insulation thickness used in this investigation are listed in Table 1. These combinations were chosen since they produce marginal conditions. In all cases rice hulls were used for riser insulation. The thickness of the rice hull cover was controlled by leaving a space of the required height above the riser cavity as shown in Fig. 6. The rice hulls were added as soon as the level of the molten metal reached the desired riser height. No attempt was made to keep the upper surface of the layer of rice hulls flat; therefore, where the insulation thickness is given as, say, 1½ in., the actual thickness varied from 1½ in. at the perimeter of the layer to about 2 in. at the center.

TABLE 1 — COMBINATIONS OF RISER DIMENSIONS AND INSULATION THICKNESS

Riser Diameter (in.)	Riser Height (in.)	Insulation Thickness (in.)
4½	4	Nil
4½	4	1½
4½	3	3

*Published by permission of the Deputy Minister, Department of Mines and Technical Surveys, Ottawa, Canada.

**Metallurgist and ***Head, Ferrous Section, Physical Metallurgy Division, Mines Branch, Department of Mines and Technical Surveys, Ottawa, Canada.



(a) Hypo-eutectic Irons *(b) Hypereutectic Irons*

Fig. 2 — Effect of pouring temperature on shrink depth of nodular iron test castings. Riser height, 4 in.; no insulation.

The castings with the risers still attached were sectioned through the gates and along the longitudinal axis. Each half casting was radiographed to 1½ per cent sensitivity. The radiographic plates were used for the determination of the shrink depth. The distance from the casting-riser interface to the bottom of the shrinkage cavity was measured for each half casting. These measurements were made to the nearest 1/20 in. on the x-ray plate and assigned positive or negative values, depending on whether the bottom of the shrinkage cavity was located above or below the casting-riser interface; i.e. a positive sign was used to indicate a sound casting. It must be emphasized that the shrinkage cavity was considered to include spongy areas as well as pipes or gross voids.

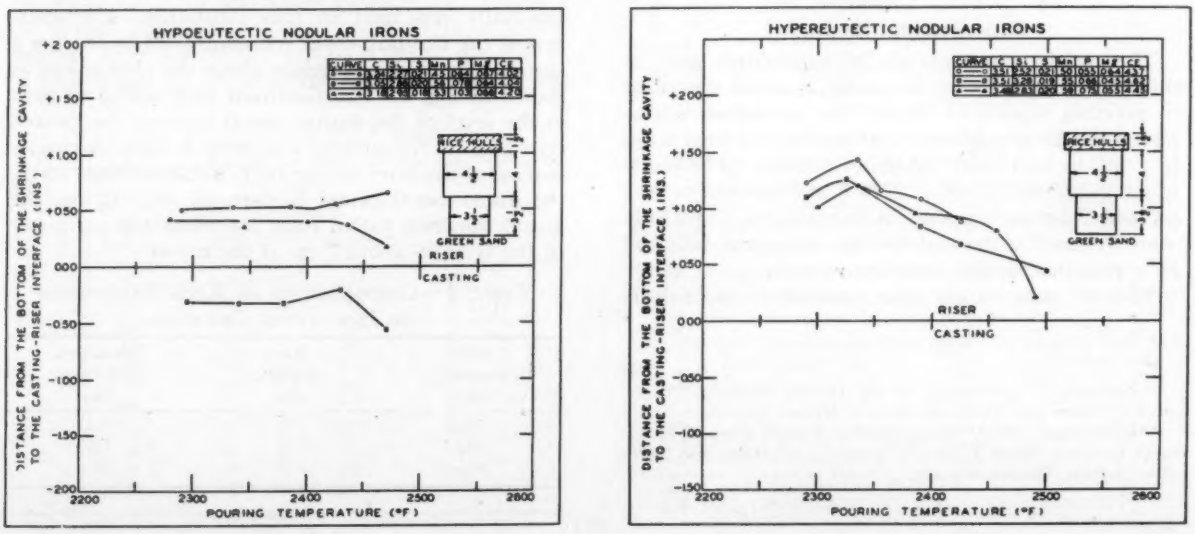
Four measurements (two for each casting) were obtained for each pouring temperature. The lowest of these was taken to represent the shrink depth at that particular pouring temperature. The samples for carbon determination were prepared in the form of small pencils, as described in previous reports.[4,6] All the samples were removed from the center of the casting, at least one inch below the bottom of the shrinkage area. The data from any heat showing an abnormal microstructure were not included in the general results.

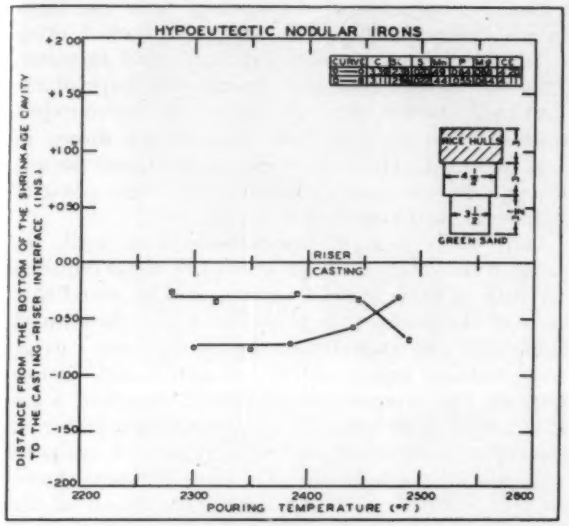
Results

(a) Tests With Green Sand Molds

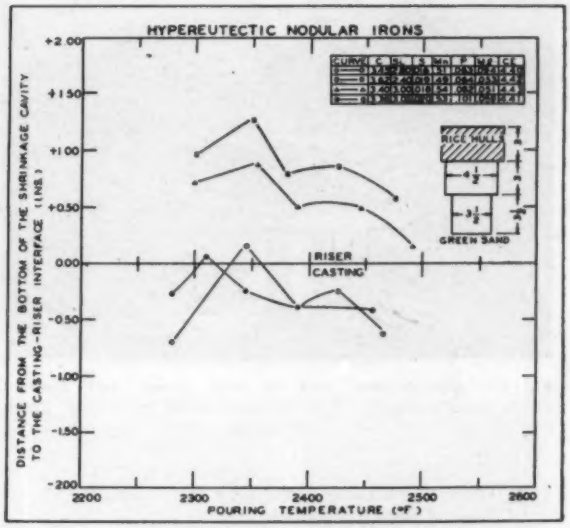
Two series of heats were carried out for each combination of riser height and insulation thickness. Hypo-eutectic compositions were used in the first series and hypereutectic in the second. The compositions within each series were varied so that a broad

Fig. 3 — Effect of pouring temperature on shrink depth of nodular iron castings. Riser height, 4 in.; insulation thickness, 1½ in.
(a) Hypo-eutectic Irons *(b) Hypereutectic Irons*





(a) Hypo-eutectic Irons



(b) Hypereutectic Irons

Fig. 4 — Effect of pouring temperature on shrink depth of nodular iron castings. Riser height, 3 in.; insulation thickness, 3 in.

range of carbon and silicon contents were covered, with the carbon equivalent remaining relatively constant. The molds were rammed by hand using a New Jersey No. 70 sand to which the following additions were made:

8 per cent flour
4 per cent western bentonite
4 per cent sea coal
1 per cent cereal

Typical properties of the sand mix are given in Table 2.

The results of these tests are illustrated in Figs. 2 to 4, inclusive.

(b) Tests With Dry Sand Molds

In order to reduce the effect of mold wall movement, a series of tests were carried out using dry-sand molds. The gating arrangement, the flask size, and the melting and inoculation procedures were the same as those used with green-sand molds. The same sand was used but with the following additions:

4 per cent western bentonite
4 per cent silica flour
1 per cent cereal

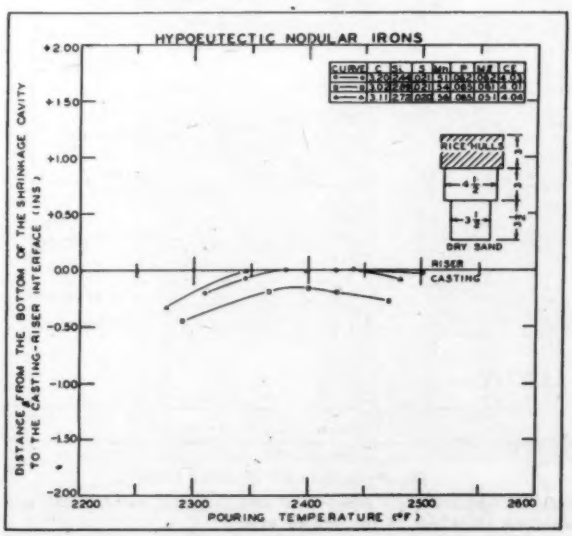
The results of these tests are shown in Fig. 5.

TABLE 2 — TYPICAL PROPERTIES OF THE GREEN-SAND MIX

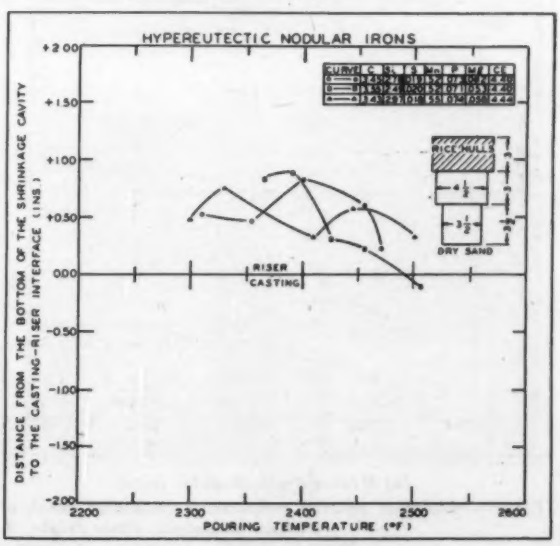
Permeability	85
Green compressive strength	5.6 psi
Green tensile strength	9.0 oz/sq in.
Green deformation	47.5 psi
Dry compressive strength	175 psi
Moisture	4.8%

Fig. 5 — Effect of pouring temperature on shrink depth of nodular iron castings. Riser height, 4 in.; insulation thickness, 1½ in.

(a) Hypo-eutectic Nodular Irons



(b) Hypereutectic Nodular Irons



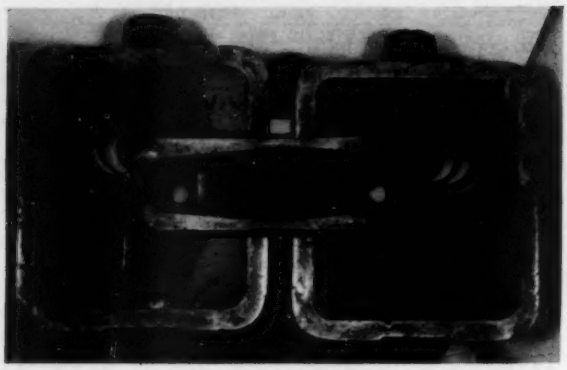


Fig. 6 — Arrangement used to pour green- and dry-sand molds simultaneously. The dry-sand mold is at left; green-sand mold at the right.

(c) Tests With Green and Dry Sand Molds Poured Simultaneously

Although the data shown in Figs. 4 and 5 are for the same riser height and insulation thickness, the effects due to the use of dry-sand molds (Fig. 5) are difficult to assess by comparing the two sets of data. This is due to fluctuations in chemical composition and other foundry variables. This difficulty was overcome by using a common pouring basin, so that green- and dry-sand molds could be poured simultaneously (see Fig. 6). The risers were 3 in. high and a 3-in. layer of rice hulls was applied to one casting in each mold. Four castings were poured at each temperature. Two of the castings were molded in green sand and two in dry sand. One of the two green-sand castings had rice hulls and the other was left uncovered. Similarly, one of each of the two dry-sand castings also had rice hulls and the other was left uncovered. The results of these tests are shown in Fig. 7.

One half of each of the castings from the heats shown in Figs. 7 (a) and 7 (b) was macro-etched, using a 1:3 solution of concentrated nitric acid in water. This was done to illustrate the size and shape of the shrinkage cavities. Photographs of the macro-etched castings with the risers still attached are shown in Figs. 8 and 9. Horizontal lines were drawn on the macro-etched sections to indicate the lowest point to which the shrinkage area extended.

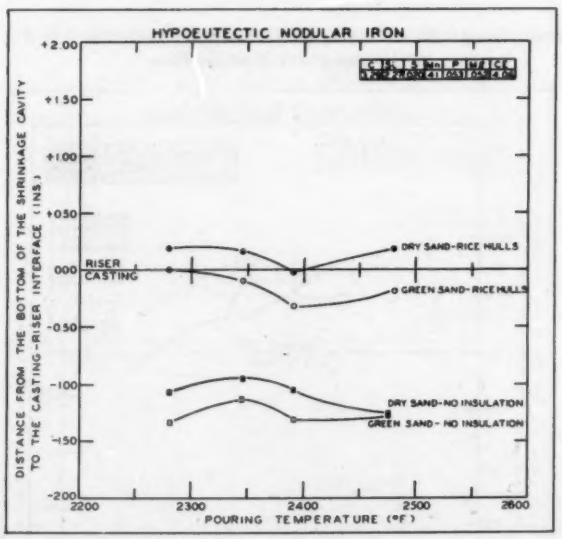
Only four pouring temperatures were used, because of the relatively large amount of metal required for each of these tests. It should also be noted that each of the points shown in Fig. 7 was determined from only two measurements (i.e., only one casting was produced under each set of conditions). Consequently, the curves shown in Figs. 2, 3, 4 and 5 are likely to be more precise than those shown in Fig. 7, since four measurements (two castings) were available for each set of conditions. However, the tests shown in Fig. 7 supply a more reliable comparison of the general advantages obtained by the use of dry-sand molds and riser insulation.

(d) Reproducibility Tests

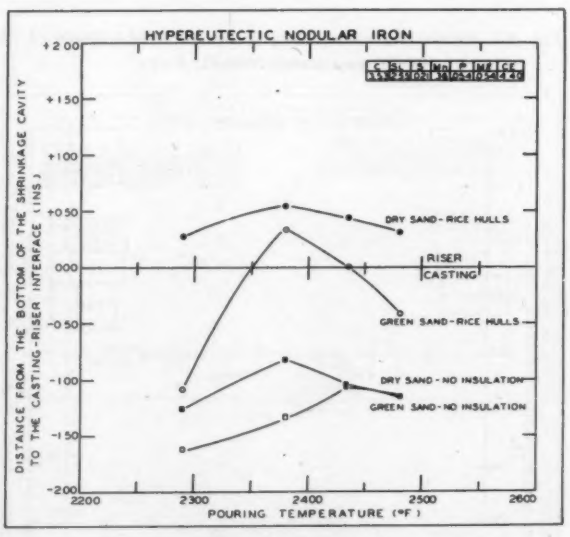
The reproducibility of the tests reported in section (a) and illustrated in Figs. 2, 3 and 4 was assessed by carrying out three successive tests with all variables held as constant as possible. Three-inch risers were used with three inches of rice hulls since this combination showed the greatest amount of variability in the previous tests (Fig. 4). The results of these tests was presented in Fig. 10, which shows a band extending over the full range of values.

(e) Tests With Small Castings

These tests were carried out to determine whether the pouring temperature effect noted in Figs. 2(b), 3(b) and 4(b) would be found in smaller castings. These might be produced in volume in a foundry and



(a) Hypo-eutectic Nodular Irons



(b) Hypereutectic Nodular Irons

Fig. 7 — Effect of pouring temperature on shrink depth of nodular iron test castings; green-sand and dry-sand molds, with and without insulation. Riser height, 3 in.; insulation thickness, 3 in. (where used).

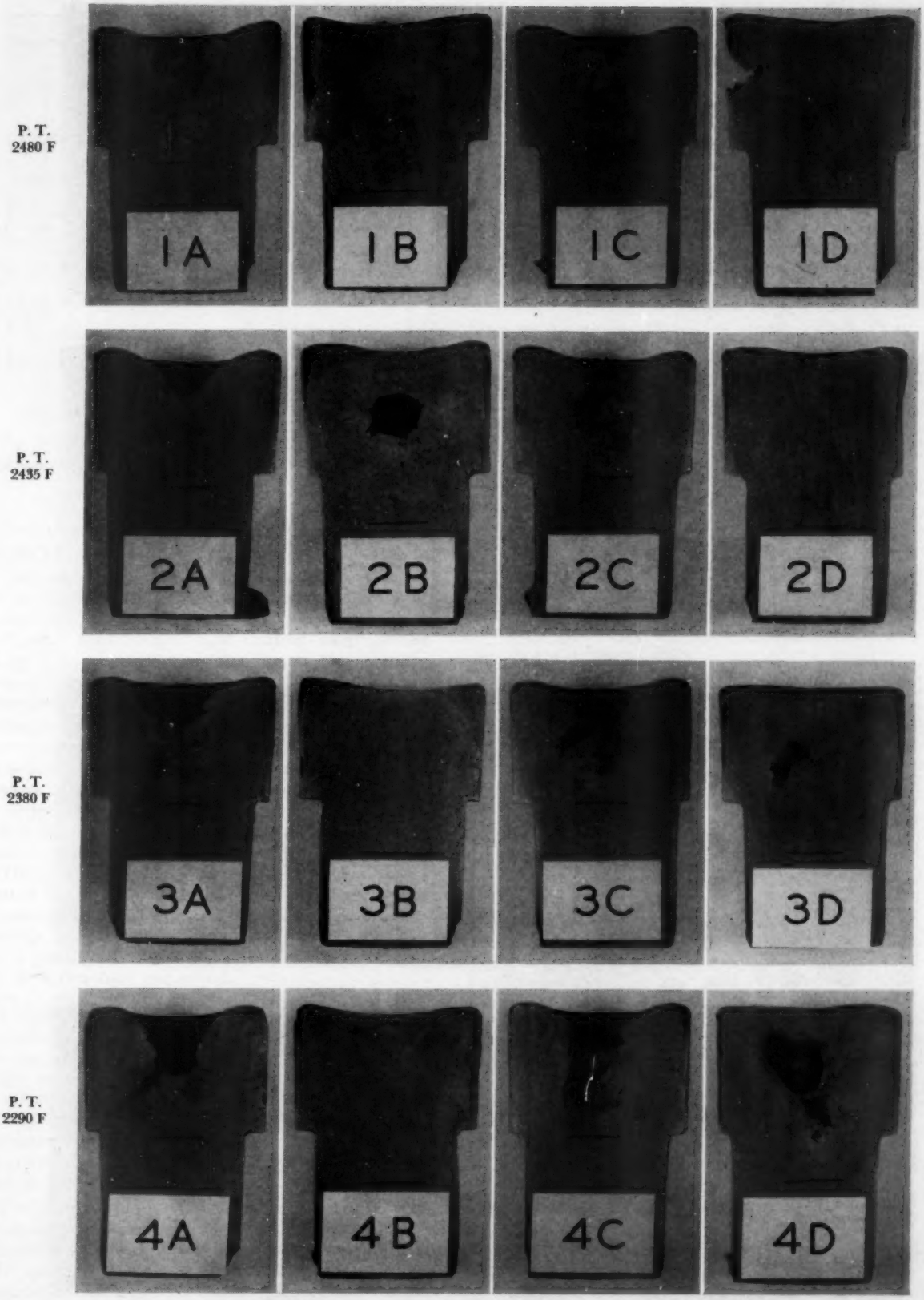


Fig. 8 — Macro-etched sections of hypo-eutectic castings. Etchant: 1:3 HNO_3 in water

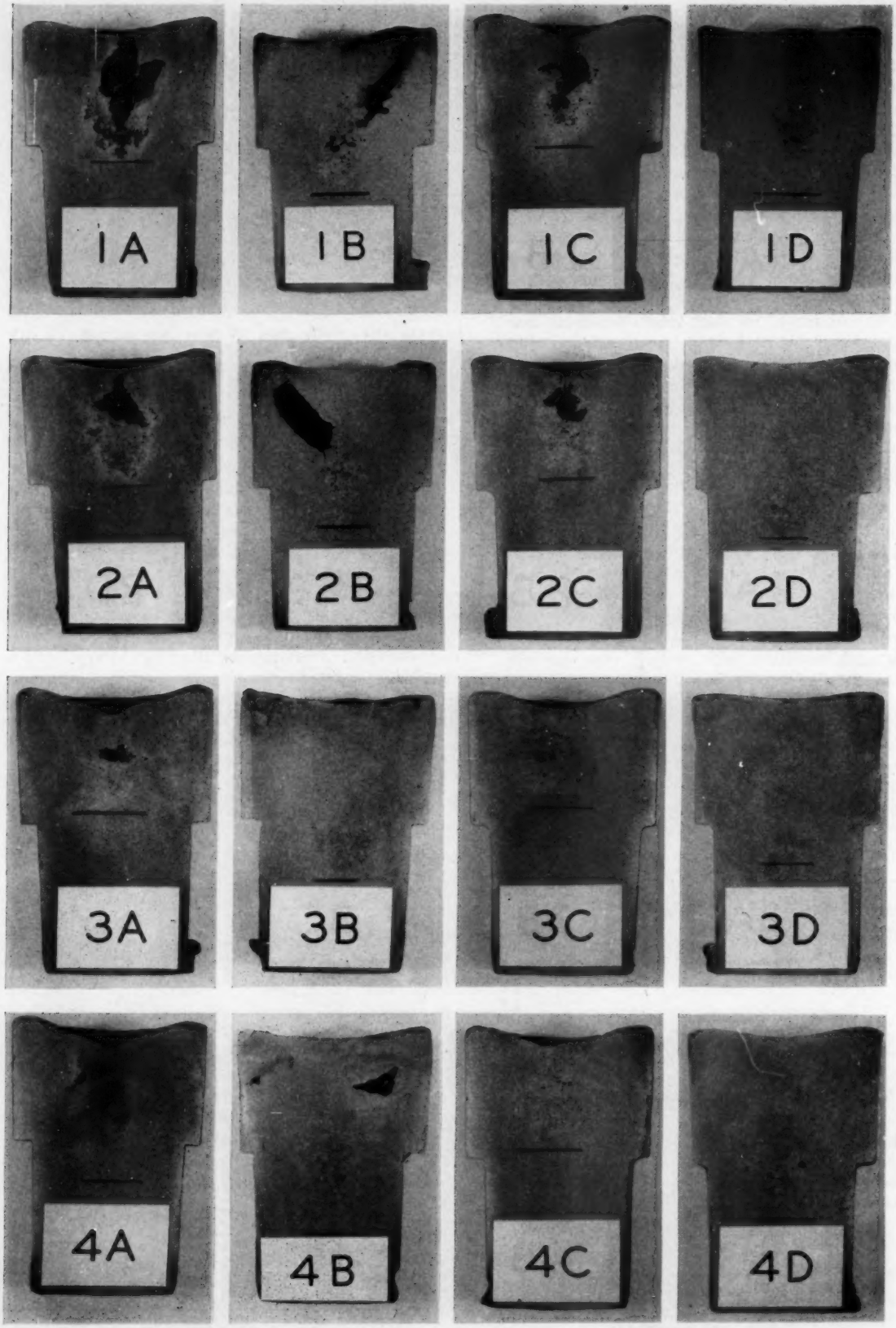


Fig. 9 — Macro-etched sections of hypereutectic castings. Etchant: 1:3 HNO_3 in water

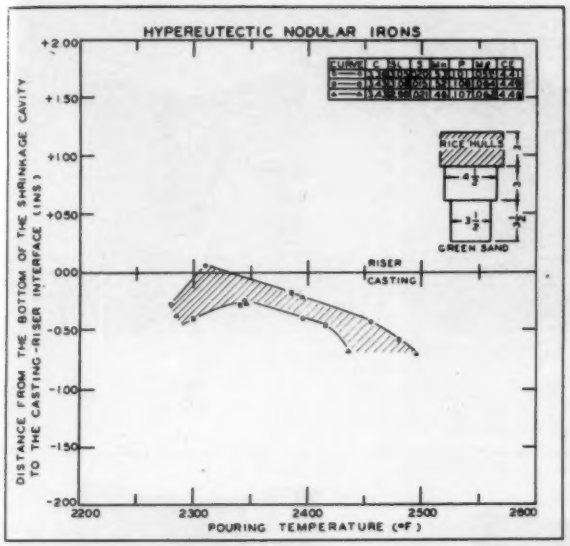


Fig. 10 — Results of reproducibility tests. The two lines are drawn to indicate range of results.

a wide range of pouring temperature might occur during any one heat.

The castings were in the form of round plates, 4 in. in diameter and 1 in. thick. The risers were 3 in. in diameter by 3 in. high and a 1½ in. layer of rice hulls was used. Eight such castings were poured using a hypereutectic nodular iron. The pouring temperatures ranged from 2480 F to 2280 F. A microstructure typical of these castings is shown in Fig. 11.

Sections were cut from the center of each casting (including the riser) and macro-etched to show the shape and the position of the shrinkage cavities. The sections are shown in Fig. 12, numbered from 1 to 8 in order of decreasing pouring temperature. Horizontal lines were again drawn to mark the lowest point to which the shrinkage had extended.

The shrinkage area extended below the riser in casting No. 1 and casting No. 2. The third casting appears to be a borderline case whereas the next two castings, No. 4 and No. 5, are not sound. Castings No. 6 and No. 7 are sound while the last casting, No. 8, is definitely not sound.

(f) Cooling Rate Studies

Cooling rate studies were carried out with standard test castings (3½ x 3½ cylinders), using a riser height of 4½ in. and a 1½-in. layer of rice hulls. A quartz-sheathed thermocouple was inserted across one of the cavities in each mold so that the bead was located in the center of the horizontal plane at the casting-riser interface. The pouring procedure was similar to that used in the previous tests.

The temperature of the metal in the ladle was measured at regular intervals by means of an immersion thermocouple. A multi-point recorder with a chart speed of 0.5 in. per minute recorded the temperature at each mold thermocouple. It was found that readings of the metal temperature could be obtained within 30 seconds of the time pouring had begun. The resulting cooling curves are shown in Figs. 13(a) and 13(b).

All the test castings were sectioned to check the final position of the thermocouples. The positions of the shrinkage areas in each pair of castings poured at the same temperature were also compared, to determine whether the presence of the thermocouple had affected the shrink depth.

It was found that the thermocouples had not shifted during the pouring and solidification of the castings. In only one instance did the shrink appear to have been affected by the thermocouple; this was in the first casting of the hypereutectic series.

Discussion

(a) Shrink Depth

The beneficial effect of using riser insulation is well illustrated in Figs. 7, 8 and 9. In every case the shrink depth is greatly decreased by using a rice hull cover.

When the risers are not covered the effect of pouring temperature on the shrink depth of the hypoeutectic nodular irons appears to be negligible (Fig. 2(a)). The addition of rice hulls to the riser surface affects the shrink depth of the castings poured at the highest temperatures as shown in Fig. 3(a) and 4(a). When the carbon content is high and the silicon content low, the shrinkage cavities in the castings poured at the highest temperatures do not extend as deep as in the castings poured at the lower temperatures. The reverse is true where the carbon content is low and the silicon content is high.

The variation of shrink depth with pouring temperature is much greater in the hypereutectic nodular irons. In every case a critical pouring temperature was found where the shrink depth was a minimum. This is shown by the peaks in the curves of Fig. 2(b), 3(b) and 4(b). This critical pouring temperature appeared to lie between 2325 F and 2375 F. When the pouring temperature was either above or below this range, shrinkage extended deeper into the casting. In some of the curves an inflection point appears

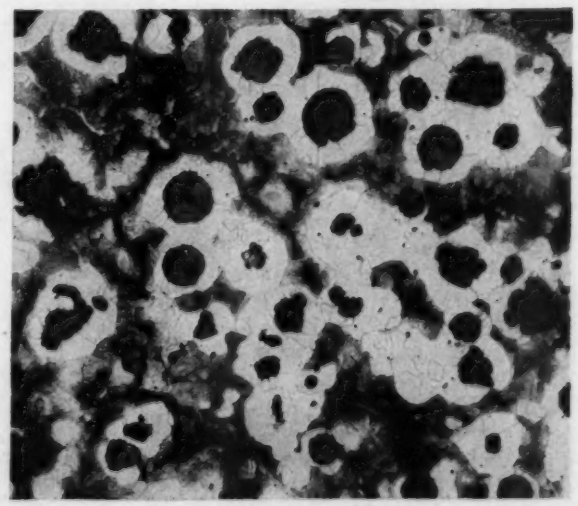


Fig. 11 — Typical microstructure of small castings. Etchant: 2% nital. Mag. — 100X.

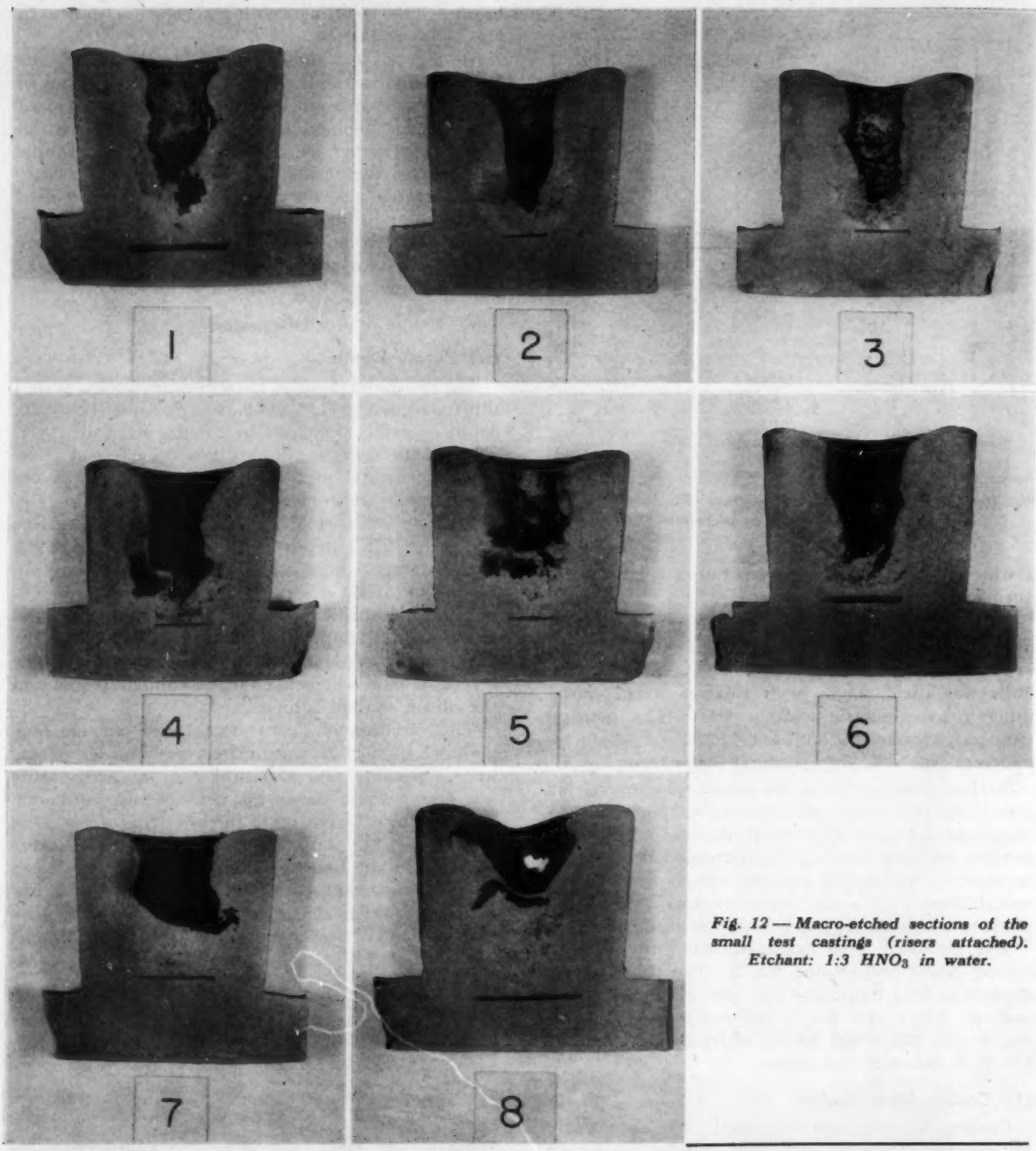
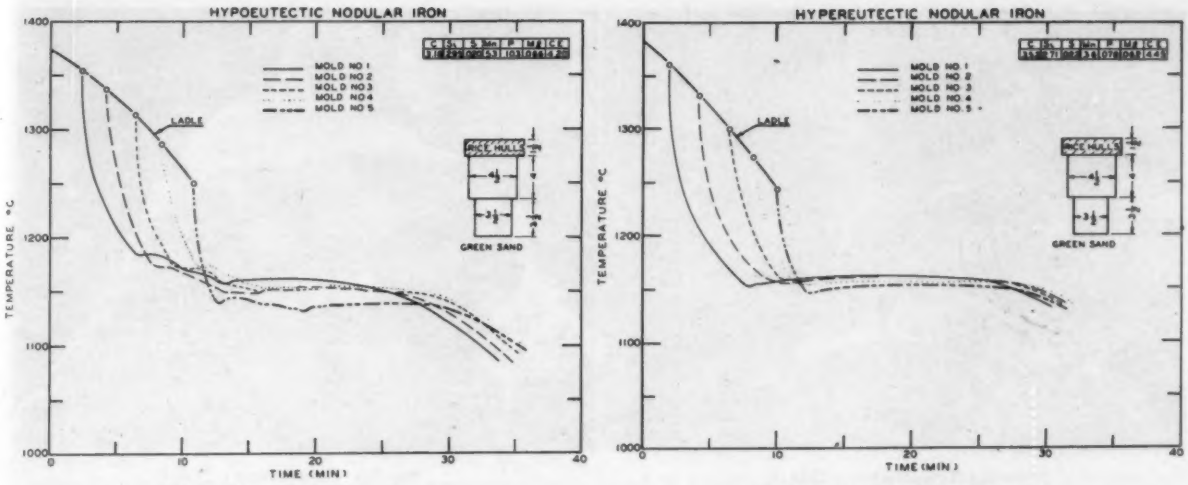


Fig. 12 — Macro-etched sections of the small test castings (risers attached). Etchant: 1:3 HNO_3 in water.

to exist between 2350 F and 2400 F where the trend seems to flatten out. This inflection is also indicated in the tests with the small castings (see casting No. 4, Fig. 12).

There are considerable differences in level between the various curves within any one series shown in Figs. 2 to 5. These differences are undoubtedly due either to the changes in metal composition or to fluctuations in the properties of molding sand. Although all the pertinent data were examined, no consistent relationships were found. This does not mean that such relationships do not exist, but rather that not enough data are available to define them. However, it is significant that the greatest amount of variation in the general level of the shrink depth curves was found in the case which was the closest to being marginal (i.e., the 3-in. riser with 3 in. of rice hulls).

The macro-etched sections in Figs. 8 and 9 are typical of the castings produced in this investigation. The gross shrinkage or pipes often appeared to decrease in depth with decreasing pouring temperature. However, the location of the spongy areas does not follow the same trend. This is particularly well illus-



(a) Hypo-eutectic Nodular Irons *(b) Hypereutectic Nodular Irons*

Fig. 13 — Cooling rate curves of nodular iron castings poured at different temperatures. The curves represent cooling conditions at the center of the casting-riser interface.

trated in the macro-etched sections shown in Fig. 12. Since the amount of liquid shrinkage decreases with decreasing pouring temperature, the gross cavities may be expected to decrease in size as the pouring temperature decreases.

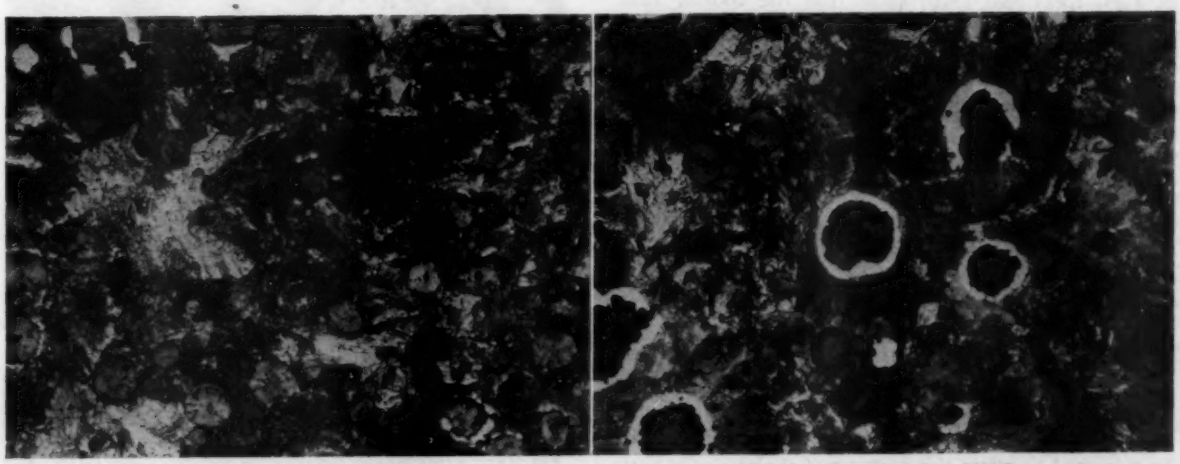
(b) Solidification Characteristics

Castings 1B and 1D in Fig. 8 show shrinkage cavities at the riser walls. This was a frequent occurrence with castings poured at high temperatures, and indicates that the wall was not completely frozen before solidification was well advanced in the body of the riser. This is an indication of the low degree of progressive solidification in nodular iron which has been reported by previous authors.[8,9]

The effect of changes in pouring temperature on the cooling rate of metal is illustrated in Fig. 13(a) and 13(b). The cooling curves in these figures show that the cooling rate was highest at the lowest pouring temperature. However, this effect seemed to diminish after the eutectic temperature was reached. A lower pouring temperature is usually associated with a higher degree of chill, but experiments carried out by Williams[10] have shown that the reverse is true in gray iron. Williams obtained the variation in pouring temperature by using different holding times in the ladle, and therefore the effects attributed to pouring temperature may have been really caused, or at least influenced, by the holding time. Somewhat different results were observed in the present investigation. Metallographic samples taken from the same point* in the castings from several heats showed an increase in the amount of ferrite and in the size of the graphite nodules as the pouring temperature was decreased. These trends are illustrated in Fig. 14.

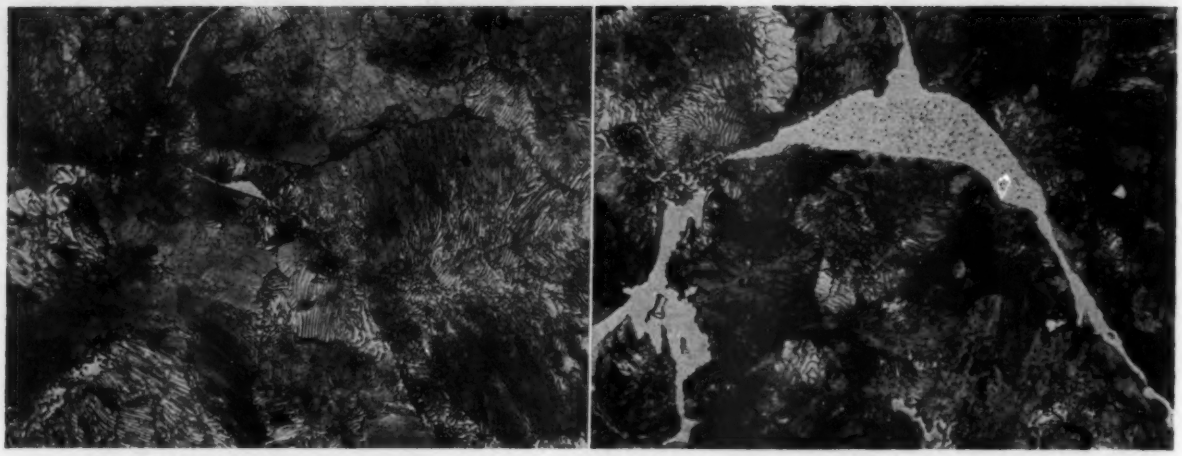
Decreasing the pouring temperatures also appeared to produce an increase in the size and occurrence of carbide particles (see Fig. 15). In addition, the tendency for carbides to form at the pearlite grain boundaries seemed to increase as the pouring temperature was lowered. This effect is illustrated in Fig. 16.

*The samples were taken from a point 1 in. below the shrinkage area and along the center-line of the castings.



Pouring Temperature — 2425 F Pouring Temperature — 2280 F

Fig. 14 — Structures of castings poured at different temperatures. Heat 2836; C. E. — 4.37. Etchant: 2% nital. Mag. — 100X.



Pouring Temperature – 2335 F Pouring Temperature – 2280 F

Fig. 15 — Photomicrographs showing the typical carbide particles. Heat 2836; C. E. — 4.37. Etchant: 4% picral. Mag. — 500X.

These differences in microstructure can be attributed to either the difference in pouring temperature, the effect of holding time, or possibly a combination of both factors. In any case it is significant that the increase in shrink depth in the hypereutectic castings is associated with changes in the microstructure. This indicates that the pouring temperature effect cannot be considered solely as a heat transfer problem. The fact that the shrink depth in hypereutectic nodular iron castings is affected by changes in pouring temperature to a much greater extent than in hypo-eutectic castings is a further indication that some complex metallurgical phenomena are involved.

Morrogh[12] has postulated two different mechanisms for the solidification of nodular iron. When the cooling rate is low, graphite nodules are said to form in the liquid between the dendrites. Under conditions of fast cooling, the liquid remote from existing nodules is said to freeze as an "acicular carbide eutectic" which may graphitize after solidification. These mechanisms can occur in hypo-eutectic as well as hypereutectic nodular irons. The cooling curves of Figs. 13(a) and 13(b) show that the cooling rate from the ladle to the eutectic temperature was highest at the lowest pouring temperatures and therefore the "acicular carbide eutectic" could be expected to form at these pouring temperatures. This appears to be substantiated by the microstructures shown in Figs. 15 and 16.

(c) Effect of Holding Time

Effect of holding time in the ladle has been reported in papers by Reynolds and Taylor[13] and Okamoto and Yoda.[14] In the first paper a decrease in magnesium content with holding time was reported. In the second paper the authors reported that the carbon and sulfur contents decrease during the first 5 minutes and 15 minutes, respectively, and then increase. Such variations in composition could be expected to affect the amount of shrinkage in castings poured after extended holding times.

(d) Test with Dry-Sand Molds

The shrink depth was decreased at all pouring temperatures when dry-sand molds were used. However, the pouring temperature effect was still present, although somewhat modified. This indicates that mold rigidity is not a major factor so far as the effect of pouring temperature is concerned; that is, although the general level of the shrink depth is somewhat higher than dry-sand molds are used, it still varies with the pouring temperature. It is possible that some of the differences between the curves shown in Figs. 4 and 5 are in part due to the absence of sea coal in the dry-sand mixture.

(e) General Discussion

The decrease in shrink depth when using dry-sand molds, and the advantages of using an insulating cover, are well known and have been reported by other authors.

In this discussion no attempt has been made to



Fig. 16 — Photomicrograph showing the formation of carbides at pearlite grain boundaries. Heat 2935; C. E. — 4.42 Etchant: 2% nital. Mag. — 100X.

present a full explanation of the effect of pouring temperature on shrink depth, but the authors have tried to bring out all relevant facts which could lead to such an explanation. There is some doubt as to whether the effects attributed to changes in pouring temperature are actually not due to differences in holding time. However, variations in pouring temperature encountered in practice are usually due to excessive holding times in the ladle. Therefore this conceptual error is probably not of great practical importance.

The most important practical application of the results of this investigation, would be in the volume production of a small casting where a large number of molds are poured from one ladle. Under these conditions there would be an appreciable decrease in pouring temperature during the operation and excessive fluctuations in shrink depth could be encountered. The results of this work indicate that risering conditions should be such as to care for any pouring temperature variations which might occur in the foundry operation.

Conclusions

1. The shrink depth of hypereutectic nodular iron castings varies greatly with the pouring temperature. Small variations in the pouring temperature can make all the difference between producing a sound or an unsound casting.

2. In the case of hypereutectic nodular irons there is a critical pouring temperature, above or below which the shrink depth increases. In this investigation the critical pouring temperature was found to lie between 2325 F and 2375 F.

3. Variations in pouring temperature have a minor effect on the shrink depth of hypo-eutectic nodular irons.

4. The effects of variations in pouring temperature are not eliminated by using dry-sand molds.

5. The use of rice hulls as an insulating cover decreases the pouring temperature effect in hypereutectic nodular irons.

6. When the risering is marginal there is a high degree of variation in the general level of the shrink depth.

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DISCUSSION

Chairman: CHARLES LOCKE, Westlectric Castings, Inc., Los Angeles, Calif.

Co-Chairman: DR. W. J. CHILDS, Lafayette College, Easton, Pa.

J. B. CAINE[1] *(Written Discussion):* The authors have reported on a significant decrease in shrinkage between hypo and hypereutectic nodular iron. It should be pointed out that simply increasing carbon and silicon to decrease shrinkage is not an unmixed blessing. Unless these hypereutectic irons are controlled very closely, flotation of graphite as nodules will occur. The test castings used in this investigation do not show this condition as they are covered with a top riser and any flotation rises into the riser and is not trapped in the casting.

Flotation of hypereutectic nodular iron becomes serious in production castings when the excess graphite cannot float into a riser and is trapped at the cope surface. If this discusser had to choose, he would prefer shrinkage in the center of the section to flotation at the surface as being less detrimental to the serviceability of the casting.

This discusser would like to continue the conjectures as to why the peculiar humped relation between shrinkage and pouring temperature. Such a relation must mean that the shrinkage is effected by at least two opposed variables. It is suggested that temperature gradients are the prime variable: that these temperature gradients are effected by two opposed variables caused by the riser and the gate.

Bishop, Pellini and coworkers at the Naval Research Laboratories have shown that gross shrinkage voids like associated with cast steel are formed when the temperature gradients are high. Dispersed shrinkage voids like associated with cast bronze are formed when the temperature gradients are low. The casting sections of Figs. 8 and 9 show both types in varying amounts.

It is perhaps significant that the most efficient feeding occurred with insulated risers when the uninsulated risers were the least efficient. If it can be assumed that at this point the temperature gradients of the uninsulated castings were at a minimum (causing all dispersed shrinkage) this indicates that the heat loss through the uninsulated open top of the riser about balanced the temperature gradients caused by the bottom gate at this particular pouring temperature.

Continuing this conjector, as pouring temperature is increased the amount of heat available in the riser increases faster than the heat input into the sand due to the metal flow as the mold is filled. This seems reasonable, but not proven. The castings of Figs. 8 and 9 indicate that this may be true, for all show an increase in gross shrinkage cavities with higher pouring temperatures. This can mean that with increasing superheat causing more available heat in the riser, temperature gradients due to the riser are increasing at a greater rate than the opposed temperature gradients due to the bottom gate.

Conversely, at low pouring temperatures the gate temperature gradients that are affected by variables such as rate of pour, minor variables as far as the riser is concerned, overshadow the riser temperature gradients to the point of adversely affecting riser efficiency. If this set of conditions is

1. Consultant, Cincinnati, Ohio.

operative, there is a point of intersection of two curves of different slopes. This point of intersection would show up as a hump in the author's charts.

It must be re-emphasized that this theory is pure conjecture. It is advanced for consideration by the authors. If such a mechanism can be proven it will be more satisfactory than explanation based on suspicion of slight changes in composition and much more satisfactory than the old whipping boy—mold wall movement. Mold wall movement is significant in gray iron due to the unique solidification of this metal. Its extension to other metals causes, in the writer's opinion only confusion.

MR. SHNAY *(Reply To Discussion By Mr. Caine):* The authors certainly agree with Mr. Cain that graphite flotation in nodular iron is a more serious problem than shrinkage. However, no evidence of flotation was observed in any of the castings produced during the course of this work. There was always an area of dispersed or poor nodule structure in the region immediately surrounding a gross shrinkage cavity, but this is not what is usually meant by graphite flotation. It has been our experience in other work that when there was a layer of high graphite concentration near the top of a casting or riser it was noticeable on either a deep-etched or freshly cut surface and also in a radiograph. None of these indications were noted in the castings prepared for this project, even though carbon equivalents as high as 4.62 were used.

During the course of previous investigations, some trouble was encountered with graphite flotation. This was usually associated with very large gross shrinkage cavities. It was found that this trouble could be avoided by reducing the proportion of graphitic carbon in the charge. However, what is true for an induction furnace such as used in the author's laboratory may not be true for cupola melting.

Mr. Caine's conjectures regarding the reason for the presence of a peak in the curves of shrink depth vs. pouring temperature are very provocative and warrant further study. It is true that the presence of such a peak or hump indicates that the shrink depth is affected by at least two opposed or non-parallel factors. However an analysis in terms of heat transfer alone is questionable, since the peaks do not occur in the curves for the hypoeutectic irons. If the shape of the curve depends only on heat transfer, it should be the same for all metals and alloys. Therefore the mechanism proposed by Mr. Caine could only be part of the answer.

Other considerations such as differences in the solidification mechanisms of hypoeutectic and hypereutectic compositions as well as between nodular irons poured at different temperatures must also be considered. Since some expansion is known to occur during the solidification of nodular iron, it is difficult to eliminate considerations of the physical properties of the mold wall from any discussion of this subject.

V. PASCHKIS[2] *(Written Discussion):* The authors present a wealth of information which will have to be taken into consideration when a comprehensive and systematic study of under riser shrinkage is undertaken. Such study would, it seems to the writer, have to coordinate findings in several fields, all of which may interact at least to some extent: *the heat transfer pattern,* establishing "isochrones i.e. lines connecting points in casting, neck and riser which reach a certain temperature simultaneously (e.g. isochrones for the liquidus, the solidus, etc.); *the fluid flow pattern,* taking into consideration the flow of a fluid with non-constant viscosity under a varying ferrostatic head; *the strength of the materials,* both casting and mold, during the solidification process, since this strength influences the progress of solidification; *the metallurgy of the iron,* which may change during solidification.

In view of the complexity of this problem, it is understandable that such comprehensive study has not as yet been undertaken; but lacking such study, experiments necessarily can only supply data to verify a later theory based on those several fields.

The writer has studied the paper mainly from a heat transfer viewpoint, and remarks offered hereafter are made in clear recognition that this is but one of the four fields involved.

Comparing Figs. 2-4 it appears as if both for hypoeutectic and for hypereutectic iron there is an optimum covering: shrinkage seems to increase as the insulation thickness goes from 1½ to 3 in. Since larger insulation thickness would probably mean later solidification of the riser, this seems baffling. But rice hulls, used as insulation in these experiments, would probably burn. Is the 3-in. layer so thick that less combustion takes place (does oxygen reach those parts of the covering which are above the ignition temperature)?

That there is a consistent difference between castings in dry and in green sand (Figs. 6 and 7) is surprising for two reasons:

(a) supposedly both casting and riser are surrounded by sand of the same conditions. The shrinkage should be explainable by the relative solidification time of casting and riser; and the two should be influenced in the same direction by any changes in sand.

(b) for steel the presence of water in the green sand seems to influence the apparent conductivity only for a very short time after pouring[1, 2]. Unless conditions are much different at the lower temperatures used for nodular iron, the marked difference between casting in green and in dry sand reported by the authors cannot be explained on thermal grounds.

The proper utilization of a riser depends, as Caine[3] has pointed out, on the relative solidification times of casting and riser. Many of the remarks in this discussion are based on the assumption of a sand (mold) thickness sufficiently large so that the outside surface of the sand and the flask do not heat at all or not significantly. It would be interesting to know the thickness of sand, both surrounding the casting and the riser. If the flask has no offsets, then the sand thickness of the riser is ½ in. smaller than that for the casting. And yet the riser should freeze after the casting. Thinner sand at the riser tends to produce the opposite effect (riser freezing before the casting).

Based on Fig. 6, it would look as if the cavity were not in the center of the mold. Moreover the sand thickness between two adjacent castings seems exceedingly small, and also the thickness between the riser and the flask appears to be only 1 in. or less (unless the proportions of the photo are deceiving).

If a few general remarks are permitted, the writer would like to suggest:

1) In papers describing castings, illustrations (such, as in this paper, the inserts in Figs. 2-4) showing casting dimensions should always contain not only casting, gates, and riser but also the surrounding mold.

2) A study should be made of the advantage, if any, of preheating the cavity for the riser prior to casting. It is probable that such preheating might permit the use of smaller risers.

3) The thickness of sand used in a mold should have some relationship to the expected solidification time and the pouring temperature.

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MR. SHNAY *(Reply To Discussion By Dr. Paschkis):* The authors recognize the complexity of this problem and as Dr. Paschkis pointed out a complete study should embrace several fields. However, the primary purpose of the work under discussion was to determine if a pouring temperature effect existed. Having established this, the authors then attempted to at least indicate the various factors involved.

The particular combinations of riser height and insulation thickness were chosen since they represented marginal conditions. A study of the pouring temperature effect under other than marginal conditions would have had little practical meaning. Dr. Paschkis has compared Figs. 2 to 4 with a view to determining an optimum thickness of rice hulls. Since the data in Fig. 3 are for a riser height of 3 in. and the data in Fig. 4 for a riser height of 4 in. such a comparison is not valid. However, it should be pointed out that the

2. Columbia University, New York.

3-in. layer was completely burnt through. Some exploratory tests with the 4-in. risers showed that increasing the thickness of the rice hull layer up to 3 in. produced only a slight improvement.

The consistent difference in shrink depth between the castings poured in green sand and those poured in dry sand is due mainly to differences in mold wall movement. A direct heat transfer approach has limited value in this case.

The flask sizes were chosen so that the thickness of the sand between the risers and the flask wall was at least 2-in. There was a layer of at least 3 in. of sand between adjacent risers and 4 in. between adjacent castings. It is very doubtful if the outer surfaces of the flasks were heated sufficiently to introduce any appreciable effect on solidification.

The heat due to the adjacent riser may have had some influence but in every case the spongy areas are centrally located in the casting or riser.

Figure 6 may be somewhat misleading since the upper cavities shown in the picture were those prepared for the rice hulls. Since these cavities were 6 in. in diameter the distance between them is only 1 in. as compared to 3 in. between the risers (4-in. diameter). The main mold cavities are not centrally located in one direction since space had to be left for sprues and runnners.

The authors agree with Dr. Paschkis that the thickness of the sand in a mold should have some relationship to the expected solidification time and the pouring temperature. However, since molding flasks are commonly available only in certain standard sizes, the practical foundryman can only choose the smallest flask that will do the job. The use of larger flasks than absolutely necessary, results in a sharp increase in costs.

GREEN-SAND CASTING FINISH

By

R. F. Meader*

How many foundrymen are satisfied with their casting finish? How many think they *cannot* do better? How many think they do not *need* to do any better to hold their present business? Because a casting may be hidden from view is no reason to try to "get by" with a poor finish. From the standpoint of pride of workmanship alone we should endeavor to get the best finish possible, and we should accomplish it in the molding and not by grinding.

Competition may force us to produce better casting finish if we are to retain green-sand casting business. Everyone knows of sand-molded jobs that are now die castings and the newest process, shell molding, has already acquired some choice squeezer jobs. Why? What do these precision or semi-precision methods have that we do not have in green sand? First is surface appearance, and second is dimensional tolerance.

We can improve our casting finish and when we do we will also produce castings truer to pattern. Such castings will fit into the jig properly so that all subsequent operations will have only sufficient machining finish, thus saving both material and machine shop time. Usually many of the same factors governing finish will also exert a strong influence on the dimensional accuracy of a casting; so whatever you do to improve surface appearance will usually provide closer dimensional tolerance.

Casting buyers are naturally influenced somewhat by appearance. If the surface is nice then the tendency is to expect that the internal structure is of acceptable quality. A foundry management that is careful to produce a good finish is almost certain to be just as careful in producing castings of proper chemical analysis and physical specifications. There probably is no foundryman who does not consider appearance when he looks at a casting.

Sand Grain Size and Distribution

The most important factor governing surface finish is grain size. The finer the sand grains the smoother the finish—within certain limits. Many foundrymen like to use a sand that is on the coarse side because then they can take considerable liberties with moisture content, hard ramming, etc., and still get castings, although the finish will not be as good as you or your customer might desire. On the other hand, the use of fine sands will definitely require more attention and closer control on your part.

Somebody always asks: "How fine a sand can be used with safety?" Sand fineness is probably dictated by the permeability needed for various section thicknesses and the moisture content of the sand. As a rough guess, ½-in. section castings of iron should not require more than 20 to 25 permeability, ¾-in. sections about 35 to 45, and 1-in. sections approximately 60 to 70. For the best finish choose a sand with as few as possible of the larger size grains. They will not help permeability and definitely will roughen the finish. If you want to open up a fine sand, you must add at least 30 per cent coarser sand to affect the permeability. Infortunately for the finish, the coarser sand grains will definitely roughen the finish before noticeably affecting permeability.

Grain distribution is a more important factor than we are apt to realize. This shows the amount of grains of each different size from the finest to the coarsest in any particular sand and these amounts are usually expressed in percentages. From these percentages is figured a kind of an average which is called the Grain Fineness Number of that sand. Thus we have an easy way of describing a sand, such as a GFN 50 sand, a GFN 110, or a GFN 150 sand, but we should always remember that these numbers are only an average. They will not tell you, for example, that a sand may have a lot of large grains which may be harmful to finish, or that there may be too many fines which may cut the permeability too low. If you want to get a complete picture of a sand examine the screen analysis. Better still, plot on graph paper the per cent by weight of sand retained on each sieve.

The Grain Fineness Number may be what you think you want but there may be several things wrong with the sand. It may be a sharp-peaked sand with too many grains on a few sieves which will encourage rattails and buckles. It may have too many coarse grains which will produce rough castings. Or, a sand may

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have too many grains on the fine screens and the pan; this will mean a close sand with a low permeability which will not stand hard ramming.

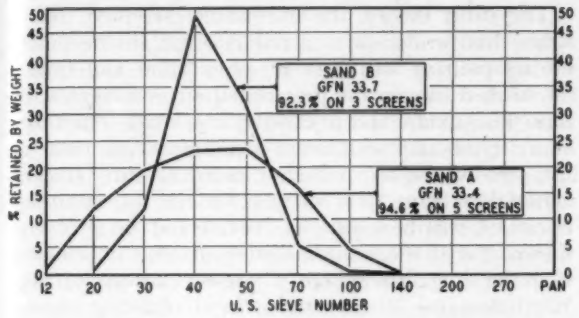


Figure 1

Figure 1 shows graphs of two sands, both with a Grain Fineness No. of 33. Sand "B" (Fig. 1) was intended for a long run slinger job (green sand). Permeability was good and the job had to be rammed hard, about 75-80 mold hardness. We made beautiful molds but the castings were terrible with plenty of expansion scabs. We put in additives such as sea coal, pitch, wood flour, and cereal, but we were not entirely successful. We continued to have too many rejected castings.

The basic trouble was that this sand was a sharp-peaked sand with 92 per cent of the grains on three screens—much too uniform. Sand "A" (Fig. 1) made a great improvement. It is a wider spread sand with 95 per cent of the grains on five screens. If you are apt to have expansion troubles when you ram hard be sure to use a wide spread sand. You may still have to use some additives but not as much as with a sharp-peaked sand.

The author is cognizant of Dietert's article "No Need for Rat-Tails" in the October, 1955, issue of MODERN CASTINGS Magazine which reports that AFS Committee (8-J) found no apparent difference in rat-tailing between a 1-screen sand, a 3-screen sand, or a 7-screen sand. In other words grain distribution cannot be correlated with rat-tailing. However, in the experience I related we believed we got better results with the wide screen sand. When you buy a sand, do not pay too much attention to the Grain Fineness Number, but look at the grain distribution.

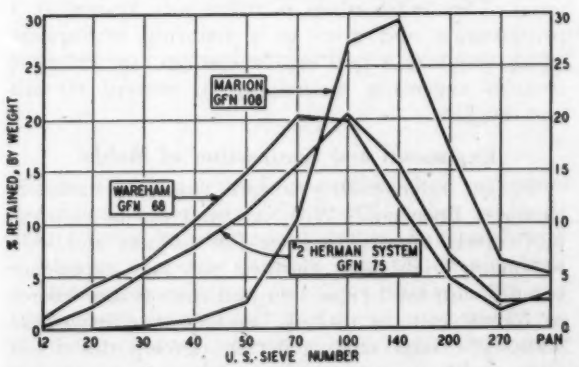


Figure 2

Figure 2 shows the grain distribution of the synthetic sand used in making large castings also of the Wareham and Marion sands that make up the mixture. We blend these two sands together to obtain a 4-screen distribution which we believe helps in controlling expansion in hard rammed molds. The system sand has a Grain Fineness Number of about 75. Moisture is kept at 4 to 4.5 per cent. Green compressive strength is from 9 to 11 psi. Permeability will be about 50; Total Combustibles from 5.4 to 6.0 per cent.

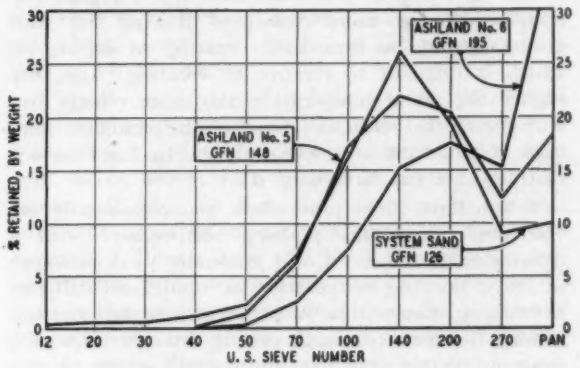


Figure 3

For our squeezer system we blend two fine sands, both local (Fig. 3). One has a Grain Fineness Number of about 148 and the other nearly 200. The system, however, seldom runs finer than GFN 125 due to our dust exhaust system being too efficient. We have under consideration a plan for reducing the exhaust without causing dusty air conditions. Where the permeability now runs 35 to 45 we would like to cut it down about 10 points to improve our casting finish.

Mold Hardness vs Finish

Mold hardness is a "must" for good finish. It is rarely possible to obtain good finish with a soft-rammed mold and close dimensional tolerance is out of the question. The author likes to see mold hardness of at least 75, and generally 85 maximum. Machine made molds will usually have higher mold hardness than hand-rammed molds. The day has passed when you check mold hardness with your finger tips pressed against the mold. Get a mold hardness indicator; they are inexpensive and a molding foreman should have one and use it frequently.

In squeezer molding machines get an extra large squeeze piston to develop maximum mold hardness. A 10-in. diameter squeeze piston at 80 lb air pressure will develop a total squeeze pressure of about 3 tons. On a 12-in. x 18-in. mold this represents a pressure of 29 psi. A 13-in. piston on the same flask will exert 5 tons, or about 49 psi. The author knows a foundry using a 16-in. piston to develop 8 tons pressure, equivalent to 74 psi. This is to get close dimensional tolerance and extra good finish. Competition from shell molding has started a trend toward higher squeeze pressures, and if the sand is fine enough the finish will be almost as good as that obtained with shell molding.

Mold hardness has a great influence on the tendency of a sand to exhibit rat-tailing. The harder a mold is rammed the greater the tendency toward expansion difficulties. However, do not reduce mold hardness if you run into rat-tails or buckles.

Molten Metal Temperature vs Finish

Temperature of molten metal will affect casting finish. As temperature of metal is increased so is fluidity. This will tend to cause more penetration between and around sand grains and produce rough finish. In the case of ferrous metals, the higher the temperature the more chance of fluxing the sand grains, forming an iron silicate coating on the casting which is difficult to remove by blasting. Also, the higher the metal temperature the more closely you will get to the refractory limit of the sand-clay mixture. If you have ever seen sand grains fused onto a casting, then you have had "it."

If you have large jobs where you consistently get "burn-on" check the pouring temperature with a pyrometer at the mold and gradually work down to as low a pouring temperature as is safe and still run the casting. You will be surprised at how well you can reduce "burn-on" on large castings by controlling the pouring temperature. At the author's plant we frequently make some 1-ton pulleys, three parted, and gated at the bottom into a loose ring. These castings are about 60 in. in diameter, approximately 40 in. deep, with a ⅝-in. section. Even though we are careful to face the ring with a strong facing we get a bad "burn-on" in front of each gate if we pour above 2600 F. This seems to be the refractory limit of our particular sand-clay-facing on that casting.

Use of Facing Additives

Facing additives contribute greatly to casting finish by producing peel and preventing "burn-on". But it is not too well understood why and how they work. The author believes that if we understand their action we can use facings to greater advantage. There are several theories that might be mentioned.

Some foundrymen believe that molten metal changes the sea coal, pitch or other carbonaceous material in the facing, from a solid to a gas containing carbon, and that this carbon is deposited on the face of the mold as a soot to form a thin refractory or insulating coating of carbon which prevents the metal from wetting the sand grains and reacting with them to form a rough surface on the casting. It is also thought that the carbon layer lubricates the mold metal interface. Molten metal, if at a sufficiently high temperature, will soften and flux the sand grains; but this is much less likely to happen if there has been a carbon layer deposited on the sand grains. It is said that molten metal cannot "wet" the silica grains, which it must *first* do before any reaction can take place between the metal and the sand grains.

Another theory, which is plausible, is that the molten metal vaporizes the facing material to form an envelope of gas at the mold-metal interface, and theoretically prevents actual contact between the metal and the sand grains, thus producing a good finish. For this theory to work, there must be sufficient back pressure of the gas to maintain such an envelop or layer until the metal starts to solidify.

The third theory, the one widely accepted today states that formation of a reducing gas in the mold during pouring will prevent penetration and burn-on. Molten iron in the presence of air or oxygen will form iron oxide, and at 2550 F iron oxide can exist as a liquid and will attack sand grains to form a thin, glassy, slag-like material called fayalite. If this material adheres to a casting surface the cleaning operation will be expensive. Taylor and Savage have shown that this condition cannot develop in a mold if a reducing atmosphere is present instead of air. What does the scientist mean by a reducing atmosphere and how can we produce it if it is so beneficial? We might define a reducing atmosphere as a gas that is extremely hungry for oxygen.

When a mold faced with a carbonaceous material like sea coal or pitch is poured, there is formed immediately a gas rich in hydrogen and carbon monoxide. Both gases will unite chemically with any available oxygen in the mold, thus leaving no free oxygen to unite with the molten metal to form iron oxide. In addition to sea coal and pitch, cereals and wood flours are a few other materials which will form reducing atmospheres. If you run high moistures in green sands you are almost certain to produce an oxidizing atmosphere which enhances making of iron oxide and you do not want this to happen.

So much for theory. In practice do not use more facing than is necessary to properly cover the parts of the pattern that need it. As to strength of facing, do not make it stronger than experience or necessity dictate. We check the strength by having our laboratory determine the amount of total combustible material. If you do not have a laboratory then watch your castings. If they are too shiny, cut down on the strength; if the castings do not clean readily, increase the strength. We have three standard facings and the total combustible material runs from about 1¾ per cent for light castings to about 5½ per cent for the heavier section castings (not over 1 in.).

Our squeezer sands contain practically no facing material. Two other systems for medium and larger castings contain a patented facing material which is a combination of a hydrocarbon and a cellulose material. The hydrocarbon is technically known as a pyrobitumen and gives us a reducing atmosphere which produces a peel on the castings; the cellulose controls expansion by helping to prevent rat-tails and buckles.

Expansion and Contraction of Molds

To get finer casting surface you will naturally choose a finer sand. With certain types of castings, particularly those with large flat surfaces and with reasonably high mold hardness you may experience trouble with sand expansion and contraction, known to foundrymen as rat-tails, buckles, and expansion scabs. The finer sands generally develop this defect more readily than the coarse sands because the smaller

grains heat up faster than the larger grains. Also, the coarser sands tend to conduct the heat away from the surface by allowing better passage of the hot gases. Hard rammed molds will be more apt to have these defects than those rammed softer.

The basic reason for this class of rejects is that as molten metal enters a mold the heat causes the sand grains to expand and increase in volume. If the mold has been rammed hard to bring the sand grains close together it is evident that as the grains expand, they have no place to go except into the mold cavity. Minor expansion will cause only a mild cracking of the mold surface; this is a rat-tail. Greater expansion will cause the surface of the mold to crack wide open and allow the molten metal to enter the crack and fill the area behind the expanded surface layer of sand. This is called a "buckle" and frequently can be chipped off, but usually leaves a depression in the casting surface.

We have read about expansion of sand grains and we talk glibly about it, but I wonder how much we really know. Sand, as we know it, is principally silica. It is extremely plentiful and is the basis of practically all of our molding sands. It is therefore almost entirely indispensable in the foundry industry. Yet, for all its advantages, it has one big disadvantage —its expansion under heat. When quartz, a pure form of silica, is heated, the expansion curve (Sosman's curve) looks like that in Fig. 4. There is a steady expansion with increasing temperature until at 1065 F there occurs a sudden expansion which may be about where molding sand goes through its worst phase of rat-tailing and buckling.

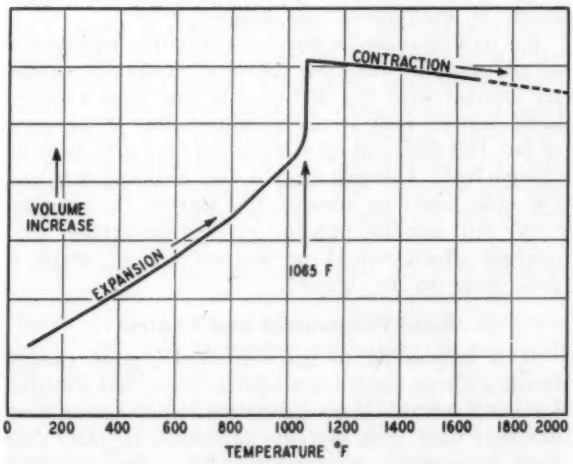


Figure 4

After the sand grains have reached their maximum expansion at around 1200 or 1300 F they gradually contract with increasing temperature. At about 1500 F (roughly) the clay starts to contract. If the total contraction is sufficient, shrinkage cracks in the sand surface may develop and produce veining. Sometimes the layer of sand underneath the surface layer may still be in the expansion stage; and if this is sufficiently strong it may push aside the surface layer and allow metal to run in and form what is known as a contraction scab. It would be very fortunate if the expansion of the silica grains took place at the same time the clay was contracting so that one action could neutralize the other; but this does not happen because the temperature ranges are not the same. What are the usual remedies to prevent these difficulties?

Most naturally-bonded molding sands are comparatively free from expansion difficulties because they have "built-in" additives which we do not have in the synthetic or formulated sands unless we deliberately add them to the mix. So one of the remedies is to use a natural sand.

I think it is advisable to have a wide spread sand according to the grain distribution. I like to see at least 10 per cent of the grains on each of four adjacent sieves. When such a sand is heated the finer grains will heat up first and will start to contract while the larger grains are still being heated up and expanding. Thus, there will be both expansion and contraction going on simultaneously and the net expansion may not be severe enough to cause rupture of the mold surface. If you have difficulty in getting a wide spread sand, you can always blend two sands to accomplish the same result.

In addition to using a proper sand, the commonly accepted method of overcoming expansion difficulties is to use various additives in the sand mixture. Sand experts tell us these will decrease the compressive strength at 1000 F and increase the hot deformation of the sand. In everyday language, the sand grains can move slightly during expansion and therefore prevent cracking of the sand surface. Sea coal and pitch have long been used for this purpose.

When heated by molten metal, sea coal passes through a plastic state which allows some movement of the sand grains as they expand. Pitch, under the same conditions, becomes liquid. However, both materials are limited in their ability to increase hot deformation because they go into a solid state before the sand grains may be finished expanding. For this reason, neither sea coal nor pitch are wholly satisfactory.

The best additives known today are cereals, wood flour, and other cellulose materials. They increase hot deformation and reduce hot strength; because of this there is a tendency toward cuts and washes at high temperature. Therefore, do not use more additives than required to do the job. Start with about ½ per cent and increase slowly.

Wood flour and cellulose materials have another advantage. As they burn in a mold they produce a reducing atmosphere which is beneficial in obtaining good finish.

Condition of Pattern Equipment

The casting surface can be no better than the finish on the pattern. If your wood patterns have not seen a paint brush since the day they were made, do not expect the best casting finish. And if your molders think that the proper way to vent a mold is to jab

the pattern full of pock marks, then you are a long way from success. As to venting, the modern thought is that a proper sand with its properties under good control does not require venting with a wire.

The author is not in favor of slicking or trowelling a rough mold. The author believes in the best of pattern equipment and we make every effort to keep it in good condition by repairing and painting at frequent intervals. The casting surface will usually show marks where a finishing tool has been used so if a pattern will not cope or draw perfectly we take it out of production and get it fixed. We do not tolerate inferior equipment because we believe the castings cost more in the long run.

The author realizes that jobbing foundries receive some poor patterns and he suggests that you try to sell your customers on the fact that better patterns will produce better castings.

Clean Sand Required

You will not get the best finish with dirty sand. The author is referring to hard clay balls, scrap metal, core butts, wood, orange peels, cigarette butts, apple cores, etc. Those of you who have mechanized systems probably have clean sand, provided your magnetic pulley is functioning properly and the sand screen is not worn full of large holes. Maintain this equipment.

Those foundries not having mechanized equipment will have to rely on the riddle and there is nothing wrong with that. Make sure that the sand next to the pattern has been put through a riddle, not too coarse, but do not permit your molders to throw the riddlings back into the heap sand or dump them into the drag to help fill it up. That will never keep the sand clean.

Flowability of Molding Sand

Good finish demands a flowable sand. Flowability has been defined as that quality of a sand which enables it to flow or pack around a pattern with the least amount of voids. Every void in the mold surface will produce a corresponding projection on the casting surface, so if you want good finish you must absolutely govern this factor.

Extra tucking, peening and ramming will do a lot to correct poor flowability but this costs extra money and you still have to depend on the human element. Fine sands usually have better flowability than coarse sands. Sands on the dry side will have better flow, than those tempered too heavily. Too much clay and too much cereal will make a sand sticky and stiff and gummy and even though you give it extra peening and ramming, chances are that you will still have a lot of voids. Synthetic sands using southern bentonite will flow better than those made with western bentonite. Wood flour will improve flowability.

Moisture in Molding Sand Mixture

One cubic inch of water will not temper a great deal of sand. As a rough guess that much sand may or may not cover a bench pattern in a 12-in. square flask. Suppose we pour that mold and all the water turns to steam. Do you know that the 1 cu in. of water will produce about 1 cu ft of steam? It expands about 1700 times. It must escape through the mold or else come out the sprue, and it has to move fast. Outside of the fact that this is a lot of steam to get out of a mold quickly, it is sometimes impossible to do it soon enough for the metal to fill the mold completely.

Molders have some tricks of the trade when their sand is heavy and they do not want to cut in some dry sand. They ram lightly over the pattern, put on whistlers and flowoffs, they leave the mold open, hoping it will dry out. Of all the ways to ruin a good finish this is without doubt one of the best. Steam is one of the best oxidizing atmospheres and with it you can get blisters, blows, misruns, burn-ons, and a few pin holes for good measure.

The best sand molders known to the author were always careful about tempering their sand. They were the men who produced the least number of reject castings.

Casting Cleaning Methods

Casting cleaning methods will definitely affect the surface appearance of castings. Star rattling or rumbling will smooth out a rough cast surface. This method of cleaning has gone out of practice to some extent because it requires more time than blasting. If done to excess, the edges and corners of the castings will be rounded over and there is more chance of casting breakage if the rattler is not loaded to capacity. The resulting casting surface while smooth, has a greasy, dirty feeling and this is not particularly pleasing in appearance. Shot blasting by its peening action will tend to flatten out rough surfaces to some extent. Grit blasting can be harmful to a nice smooth finish if the grit is too large.

For shot cleaning on horizontal rotating equipment we are presently using No. 330 shot. Annealed castings are cleaned with No. 460 grit. In our large rotating blast cleaning unit we use a combination: 25 per cent of No. 660 shot and 75 per cent of G14 grit, both of chilled iron. The grit does a fast cleaning job and the shot tends to smooth the surface by peening.

We still use the pickling process for certain thin castings which would be warped out of shape if they were to be blasted.

Sand Preparation and Control

It is impossible to get consistently good casting finish without careful sand preparation and continuous sand control. If good finish is important on your castings, then it is equally important to plan your sand mixtures to accomplish the purpose and then see that the plans are carried out. The best way to do this is to entrust your sand control to an interested and capable man in your organization and then give him time enough to do the job. But his efforts will not be effective unless he gets full cooperation of management and foundry supervision all the way down the line.

If you really want better finish then you must have faith that you can do it. Do not be discouraged by a few blows, blisters or scabs when you try a

finer sand. You will soon discover the proper molding procedures and operating limits of the sand; and then the secret is to stay within those limits. This calls for sand control and consistent molding operations. For many castings we are using a sand that is probably a little on the fine side; but we get away with it successfully by everlastingly watching the sand keeping it under strict control. Our molding operations are under control too; you cannot ram hard one day and soft the next and expect the same results.

Most large and medium size foundries, and many of the small ones, are equipped to do sand testing, but the foundryman pays little attention to what goes on in the sand laboratory or the results of the daily routine sand tests. The young man from the laboratory comes into the foundry at certain times of the day and picks up the sand samples, maybe in nice moisture-tight containers and takes them to the laboratory, in a special cart or hand truck. There he runs the tests and records the results on nice record cards: Moisture, Permeability, Green Strength, Deformation, and Dry Strength. Then what happens? I will tell you, confidentially. He neatly files them away. Yes sir, he files them away. Why? Well, the superintendent is too busy to look at them. He has more important things to do. The foremen are too busy getting out production, watching quality, and a hundred other things, including being a practicing psychologist to his men; and perhaps the supervisors do not understand what the tests mean anyway.

A friend of mine, a sand expert, was called into a large well-known foundry on some sand problem. He talked with top management about their trouble and when he asked about various sand properties which were tested for each day, the man pleaded ignorance and passed the expert along to his assistant. This fellow did not know either so they went to see the foreman of the section where they were having their trouble. He did not know a thing about the sand except that he was expected to produce molds with it. He could not tell the moisture, the permeability, the strength or anything else. But, he knew that the laboratory technician had the figures. Yes, the figures were there, neatly filed away.

You have a sand laboratory that cost you perhaps $5,000 to set up, and you have a salaried man in charge. What return are you getting from it? Practically none. You do not use information that is available to you each day.

You wait until you are in real trouble, and then you call in the sand man and cast the blame on him for getting you into difficulty. Let us be honest; most of the trouble is of our own doing, or not doing. If we take the advice of our sand man, we buy a sand of the proper fineness and distribution for our work. Everything goes just fine for a while and you do not bother to look at the sand tests any more.

But it is necessary to check the sand tests and keep checking them every day. Watch for any slight variation in any of the properties. If you find any, try to discover the reason before you make changes in anything. Stay on friendly terms with the sand technician. This does not imply that the sand technician should have complete authority over the whole sand situation but delegate some responsibility and let him answer for troubles due to sand.

Sand control requires cooperation and team work between the sand technician and foundry molding supervision.

What can be done to get that teamwork? One way that will work is for the sand laboratory to furnish the foremen each day with a record of the tests of their sand. At the top of each day's card list the arbitrary standards for the various properties, such as: Moisture, Permeability, Green Compressive Strength, Dry Compressive Strength, Total Combustible Material, Grain Fineness Number. Teach the foreman to note any deviation between the actual test results and the standards and to try to discover the reason for such deviation. Daily changes in sand properties are usually small and occur gradually but if unnoticed or uncorrected for several weeks may be the cause of major difficulties.

In order that foremen and supervisors may be able to recognize sand changes and what they mean, it will be advisable to conduct a short educational program headed by the sand technician. This need not be intensive because you do not expect the foremen to have to make their own tests. At each meeting, take up one sand property, tell what it is, show the actual test and correlate it with molding procedure.

Producing consistently good finish is no accident—it requires steady everyday control of sand. You may not have to do it today but the trend toward better finish is apparent and you might as well plan on it in the near future.

DISCUSSION

Chairman: R. H. JACOBY, St. Louis Coke & Foundry Co., St. Louis, Mo.

Co-Chairman: E. C. ZIRZOW, Werner G. Smith, Inc., Cleveland.
Secretary: J. B. STAZINSKI, Foundry Div., General Electric Co., Everett, Mass.

F. S. BREWSTER[1] *(Written Discussion):* The author is to be commended for a very practical article on sand. He combined close observation with a thorough knowledge of much that has been written on sand.

He quite properly appeals to management to use the control test figures from the laboratory! If sand is adequately tested and the results buried, is this not the fault of management?

The picture of the sand technician presented here is all wrong. If he resides in an ivory tower only coming down to the foundry for the necessary samples, he is not much good. Neither can he function properly if he has only responsibility and no authority. How can we expect him to answer for sand troubles over which he has no control? I am sure the author has a much better organized sand control program than could be assumed from this one statement.

A good sand "man" lives in the foundry. He knows about how the castings are coming almost as soon as the inspector. As the author indicates, he has to work close to the rest of the supervision, learning from them while he teaches. Obviously he cannot be completely independent in his decisions but must come under control of top supervision the same as everyone else.

1. Director of Research, Brumley-Donaldson Co., Los Angeles, Calif.

ANNEALING ON FIVE-DAY WEEK BASIS

By

R. V. Righter*

Use of bismuth-boron inoculants and high-nitrogen atmosphere in the annealing of malleable iron make possible tremendous advantages in plant operation and efficiency. This paper discusses that subject from a management rather than a technical view point.

The malleable operation at the author's plant consists of 234,000 square feet of floor space, with four 112-in. cupolas, two 15-ton electric furnaces, four pouring conveyors, ten atmosphere-controlled annealing kilns and other necessary foundry equipment. Average hourly melt is approximately 65 tons. Prior to the use of bismuth-boron and high-nitrogen atmosphere it was necessary that the annealing operation be continuous, using ten kilns around the clock, seven days a week, to keep pace with a 5-day foundry operation. Today eight kilns have the capacity to anneal a like amount of iron in a continuous operation of only five days a week.

While the results of bismuth-boron and high-nitrogen atmosphere complement one another, their uses were developed separately for different reasons. As early as 1945, experiments were made using bismuth as a mottle suppressive agent. These were the forerunners of the development of a bismuth-boron combination which proved satisfactory for this purpose under production conditions. In 1953 additional boron was used to counteract the detrimental effects of certain types of pig iron on annealability. Results of this instigated an investigation into the use of boron as a means of reducing annealing time. This investigation, together with a desire to eliminate the use of toxic and explosive high-CO gas used in the kilns, led to the installation of high-nitrogen atmosphere units to reduce the decarburization of casting skins. These latter experiments and with an increase in kiln burner size and number which increased heating capacity by 1,650,000 Btu's or about 24 percent, resulted in over a 30 percent reduction in annealing time. This also made it possible to leave iron in the kilns for prolonged periods of time without decarburization.

The resulting advantages of the use of these processes are an improved and more flexible product, a more efficient operation, and a substantial improvement in plant working conditions. Prior to the use of bismuth-boron the only method of suppressing mottle when carbon and silicon contents approached the high point on the iron chemistry range was a large addition of steel bundles to the electric furnaces. This was a difficult operation with definite limitations. However, bismuth-boron additions were made to selective castings of heavy section thickness experimentally and were found to be very effective in suppressing that mottle which was normally present. The first of these castings were automotive hubs havings section thickness ranging from ¾ in. to 1½ in. and weighing approximately 60 pounds.

On the basis of this success a 100 percent inoculation program was started and at the present time sections of 3 in. are being cast without mottle difficulties. Customer complaints about mottle no longer exist. This ability to control mottling tendencies in sections much thicker than had before been possible, had opened an entirely new field in the type of casting that it is possible for us to make.

Bismuth-Boron Controls Mottling

Before the use of bismuth-boron, casting weight and section thickness were determining factors in the type of malleable iron casting that we were able to produce. With this restriction removed, we have been able to increase the number of castings that it is possible for us to make successfully. Probably the best example of this new flexibility is the automobile crankshaft. In the past year we have been able to place in production the crankshaft for a V-8 automobile engine. This casting has sections as thick as 3 in. and the mottle has been eliminated.

Aside from these direct improvements in our product there has also been a great change in some of our processes. Before, we were forced to work within a very narrow chemistry range of carbon and silicon. We now operate in a much broader range with confidence that as we approach the high side of the iron chemistry range, bismuth will protect us from normal mottle and on the low side of the range, boron will prevent off analysis iron that must be specially annealed. This has greatly simplified our chemistry control and allowed us to use many less expensive melting materials.

*Plant Manager, Central Foundry Division, General Motors Corp., Danville Plant, Danville, Ill.

With the rising prices in such materials as pig iron, steel bundles, and steel scrap, we have been able to go to an "all steel" charge and eliminate the use of pig iron. Ferrosilicon has been substituted to provide the necessary silicon. Our present charge consists of approximately 55 percent remelt, 25 percent steel bundles and 20 percent steel scrap with some addition of ferrosilicon and ferromanganese. This is a substantial increase in steel scrap which is much less expensive than our previously used pig iron. The boron has been particularly successful in suppressing the effects on our annealing cycle of the tramp elements brought in with this scrap.

As mentioned above we have been able to realize the large increase in annealing capacity by shortening our annealing cycle. Before the use of boron we operated on a 32-hour cycle. We are presently on a 22.3-hour cycle and do not feel that we have as yet reached the minimum annealing time. The outstanding benefit we have derived from this increased capacity is being able to start production on the automotive crankshaft without increasing our annealing equipment. We have been able to convert two of our kilns for this production, which is pearlitic malleable iron, and could anneal the same amount of malleable iron adequately in the eight remaining kilns, on a much shorter cycle.

Breakdowns Now Are Not As Serious

While breakdowns are always expensive we feel that the use of bismuth-boron and high-nitrogen atmosphere has substantially reduced such losses. The largest benefit is, of course, in the event of a breakdown of one of our annealing kilns. Prior to the use of high-nitrogen atmosphere it was impossible for the iron to remain stationary in a kiln for more than 12 hours without resulting in almost total loss. With our high-nitrogen atmosphere we have experienced a 72-hour interruption in kiln production with no detrimental decarburization. We have reason to believe that we would be safe in going even beyond a period of 72 hours. Also we are experiencing a longer alloy life in our kilns because of the use of high-nitrogen atmosphere.

Naturally, these benefits have not been brought about without some additional expense. However, this expense has been limited to the actual cost of the bismuth boron inoculant, which is approximately 35 cents per ton of iron inoculated, and to the actual capital investment required for the installation of the high-nitrogen atmosphere units. While the preparation of the inoculant is a very exacting process, it is so insignificant that we have been able to incorporate its preparation into the work patterns of our regular laboratory personnel and are accomplishing it without additional manpower. The inoculation itself is made by the iron pourer at the time his ladle is filled and was easily incorporated in his regular duties. It has been our experience that the cost per cubic foot of high-nitrogen gas is almost identical with the cost of our previously used high-CO system. The cost to increase the heating capacity of the kilns amounted to about $4500.00 per kiln.

In addition to these improvements in the product and in the processes of making malleable iron, the use of bismuth-boron and high-nitrogen atmosphere have brought about changes in plant working conditions that are as valuable as the direct economic advantages. High-nitrogen atmosphere is non-toxic and non-explosive. This has eliminated a hazard existing with the use of high-CO gas.

Annealing Schedules Improved

However, the greatest improvement in this field has been the advantage of being able to operate the annealing department on a schedule similar to the other production departments. Elimination of a seven-day operation which necessitated Saturday, Sunday, and holiday work has resulted in a reduction in necessary labor and has made it possible for the people in that department to work the normal hours of other employes. They are able to participate in employe activities, transportation pools and a number of things that were impossible before. As a morale factor, this has brought about a noticeable improvement in the department efficiency, substantially fewer grievances and an overall improvement in the quality of their work.

While these improvements are difficult to measure, the five-day week is responsible for one direct advantage, the elimination of material handling, that is readily measurable. Previously, it was necessary that the excess iron from the foundry, over and above annealing capacity, be stored for Saturday and Sunday use. With the decrease in the annealing time schedule, only that iron necessary to production on the third shift need be taken from the normal flow. This has reduced the handling of hard iron nearly 30 percent, with a resultant saving in labor and reduction in scrap.

Aside from the cost, bismuth-boron has two disadvantages which should be mentioned. The addition of another material increases the possibility of throwing the chemistry of the iron out of range. To keep the bismuth-boron quantity consistent, the laboratory has packaged this material in small envelopes so that the iron pourer can drop one as the ladle is being filled with iron.

There is also the problem of excessive smoke and fumes when the bismuth-boron is added. Increased exhausting facilities in the iron pouring area where the inoculant is added has helped remedy this.

While through the use of bismuth-boron and high-nitrogen atmosphere we feel that we have made amazing strides in the improvement of the foundry industry, it is obvious that we are merely beginning to recognize the advantages that are possible through their use and we expect to make many more improvements in the future.

DISCUSSION

Chairman: W. D. McMILLAN, International Harvester Co., Chicago.

Co-Chairman: MILTON TILLEY, National Malleable and Steel Castings Co., Cleveland.

E. H. BAKER[1]: Do you need oven operators over the week-end?

MR. RIGHTER: Yes, however we use only two instead of ten.

MEMBER: What percent of bismuth and boron do you use?

CARL JOSEPH[2]: We use between 0.005 to 0.02 per cent bismuth and between 0.0025 to 0.003 per cent boron.

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1. Dayton Malleable Iron Co., Ironton, Ohio.
2. Central Foundries Division of G.M.C., Saginaw, Mich.

H. LEICKLY[3]: Do you leave work in the ovens over the weekends?

MR. RIGHTER: Yes we do. We have found no additional maintenance problems as a result of doing so.

MR. LEICKLY: What do you mean by the higher approach to the chemistry range?

MR. JOSEPH: I mean a higher total of carbon and silicon percentages.

R. J. STICKLE[4]: Does boron allow you to carry more chromium in your iron?

MR. JOSEPH: Boron aids annealing when chromium is high.

MEMBER: On your older kilns, did you reduce your cycle by using boron?

MR. JOSEPH: Yes.

MEMBER: Did you increase the amount of heat going into your ovens when you started to use boron?

MR. RIGHTER: Yes we did.

MR. LEICKLY: Did you use high nitrogen atmosphere in the vestibule only?

MR. RIGHTER: No throughout the ovens.

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3. Fanner Manufacturing Co., Cleveland, Ohio.
4. Meadville Malleable Iron Co., Meadville, Pa.

HOT TEARING CHARACTERISTICS OF ACID AND BASIC STEEL CASTINGS DETERMINED BY HIGH TEMPERATURE TESTING

By

C. F. Christopher*

ABSTRACT

Presented here is a study of the high temperature dynamic properties of a series of steels varying in chemical composition and methods of manufacture.

Results show that all steels exhibit a limited brittle zone under the solidus which is related to the element carbon, and that the temperature limits of this zone are proportional to the per cent of carbon.

The investigator concludes that the hot tearing of castings can occur only during the period when liquid is present at the grain boundaries.

Introduction

It has been estimated that cleaning room costs constitute as high as 30 per cent of the total cost of production of steel castings. The amount of hot tearing repair, which is responsible for part of these costs, varies to some extent upon the type product being produced. Much of this repair could be reduced if the fundamentals of hot rupture were more fully understood.

Before attacking this problem one must be aware of the fact that hot tearing is only one form of failure which originates from the same cause. Hot tearing is a strain rupture occurring during the cooling cycle from liquid to room temperature. Piping, shrinkage, hot tearing, cold cracks, blow-ups, etc., are all ruptures during this cycle and are all identically caused. The only difference between them is the temperature level at which they occur.

The temperature of the metal, during the cooling cycle, determines the physical characteristics of the metal at any particular time. If strained, every defect down the line will look different even though they are caused by the same thing.

If, during the cooling cycle, certain ranges of temperature produce different types of strain, it is important that we know the width of each temperature range and the type of defect in each.

Since this investigation concerns hot tearing, we are interested in knowing how far under the solidus this defect can occur. The location and width of this range has much to do with effective methods of eliminating it.

The ingot industry has made great progress in eliminating hot tearing in ingots. This has been done chiefly through their knowledge of pouring temperature and design. Their problem, however, cannot be compared to that of a foundry which must be able to produce sound castings free from hot tearing without much help from design.

In order to eliminate or minimize hot tearing, we must first know what a hot tear is and all the conditions surrounding its formation. Only after this is clarified can we develop intelligent means of eliminating it.

This paper is a report on an extensive investigation dealing primarily on the definition of a hot tear and its characteristic nature in steels of different compositions made under different processes.

If hot tearing is to be intelligently controlled or eliminated it must be considered as originating from some fundamental cause. Steel quality, sand collapsibility, design, etc., are naturally contributory to hot rupture, but the mechanism of failure, temperature ranges, and metallurgical factors are far more important in concentrating everyone's efforts toward a common end.

In this investigation we selected 23 heats of steel varying widely in chemical composition. Some of the heats were made in the acid open hearth, some in the acid electric, and some by the basic electric process. The acid steels were deoxidized with silicon only while the basic were deoxidized with either straight aluminum plus silicon or aluminum plus silicon plus rare earths.

The manufacture of good steel castings or other products is basically divided into two distinctively separate parts. One is the actual manufacture of the steel, and the other is the ability to cool that steel to room temperature in a manner which will keep it sound. The characteristic conditions existing in steel at various temperatures during this cooling cycle, may or may not have anything to do with the actual steel making.

Steel Making Practice

This investigation was not meant to be one which would prove or disprove the effect of steel quality on hot tearing. However, we have certain steelmaking

*Director of Research, Continental Foundry & Machine Division of Blaw-Knox Co., Chicago Works, E. Chicago, Ind.

practices which we developed as standard for quality. We have learned through the years, however, that inherent steel quality has nothing to do with hot tearing as long as the definition does not include temperature.

It is fortunate that our beliefs in steelmaking principles regarding temperature pouring levels, coincide with our principles of hot tearing elimination. We believe that good steelmaking practice requires temperature control and maximum temperature levels. This is also where we believe hot tearing control begins.

In previous investigations by the author, conclusions are that temperature plays an important part in the elimination of porosity and the formation of proper inclusion types in all steelmaking processes.

All deoxidizers or elements which remove oxygen, do so under the influence of temperature, with the exception of carbon. This relationship of temperature to oxygen removal by elements has been carefully investigated and published. A practical summary of the basic principles of oxygen removal and its effect is shown in Fig. 1.

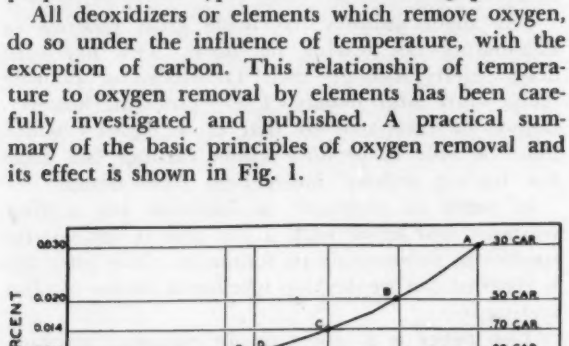


Figure 1

This figure shows the relationship of carbon and silicon to temperature and the pouring temperature limitations of certain carbon steels which produce proper inclusion types and insure against porosity. The figure refers to steels containing 0.40 per cent silicon.

In the manufacture of relatively large steel castings and plain designs it is possible to produce steel free from porosity and with proper inclusion type with straight silicon deoxidation. This can be done if the composition is under 0.50 per cent carbon and with certain pouring temperature limitations. For instance, the maximum pouring temperature of a 0.50 per cent carbon, 0.40 per cent silicon is 2755 F. For a 0.30 per cent carbon steel the maximum pouring temperature is 2855 F.

Many plants make higher carbon steels than 0.50 per cent, and due to small size castings, may be forced to pour at higher temperatures even on lower carbon steels. These conditions require that the steel be deoxidized with some deoxidizer stronger than silicon, the most common being aluminum. Figure 2 shows the conditions prevailing when aluminum is added to various carbon steels which already contain 0.40 per cent silicon. This figure shows the inclusion type present with various aluminum additions and the proper amount necessary to produce the best inclusion type, that is Type 3.

It can be seen that aluminum or other strong deoxidizers permits higher temperature operation to those companies which require this type of operation.

Naturally there are objections to high temperature practices due to danger of burnt-in sand and porosity, not to mention the effect of high temperature on hot tearing.

Since the deoxidizing power of all elements used for the purpose are related to temperature, weakening as the temperature rises, it is important to operate at a temperature no higher than necessary to avoid porosity and the use of excess deoxidizers.

It can be seen that temperature plays a vital part in the chemistry of steelmaking, always favoring the lowest temperature of operation for any given practice. Temperature also plays an important part in hot tearing of steel castings, since it is the contraction of steel from various temperatures which is the source of stress. Further discussion on the effect of temperature levels will be shown in the report which follows.

Discussion of Testing Procedure

Before the problems of hot tearing and cracking can be attacked, a study must be made of the characteristics of steel from its freezing point down through all temperatures.

It is known that all steels will tear as long as any liquid metal exists during freezing, if stressed by adjacent metal contraction, stress concentration, or any other mechanical reason. The important problem is to determine, under the most severe stress concentration conditions, the strengths and weaknesses of steel at various lower temperature ranges, and to determine the limiting ranges of brittleness.

Hot tearing is a natural phenomenon occurring when stress is applied to a steel which is liquid around its austenitic grain boundaries and ceases when the strength of the grain boundaries resists separation sufficiently to cause the grains themselves to strain.

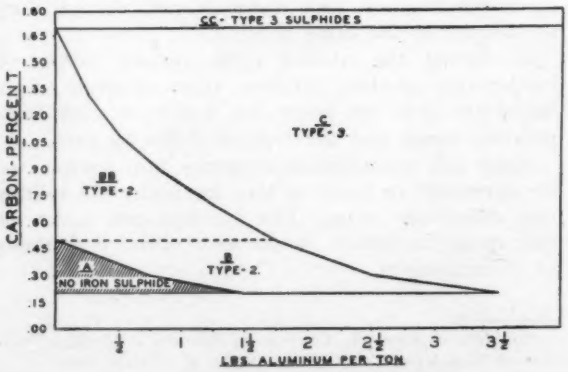


Figure 2

When the grain boundaries are weaker than the grains, it is safe to assume that the speed at which the stress is applied would be immaterial. A casting would tear around the grain boundaries if the metal was in this state, regardless of whether the stress was applied slowly or rapidly. Therefore, any mechanical test which could be devised for use at the temperatures in question should bring out the weaknesses existing between grain and grain boundaries.

Furthermore, if stresses, in any concentrated form, magnify these weaknesses around grain boundaries, a severe stress concentration test would be advisable in determining these weaknesses to bring them out to their fullest extent. Any severe dynamic test, with severe stress concentration, would determine complete embrittlement. At any rate, steels which would show no grain boundary weakness with this test would certainly not hot tear under lesser or under shop conditions.

A method of using the Charpy impact test at all temperatures up to the freezing point of the steel has been devised. The test specimens are completely machined and carefully notched. They are heated under accurate thermocouple control, either in salt baths or in argon atmospheres, depending upon the temperature range. All tests, regardless of temperature ranges, are fractured within two to three seconds after they are removed from the salt or argon atmosphere.

Not only are the foot-pound strengths recorded, but more important, the nature of strain resulting.

Most important of all, these tests will determine if there is a temperature where steels of various compositions cease to hot tear during their solidification cycle.

Furthermore, this test gives us the opportunity to definitely compare one steel composition with another regarding hot tearing susceptibility.

Any type of open hearth practice, various deoxidizers, additions, etc., can be definitely evaluated regarding hot tearing susceptibility. This test makes it possible to classify steel compositions regarding hot tearing and to help determine the source, whether it be due to composition, methods of heading and gating, or temperature.

Chemical Compositions

Table 1 shows the chemical compositions of the acid steels selected for hot tear investigation. These steels represent typical compositions made in the author's plant, including rolls, wear resistant steels, armor plate, and casting steels for various uses. All these heats are deoxidized with straight silicon.

TABLE 1 — ANALYSIS OF TEST HEATS

Heat	C	Mn	Si	S	P	Cr	Ni	Mo	Cu	B
A	1.83	.96	.37	.043	.080	1.00	—	.34	—	—
B	1.30	.82	.39	.046	.053	.87	.61	.35	—	—
C	.90	.83	.43	.053	.052	.93	—	.42	—	—
D	.29	.77	.47	.045	.041	2.86	—	.45	—	—
E	.47	.78	.41	.045	.040	—	—	—	—	—
F	.30	.82	.42	.038	.044	—	—	—	—	—
G	.28	.86	.38	.044	.040	.97	.90	.52	—	—
H	.26	.80	.47	.048	.049	.29	.61	.39	—	—
I	.28	.85	.47	.035	.022	.70	.32	.48	.80	—
J	.30	1.64	.19	.013	.016	—	—	.49	—	.0008
K	.27	2.20	.46	.034	.020	—	.28	.49	—	—

Heats A, B, C, D, and E were commercial heats made in a 65-ton acid open hearth.

Heats F, G, H, I, and K were acid electric arc heats experimentally made in a 5,500-lb furnace.

Heat J is from purchased rolled plate of which we have no records.

It is obvious that if alloy composition was a big factor in hot tearing, it would be difficult to arrive at any conclusions from this list of heats, due to the various combinations of alloys. However, they do cover a wide range of carbon contents, (carbon greatly affects the melting point and the amount of selective freezing), and represent steels at various locations on the iron-carbon constitutional diagram.

TABLE 2 — CHEMICAL COMPOSITIONS

		Basic + 2½ lb Al per Ton								
		C	Mn	Si	S	P	Cr	Ni	Mo	Cu
L	—	.90	.85	.43	.020	.030	1.59	—	.36	—
M	—	.50	.82	.43	.023	.023	—	—	—	—
N	—	.34	.72	.38	.022	.012	2.98	—	.50	—
O	—	.29	.87	.44	.024	.018	—	—	—	—
P	—	.28	.81	.45	.028	.032	.83	.80	.49	—
Q	—	.30	.91	.36	.026	.019	.41	.66	.37	—
R	—	.30	.53	.15	.030	.018	.53	.43	.53	.82

Table 2 shows the chemical compositions of the seven basic electric steels used in this hot tear investigation. All of these heats were made with the idea of comparing their characteristics with those of the acid steels. Therefore, it can be seen that the compositions compare approximately with the corresponding grades made in the acid group.

These steels all have a silicon specification, but in addition are deoxidized with 2½ lb of shot aluminum per ton in the ladle.

TABLE 3 — CHEMICAL COMPOSITIONS

		Basic + 2½ lb Al and 2 lb Rare Earth/Ton								
		C	Mn	Si	S	P	Cr	Ni	Mo	Cu
S	—	.28	.78	.43	.020	.032	.59	.78	.48	—
T	—	.30	1.17	.45	.018	.014	2.80	—	.45	—
U	—	.30	.72	.42	.018	.032	1.00	1.10	.46	—
V	—	.30	.90	.47	.026	.023	—	—	—	—
		2 lb Ca-Si-Al per ton (acid)								
W	—	.46	.79	.54	.034	.027	—	—	—	—

Table 3 lists four heats of basic electric steel deoxidized with 2½ lb aluminum plus 2 lb of rare earth per ton. These steels also require a silicon specification.

Included also is an acid steel deoxidized with 2 lb of Ca-Si-Al per ton.

These heats have compositions comparable to the acid steels killed with silicon and the basic with aluminum, although they include only the lower carbon grades.

Experimental Procedure

Since the purpose of this experimentation was to investigate the inherent characteristics of hot tearing, it was felt important that there should be some standardization of the coupons from which the im-

pacts would be taken. By so doing, most of the physical variables which would tend to confuse results would be eliminated.

Uniform-size keel blocks were poured on all heats. These blocks consisted of a head with two 3 in. x 2 in. x 10 in. coupons attached, and fan gated to insure clean tests. All keel blocks were sand cooled, sawed off, and annealed at 1750 F.

The impacts were all cut from these coupons, ground to exact specifications, and notched. All notches were inspected at 25 magnification to meet government specifications.

Table 4 contains the test results for all acid steels from 2000 F up to the temperature where the steels became brittle. Each figure shown is the average result of several tests on each steel at the specific temperatures.

TABLE 4 — IMPACT PROPERTIES, FT-LB, OF TEST ACID HEATS AT ELEVATED TEMPERATURES

	Temperature, F						
Heat	2000	2100	2200	2300	2400	2500	2600
A —	20	15	—	—	—	—	—
B —	76	19	—	—	—	—	—
C —	80	70	59	47	26	—	—
D —	83	74	58	53	17	—	—
E —	84	74	59	56	50	38	4
F —	82	71	54	49	44	39	4
G —	74	68	59	47	46	42	—
H —	83	68	61	51	44	39	6
I —	79	72	63	56	48	42	6
J —	82	77	61	55	48	40	20
K —	84	70	61	52	45	40	20
Aver.	80.7	71.6	59.5	51.6	46.4	40.0	—

The double solid line shown is the separation line between ductility and brittleness for all the acid steels. Below and to the left of this solid line there was no indication whatsoever of brittleness. It can be seen that for any given temperature the impact results are remarkably close, regardless of composition, as long as the steels are in the ductile zone.

Above the solid line, all tests show a brittle, granular fracture, and vary in strength from zero to a few foot-pounds.

Referring to the figure, composition A, the 1.83 per cent carbon steel, is brittle and crystalline at all temperatures tested. From 2000 F up to the temperature of complete melting the fractures on this steel indicate that failure is occurring around the austenitic grain boundaries. All grains remain intact with evidence of grain boundary melting.

Decreasing the carbon to 1.30 per cent, composition B, shows that the grain boundary melting now occurs between 2000 and 2100 F. The temperature at which brittleness appears has increased due to the decrease in carbon. At 2000 F the 1.30 per cent carbon steel is very ductile, whereas it is extremely brittle at 2100 F. In this 100 degrees the fracture has suddenly changed from a ductile type to a brittle, grain boundary type.

A further lowering of carbon to 0.90 per cent, composition C, further increases the ductile zone to between 2300 F and 2400 F. The brittle fracture at 2400 F shows the same evidence of grain boundary weakness, originating from apparent melting around these boundaries, as compositions A and B.

When the carbon is lowered to the 0.30 to 0.50 per cent range, compositions E to K, all steels become brittle at 2600 F, with extreme ductility at 2500 F.

It is noteworthy to observe that the brittleness is a carbon phenomenon. In spite of the wide variation in chemical composition, this physical property change, occurring at specific temperatures, follows only the per cent carbon. Furthermore, the fact that the fracture in the brittle zone occurs around the grain boundaries indicates that there is a heterogeneity of carbon from the center to the surface of the grain. It is apparent that brittleness occurs the instant the melting point of the grain boundary is reached.

It is obvious that the impact-temperature testing method used in this experiment is simply the determination of grain boundary melting points by measuring the foot-pounds and observing the change of fracture above and below the temperature where the first melting occurs.

Selective freezing during solidification causes this heterogeneity of carbon. On higher carbon steels the carbon around the boundaries of the grains reaches a much higher content than on lower carbons. This accounts for the various temperatures of grain boundary melting, and consequent brittleness, with changes in the carbon content of the steel.

Table 5 shows the test results of heats made in the basic electric furnace. It will be noted that these heats were made to correspond in composition with the acid heats for comparison reasons; however, not all of the acid heats were duplicated in basic composition.

TABLE 5 — ELEVATED TEMPERATURE IMPACT PROPERTIES, FT-LB, OF BASIC HEATS DEOXIDIZED WITH ALUMINUM

	Testing Temperature, F							
	2000	2100	2200	2300	2400	2500	2600	2700
A —	—	—	—	—	—	—	—	—
B —	—	—	—	—	—	—	—	—
L —	90			56	48	2	—	—
M —	76			52	44	38	10	—
N —	86			62	49	44	39	2
O —	76			50	46	40	36	0
P —	81			54	51	43	38	2
Q —	76			52	38	39	37	0
R —	74			48	48	41	38	0
J —								
K —								
Ave.	80			53	46	40	38	—

Basic + Al

This particular group of heats was deoxidized with 2½ lb of aluminum per ton and tested under the same temperature schedule as the acid steels, that is, from 2000 F up to the temperature of brittle rupture.

The double solid line indicates the dividing line between ductility and brittleness and shows the temperature where each steel completely lost ductility. The dotted line indicates where the same steel grades in acid steel became brittle.

The amount of ductility and strength in the basic corresponds very closely to that of the acid steel at any given temperature. However, as can be seen, the basic retains ductility approximately 100 degrees higher.

Table 6 shows the test results on the basic heats deoxidized with aluminum plus rare earths. In this table again the solid line indicates the dividing line between ductility and brittleness while the dotted line indicates where the same grade of acid steels became brittle.

TABLE 6 — ELEVATED TEMPERATURE IMPACT PROPERTIES, FT-LB, OF BASIC HEATS DEOXIDIZED WITH ALUMINUM PLUS RARE EARTHS

	Testing Temperature, F							
	2000	2100	2200	2300	2400	2500	2600	2700
A	—	—	—	—	—	—	—	—
B	—	—	—	—	—	—	—	—
C	—				—	—	—	—
W	— 80	Ca-Si-Al Acid		53	52	42	4	0
T	— 88			57	55	46	40	0
V	— 78			51	48	41	36	0
U	— 74			52	50	44	36	0
S	— 78			53	52	42	41	0
Ave.	80			53	51	43	38	0
	Basic + Al + Rare Earth							

Heat "W" is an acid heat deoxidized with 2 lb of Ca-Mn-Si. It was introduced in this table because of its special deoxidation.

It will be noted that these basic heats also retain their ductility 100 degrees higher than the acid heats, with exception of Heat "M", a 0.50 per cent carbon basic heat killed with aluminum.

In summarizing the test results, all indications lead to the fact that the basic steels have a narrower range of hot tear susceptibility than the acid. At least the basic steel loses its brittle characteristics at a higher temperature than the acid steels.

Whether this condition favors basic casting manufacture over acid depends upon whether or not this favorable characteristic is utilized. It must be remembered that both acid and basic steels have a hot tear range, the only difference being that the acid range is possibly wider than the basic. A bad stress condition at the wrong temperature would tear a casting regardless of the width of the brittle range.

A steel that retains ductility at a higher temperature has, without a doubt, a very favorable characteristic. This means that the surface of the casting is more likely to drop into a ductile state before major stresses appear, that is, for a given pouring temperature. It also means that good heading, gating, and chilling practices would be very effective in eliminating hot tearing.

The narrower range of brittleness in the basic steels may be the result of several things. It is known to lose its temperature faster and consequently freeze faster. Basic steel also contains less phosphorus, which decreases its fluidity and accelerates its setting characteristics.

The average phosphorus and sulphur composition of the heats involved in this investigation are as follows:

Acid		Basic plus Aluminum		Basic plus Al plus Rare Earth	
S, %	P, %	S, %	P, %	S, %	P, %
0.043	0.044	0.025	0.022	0.021	0.025

It is believed by the author that sulphur in the above amounts, with the proper deoxidation practice, has less effect than the phosphorus on the fundamentals of hot tearing. This report strongly indicates that hot tearing, in itself, is a carbon phenomenon, and that the width of the hot tearing range is due to the amount of selective freezing involved. The amount of selective freezing influences the heterogeneity of the grain versus the grain boundary.

Naturally, such items as fluidity, setting characteristics, nucleation of grains during freezing, etc. are selective freezing influences. Phosphorus, aluminum additions, etc. are a big part of this picture, and probably more influential than sulphur.

It is possible that pouring temperature is as influential in controlling cooling rates and selective freezing as is phosphorus. We may be just as sure of controlling hot tearing in acid steel, with its higher phosphorus, by controlled lower pouring temperature, as we are with low phosphorus basic steel with its aluminum nucleation and quicker setting.

To further stress the value of high temperature dynamic testing in clarifying the characteristics of hot tearing, the following curves and micrographic study near the hot tear temperature ranges are of interest.

Figure 3 refers to the high temperature impact values of the acid steel group. The shaded area represents the range of impact values of all the 0.30 to 0.50 per cent carbon steels at individual temperatures from 1100 F to the melting point.

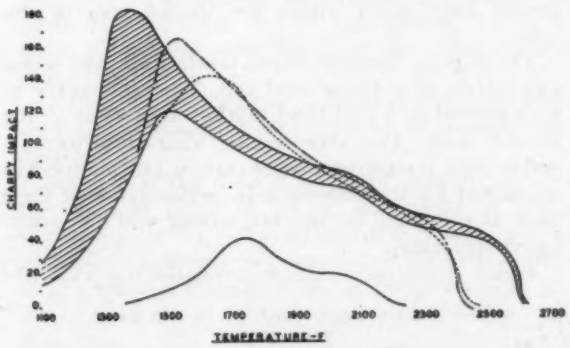


Figure 3

Testing the steels at lower temperatures shows that they all possess extreme brittleness around 1100 F. This brittle range is believed to be caused by carbide precipitation around the grain boundaries. This carbide dissolves as the temperature is raised, as indicated by physical testing at higher temperatures. The impact values reach a peak at the critical transformation temperature. The higher manganese steels give higher values and transform at lower temperatures than lower manganese steels. On the curve, the higher manganese steels lie at the top of the shaded area and

the lower manganese at the bottom. All the other 0.30 to 0.50 per cent carbon steels fall somewhere within the shaded area.

As the steels are further tested at increasingly higher temperatures, all values gradually decrease and approach the same value. In other words, the wide range of impact values at the lower temperatures are affected by composition which affects the transformation temperatures. Above the transformations all the compositions become austenitic and produce equal physical property values at a given testing temperature.

All values gradually decrease as the testing temperature increases, simply because the austenite softens. After the steels transform, the tests indicate extreme ductility with no sign of brittleness until grain boundary melting occurs.

This grain boundary melting occurs suddenly after the temperature passes 2500 F on the acid steels in the 0.30 to 0.50 per cent carbon range.

Referring to the figure, the solid line shown at the bottom is the 1.83 per cent carbon steel. This steel is not ductile due to free cementite and shows brittleness at all testing temperatures.

The dotted line represents the 0.90 per cent carbon steel. It can be seen that this steel has a higher critical temperature as shown by the hump in the curve at about 1600 F. After transformation the values gradually decrease until grain boundary melting occurs. However, this higher carbon steel reaches this grain boundary melting sooner than the lower carbon ranges. This melting condition begins after the temperature passes 2300 F instead of 2500 F as was the case with the lower carbon.

The broken line shown in Figure 3 is of interest. These tests were taken from a high chromium steel with a ladle analysis of 0.30 per cent carbon, and should have fallen within the shaded area of the curve.

These tests, however, were taken from the segregated body of a 40-ton steel roll. The same grade of steel poured in a keel block test fell uniformly in the shaded area. The slow setting characteristics, together with the segregation of carbon and phosphorus accounted for the difference in performance of these tests as compared to quicker setting and less segregated specimens.

Figure 4 represents the accumulated test values of basic heats from below the transformation temperatures to 2000 F. These tests are taken from all the medium-carbon grades deoxidized with aluminum and with aluminum plus rare earths. These values agree very closely with those found in acid steels.

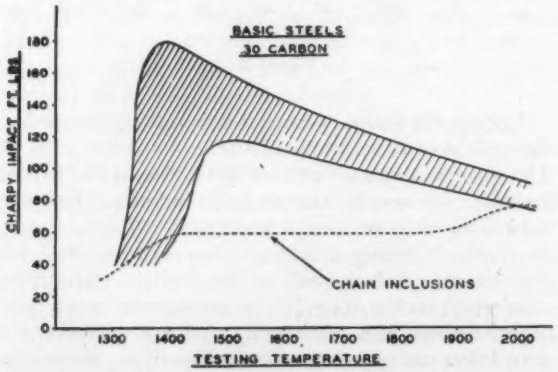


Figure 4

Later in this report it will be seen that the test results above 2000 F are different in basic and acid steels. Reference will be made to the temperatures at which these two types of steel melt around the grain boundaries.

There have been many discussions and differences of opinion regarding the effect of chain inclusions upon hot tearing. Chain inclusions are a menace to steel quality, but are sometimes produced accidentally in practice. This condition seldom occurs in acid steel below 0.35 to 0.40 per cent carbon with a straight silicon deoxidation practice. Porosity can occur if the steel is poured too hot, but it is practically impossible, theoretically, to produce grain boundary sulphides at normal acid steel sulphur contents with the above practice.

On the other hand, other acid practices which use strong deoxidation, as well as killed basic steels, are always in danger of chain inclusions even with lower sulphur. When deoxidizing strongly, there is always danger of losing some of the deoxidizer in the slag, which may throw the metal into Type II inclusions.

It was not the intent in this investigation to study the effect of chain inclusions on hot tearing. This investigation is a study of the hot tearing susceptibility of good metallurgical products.

One of the experimental heats in this investigation, however, lost some of the deoxidizer in the slag, chain inclusions resulting. This was a 0.30 per cent carbon alloy heat made by the basic practice, deoxidized with 2½ lb of aluminum and 2 lb of rare earth per ton. Out of curiosity, this heat was carried along with other basic heats to compare the effect of this grain boundary condition upon hot tearing.

The dotted line in Fig. 4 represents the test properties of this steel at various temperatures as compared with properly deoxidized steels. It can be seen that this grain boundary condition destroys the strength and ductility. The fact that this line remains practically constant from 1400 F to 2000 F and, therefore, approaches the shaded curve as the temperature rises, indicates that chain inclusions lose their effect at temperatures near the hot tear range. At 2000 F, this heat completely loses its identity as far as properties are concerned. The physical characteristics of this heat, above 2000 F, acted exactly as did all the other 0.30 per cent carbon basic heats near the hot tear range.

This heat contained 0.018 per cent sulphur, a percentage which apparently did not affect the hot tearing condition. We are aware, however, that high sulphur as found in screw stock, for instance, does affect hot tearing. As the sulphur increases materially the amount of sulphides increases with improper deoxidation, and apparently the purity is also affected. Pure iron sulphides melt at 2187 F.

The following figures deal with the steels at temperatures near the hot tear range and show how these

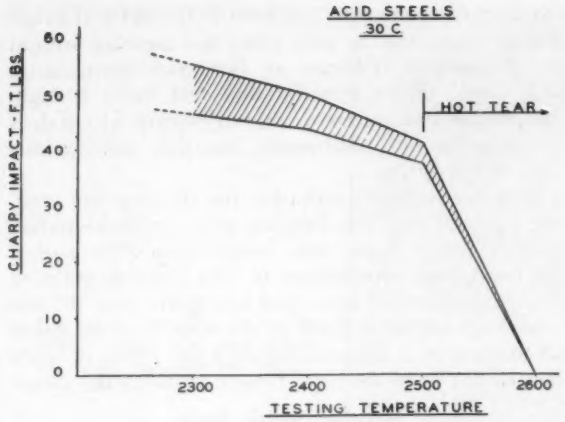


Figure 5

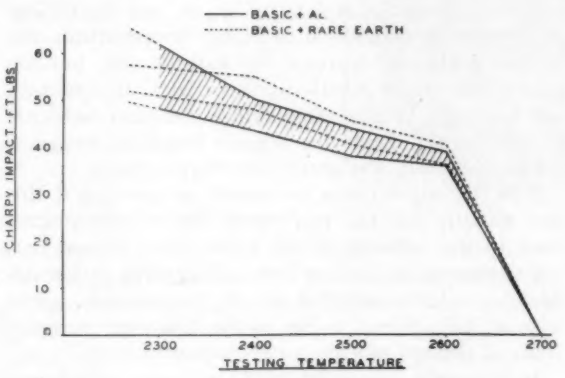


Figure 6

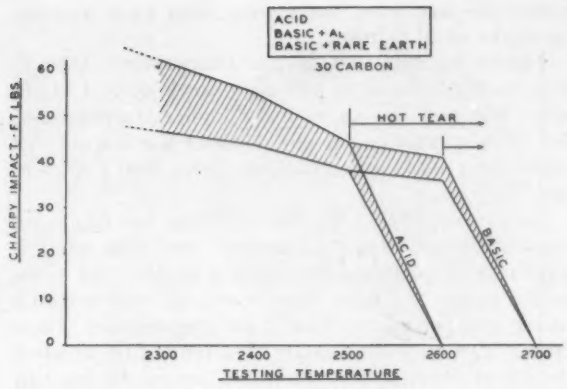


Figure 7

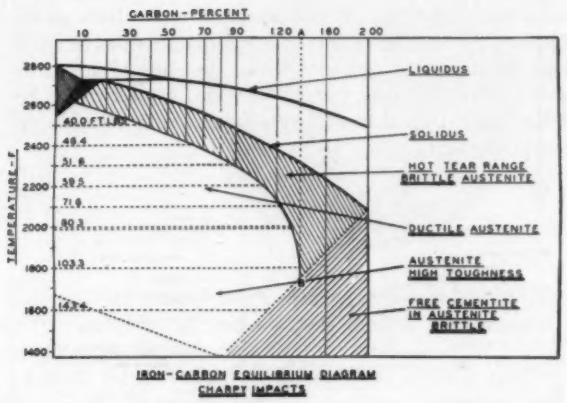


Figure 8

conditions apply to the iron-carbon constitutional diagram.

Figures 5, 6, and 7 are the impact test values of acid and basic steels and their comparisons in the temperature ranges near the hot tear zone. For comparative purposes these curves represent only those steels of the lower carbon range. They include eight acid heats, five basic plus aluminum, and four basic plus aluminum plus rare earth.

All impact values drop sharply when a given temperature is reached. This particular condition develops in acid steel, (See Fig. 5) when the temperature exceeds 2500 F.

On both the basic with aluminum and the basic with aluminum and rare earths, the condition develops when the temperature exceeds 2600 F (See Fig. 6)

The differences in practice are more clearly shown in Fig. 7. This figure shows that there is a consistent difference between acid and basic, a difference that is unaffected by rare earths. Lower phosphorus and sulphur exist in both basic practices and both used aluminum deoxidation. The acid is a coarse grain steel with straight silicon deoxidation and contains approximately double the phosphorus and sulphur present in the basic heats.

Since this entire investigation shows that hot tearing follows carbon and is a selective freezing phenomenon, it is believed that the difference in hot tearing between these two practices is a difference in grain boundary heterogeneity of carbon. The nucleating effect of aluminum and/or the quicker setting of lower phosphorus steel may favor less carbon heterogeneity.

A better picture of the physical property data can be seen by transposing the results to the constitutional diagram.

Figure 8 is the pure iron-carbon diagram showing the hot tear range of all the acid steels. It is not necessarily the correct diagram for each steel investigated, but is shown to demonstrate the relationships of properties to selective freezing, the physical properties in austenite at various temperatures, the hot tear zone, etc.

Various carbon steels up to a given carbon content give similar properties at given temperatures below the brittle zone. Likewise, each carbon content remains ductile as the temperature is raised up to some definite point. This set of conditions, superimposed on the diagram, allows us to predict the properties and hot tear limitations of all steels.

As shown in Fig. 8, for instance, a 0.60 per cent carbon acid steel hot tears above 2400 F. All steels of 0.60 per cent carbon and lower will not tear at or below 2400 F, and will all give about 46.4 ft-lb strength at this temperature.

At 2300 F only those steels 0.90 per cent carbon and

under are free from hot tearing, and have uniform strengths of 51.6 ft-lb.

Figure 8 shows that at low temperature, 1800 F, hot tearing is absent in all carbon steels up to 1.30 per cent. However, at increasingly higher temperatures, less carbon must be present to avoid hot tearing. All carbon steels, as shown, hot tear above 2600 F on acid steels.

As shown in Fig. 8 there is a definite hot tear band apparently following the solidus, and this band is evidently below the solidus, with a tendency to widen as the carbon increases. Apparently, all steel is brittle above 1.30 per cent carbon at all temperatures. These higher carbon steels contain free cementite, which is brittle at lower temperature, and are in the hot tear zone at higher temperatures.

Figure 9 shows the same constitutional diagram with the properties of both acid and basic steels superimposed. It can be seen that the impact values of these two practices are very similar in the ductile austenite area. The hot tear range, however, is narrower for basic steel since this steel retains ductility to a higher temperature.

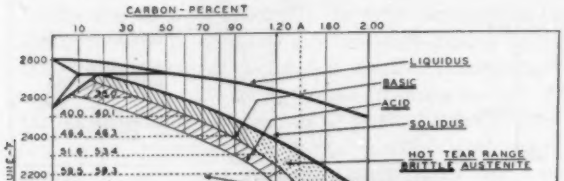


Figure 9

The constitutional diagram, as we know it, shows the temperatures at which various carbon steels begin and complete freezing with the corresponding compositions of liquids and solids during the freezing cycle. The curve is produced under equilibrium conditions during slow cooling. However, it is questionable whether these exact conditions prevail in the cooling of a casting in practice. We believe that the bottom of the hot tear range, as it has been determined in this experiment, is the actual solidus and is below the theoretical solidus.

When we heat impact tests to increasingly higher temperatures and test them at these temperatures, the grain boundaries resist rupture until melting takes place. This melting takes place at the grain boundary first, apparently because of the higher carbon at this point. There are other theories why melting takes place at the boundaries first, but since the temperature at which this melting takes place follows carbon, we believe that it is caused by higher carbon close to the boundary.

It is interesting to note that the temperature at which this melting takes place never changes. We have reheated the same steel specimen to the hot tear range several times, but in each case, this melting around the boundaries re-occurs at the same temperature each time. If the material that first melts is high-carbon material, we are unable to explain why it does not migrate, but consistently identifies itself at the same temperature.

It is interesting to note that the freezing and melting point of this low melting point grain boundary material occurs at the same temperature. The author has conducted experiments on the freezing point of these experimental heats and has found that the hot tear range identifies itself at the same location either when the steel is reheated through this point, or when it is cooled from the liquid down through the range.

Metallographic Study

As explained previously, the hot tear range was determined by heating impact tests to the various temperatures in an atmosphere of argon, and fracturing. As the testing progressed to higher temperatures, the values gradually dropped in foot-pounds, but remained extremely ductile until a critical temperature was reached. At this temperature the steel suddenly became brittle, displaying a grain boundary weakness which obviously was grain boundary melting.

The investigator was interested in knowing if this was actually the hot tear range and if the method used in determining it was valid, since in practice, hot tearing occurs during the cooling cycle. It was important to know whether or not the research agreed with actual practice as far as the lower temperature limit of the hot tearing range was concerned.

It is actually impossible to do a precise job of fracturing impacts on a cooling cycle from the liquid steel temperature. It is difficult to produce tests which are dimensionally perfect, and temperatures which are either accurate, or accurately uniform from surface to center. This can, however, be accurately done on reheating.

The most conclusive method for studying the hot tear range, and the different effect, if any, caused by heating from the solid or cooling from the liquid, is by accurate microscopic studies of the steel under each condition.

In order to study the microstructure of the steel during the heating cycle at various temperatures near the hot tear range, small sections were heated in an argon atmosphere. When the samples reached the desired temperature, they were soaked for 20 minutes at temperature, and then quenched in ice water. These structures were examined microscopically for changes which might have taken place just below, and in the hot tear range. The temperatures selected for study were the same as those used in impact testing. In other words, a microscopic study was made of the steel as it changed from the ductile zone to the brittle zone as determined by the impact test. Studies were made on 0.30, 0.50, and 0.90 per cent carbon steels just under, and within the hot tear range. These microscopic studies were made on both acid and basic steels.

For a study of the cooling cycle, small drillings were taken from corresponding acid and basic steels. These drillings were placed in a refractory boat and heated

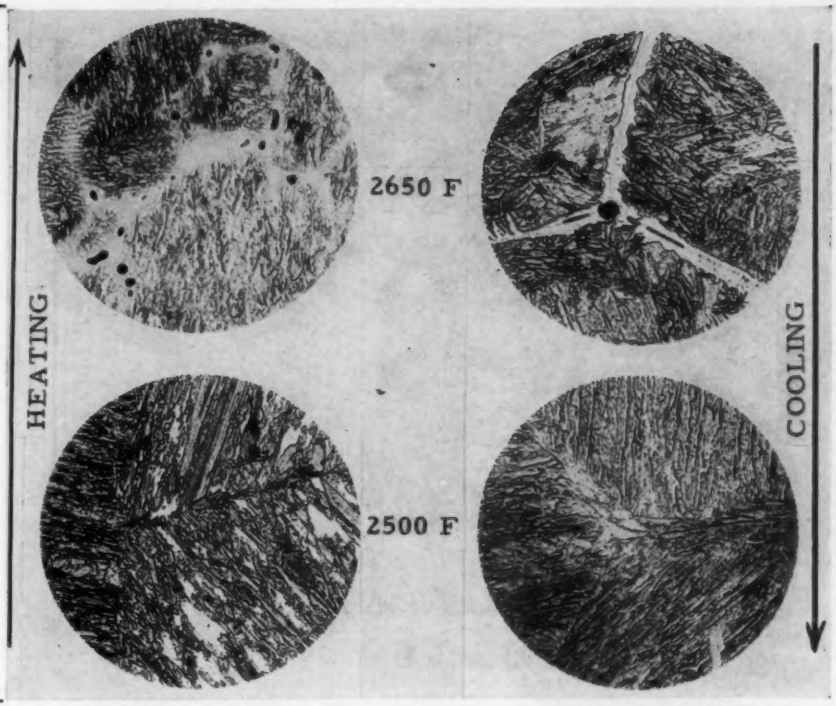


Fig. 10 — Microstructure of 0.30% carbon acid steel in the hot tear temperature range. Etched. Mag. — 1000X.

in an argon atmosphere to the liquid state. They were then cooled to either the hot tear range or just below the hot tear range, and held there for ten minutes prior to quenching in ice brine.

In studies of both the heating and cooling cycles, the same temperatures were used for micrographic study. These temperatures originated from those determined by impact studies.

This work made available for study impacts and heating and cooling micrographs at known temperatures. It was important to know if the micrographs agreed on heating and cooling and if the impacts agreed with the micrographs.

Figure 10 shows the heating and cooling micrographs, within and below the hot tear range. These represent a 0.30 per cent carbon acid steel at 2650 F and 2500 F, at 1000 magnification.

The two photomicrographs on the left represent the heating cycle and show the structures at 2650 F and 2500 F, respectively.

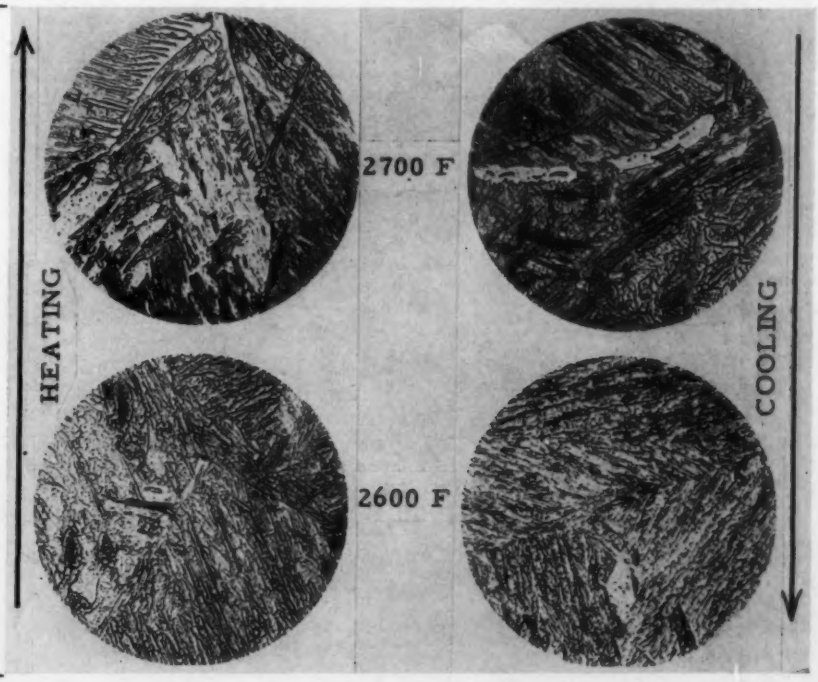


Fig. 11 — Microstructure of 0.30% carbon basic steel in the hot tear temperature range. Etched. Mag. — 1000X.

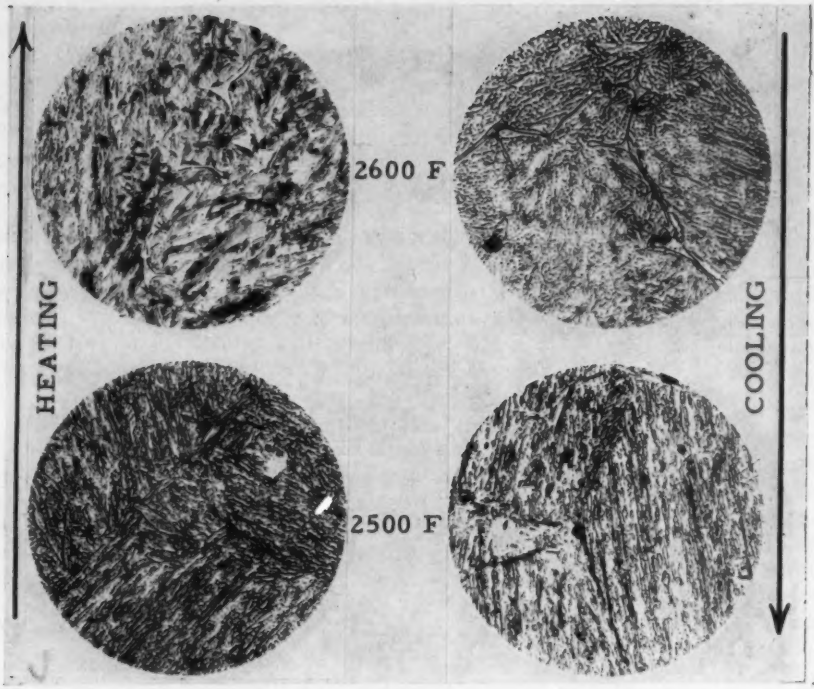


Fig. 12 — Microstructure of 0.50% carbon acid steel in the hot tear temperature range. Etched. Mag. — 1000X.

At 2500 F the impact value was 39.0 ft-lb. It can be seen that the grains are large and that no melting has taken place.

At 2650 F, however, the structure shows that melting has begun to take place, and the impact has dropped to 4.0 ft-lb.

On the cooling cycle from the liquid, the two photomicrographs on the right represent temperatures of 2650 F, above, and 2500 F, below.

It can be seen that the structures again show melting at 2650 F and complete solidity at 2500 F. Even though the melting condition is more pronounced at 2650 F on the cooling, it has completely frozen at 2500 F.

In general, the photomicrographs show the same melting condition at 2650 F on both heating and cooling, and the same completely frozen condition at 2500 F. More liquid is in evidence at 2650 F on cool-

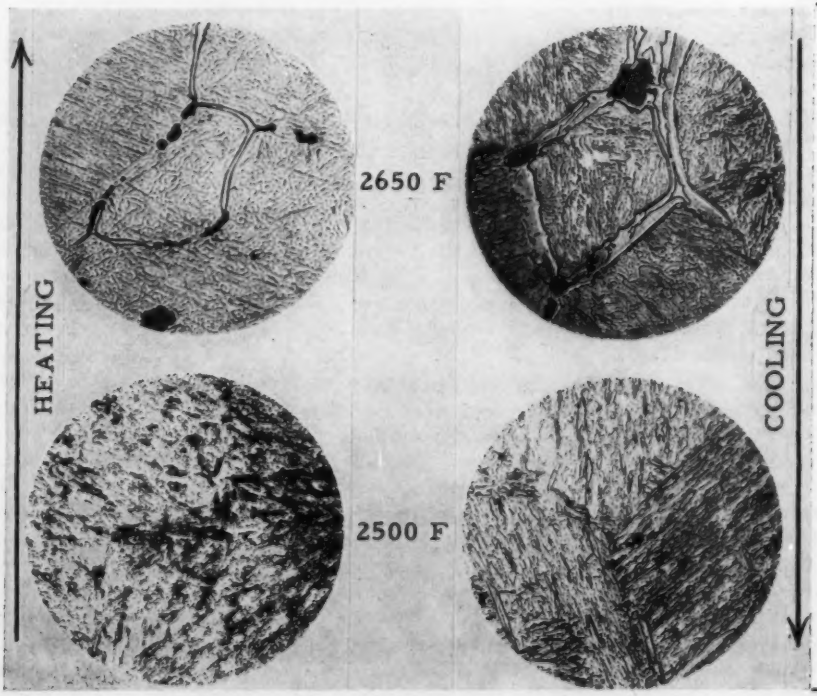


Fig. 13 — Microstructure of 0.50% carbon basic steel in the hot tear temperature range. Etched. Mag. — 1000X.

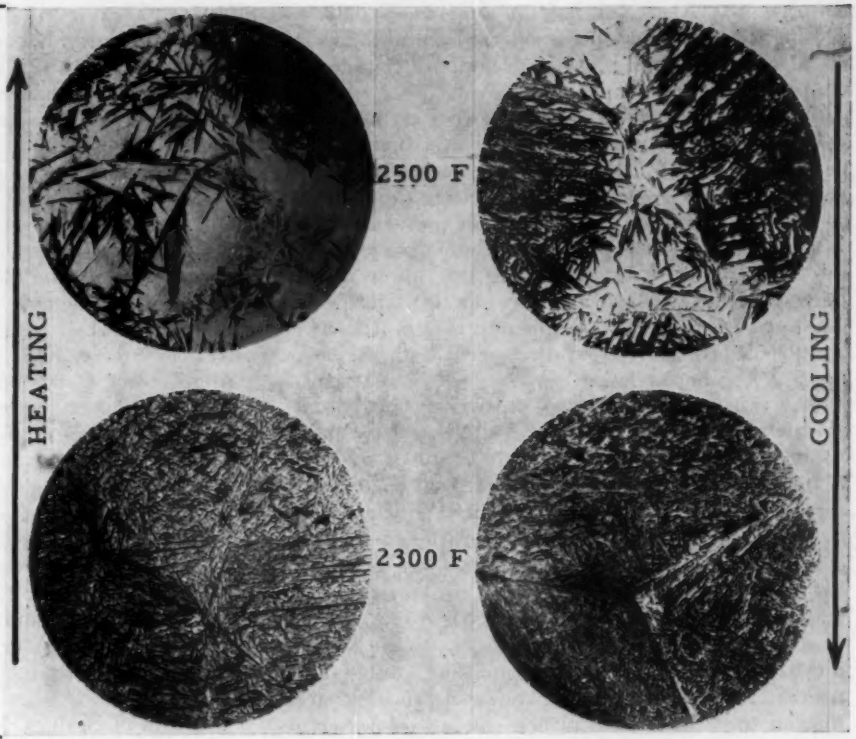


Fig. 14 — Microstructure of 0.90% carbon acid steel in the hot tear temperature range. Etched. Mag. — 500X.

ing from the liquid than is shown upon heating from the solid to the liquid, at the same temperature.

Figure 11 is a 0.30 per cent carbon steel made by the basic electric practice with photomicrographs below and within the hot tear range, on heating and cooling cycles.

The two photomicrographs on the bottom show the structure at 2600 F on both the heating and cooling cycles, and are below the hot tear range. The two at the top are at 2700 F and within the hot tear range. The impact values at these temperatures are 38.0 and 2.0 ft-lb, respectively, as determined by heating.

Again on this steel, the grains show melting at the higher temperature and complete solidity at the lower

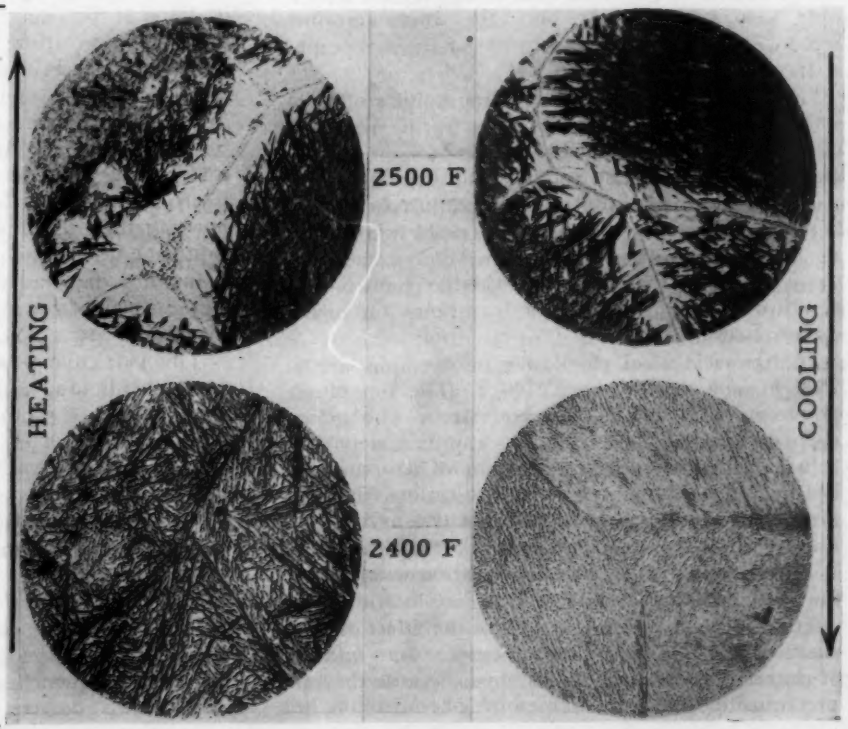


Fig. 15 — Microstructure of 0.90% carbon basic steel in the hot tear temperature range. Etched. Mag. — 500X.

temperature. A greater amount and more defined melting, however, does occur on the cooling cycle at the higher temperature, although both are completely solid at the lower temperature.

It will be noted that the 0.30 per cent carbon basic steel showed this melting condition at a higher temperature than the acid steel. Both the micrographs and the impacts proved that this melting condition is at a somewhat higher temperature for the basic steels.

Figure 12 is a 0.50 per cent carbon steel made in the acid open hearth. Again, the two lower photomicrographs are below, and the top two within the hot tear range. The two temperatures are at 2500 F and 2600 F, respectively, and represent both the heating and cooling cycles.

The two top photomicrographs show grain boundary melting at 2600 F, and give 4.0 ft-lb as determined by impacts on heating to this temperature.

It is obvious that if the steel had been heated to a slightly higher temperature, more melting would have been in evidence. At 2500 F, however, the grains have completely frozen and give impacts of 38.0 ft-lb.

Figure 13 is the same 0.50 per cent carbon grade made in the basic electric furnace. The two bottom photographs are at 2500 F and the top two at 2650 F; they represent both the heating and the cooling cycles. A greater amount of melting shows in the basic steel than in the acid for this carbon content. However, it will be noted that these tests are taken at 50 degrees higher.

Here, as before, the grains are completely frozen below the hot tear range, while they are beginning to melt within the range. Again it can be seen that more melting is present on the cooling cycle than on the heating, as possibly should be expected.

The impact values of the steel at 2500 F and 2650 F were 40.0 and 0 ft-lb, respectively, which certainly agree with conditions shown in the photomicrographs at these temperatures.

Figures 14 and 15 are photomicrographic studies of a 0.90 per cent carbon steel made by the acid and basic processes, respectively. These figures show clearly that heating and cooling from the liquid produce the same condition within and just below the hot tear range. These photomicrographs were taken at 500 magnification and show the tremendous changes that take place with a relatively small temperature change near the hot tear range on higher carbon steels.

On the acid steel the lower photographs are at 2300 F, and on the basic, 2400 F. The top photographs are taken at 2500 F on both steels. The sudden change of impact values from approximately 50.0 ft-lb to 2.0 ft-lb and the agreement of structures on heating and cooling are good indications that the impact test is a good measure of hot tearing susceptibility.

The purpose behind this investigation was to study the characteristics of hot tearing, its location with reference to temperature, as well as the effect of composition and practice. These questions were and are of prime interest to the foundryman because they are prerequisites to means and methods of controlling hot tearing. There is no doubt that the foundryman can greatly improve hot tearing conditions if he first knows what he is dealing with and its characteristics.

For a given practice, the first point that a foundryman should know, as proven in this report, is that hot tearing is not a steel quality proposition that varies from heat to heat. Both the foundryman and the open hearth, or steelmaking department, must learn to separate steel quality from steel temperature. Hot tearing susceptibility for any given method of steelmaking is a uniform characteristic. Whether a steel hot tears or not does not depend upon susceptibility, but upon whether the steel is stressed or not at the wrong time. Neither stress nor hot tear susceptibility are "cooked" into, or out of, an open hearth heat. Stress is only produced on cooling after the steel leaves the furnace, and hot tear susceptibility is a natural phenomenon, uniform for any given practice.

The foundryman is confronted with the difficult proposition, whether he admits it or not, of avoiding stress concentrations at the wrong time-temperature. Those temperatures are explained and tabulated carefully in this report. The only help that he can get is the tapping and pouring temperature level supplied by the open hearth or other steelmaking furnace. He cannot rely too heavily upon this, because he may be pouring for long periods of time with various types and sizes of castings. He can only be supplied with a practical temperature level which will do the job as a whole. There is no pouring temperature level that will not generate a stress in intricate castings with light and heavy sections, nor is there a temperature that eliminates hot tear susceptibility. There is, however, a proper temperature level of pouring which is best for him and one that makes it possible for him to do his job correctly. A steel must be hot enough to feed and at the same time low enough in temperature to respond to chilling effectively.

Heading, gating, and chilling are all factors in feeding a casting for soundness, but they are also factors instrumental in generating stresses which produce hot tearing. When feeding a casting, certain sections of that casting precede other sections in cooling and freezing. It makes a big difference when, where, and how fast these sections freeze, while other sections are liquid or near liquid.

Just how fast and far certain sections have to cool without being stressed is of vital importance in avoiding hot tearing.

This particular point is brought out clearly in Fig. 16. In this figure is shown the effect of cooling rate in cooling from the liquid state into and through the hot tear range. These conditions are also compared micrographically with heating up to and into this range.

Turnings were taken from a 0.50 per cent basic carbon steel and heated in a refractory boat under an argon atmosphere to 2800 F until they were in the liquid state.

In one case the steel was cooled extremely slow, that is, two degrees per minute to 2650 F, held ten minutes and quenched in ice brine. Another melt was cooled two degrees per minute to 2500 F and

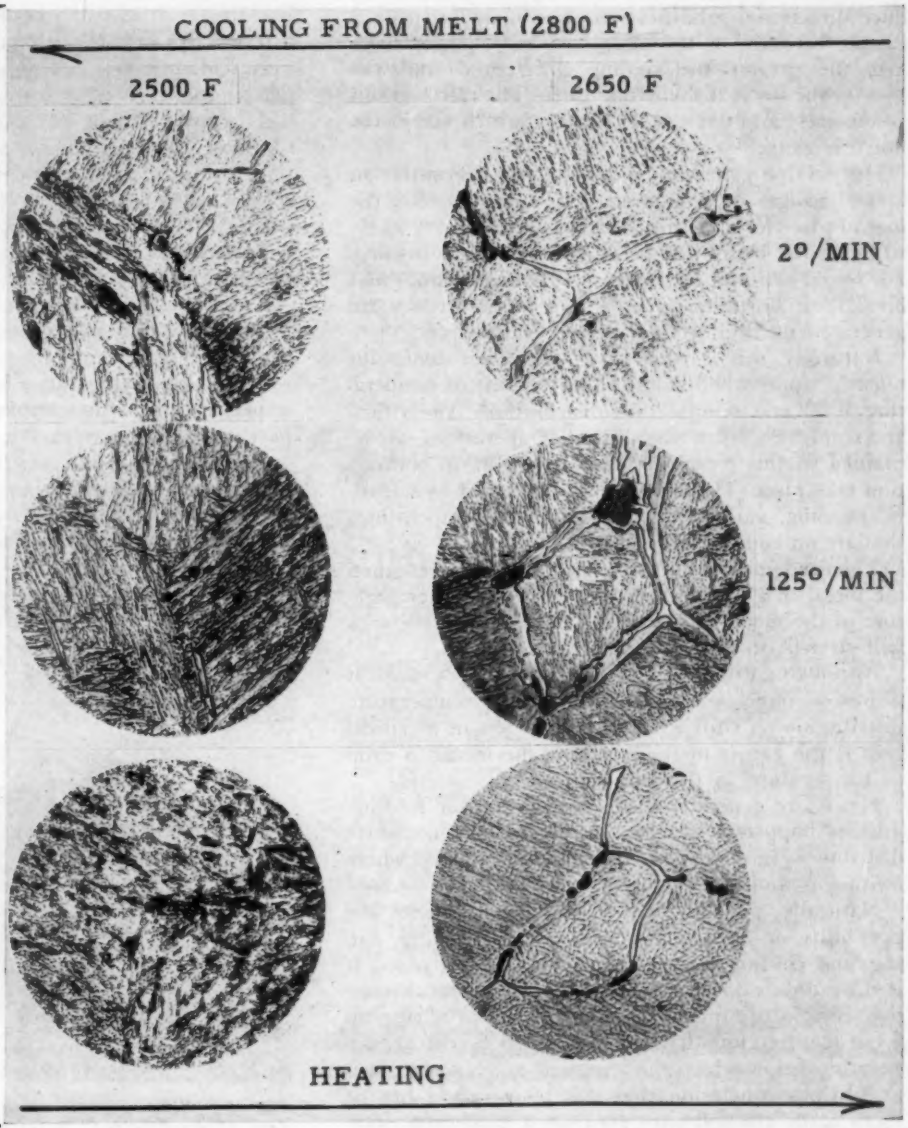


Fig. 16 — Effect of cooling rate on hot tear temperature range. Photomicrographs are of a 0.50% carbon basic steel. Etched. Mag. — 1000X.

quenched in ice brine. In other words, the steel was cooled from the liquid to the hot tear range in one case and into the ductile range in the other, at an extremely slow rate.

These experiments were duplicated using a cooling rate of 125 degrees per minute, an extremely rapid rate.

In Fig. 16 the two top photomicrographs are the structures produced under slow cooling. The two center photos are those of very fast cooling.

As a comparison, the bottom two photomicrographs are the structures produced on the same steel by heating to the same respective temperatures.

All photomicrographs in this figure were taken at 1000 magnification.

Neither heating to temperatures within or just below the hot tear range, nor cooling from the liquid at widely different rates displaces the position of the hot tear range. All samples show melting at 2650 F and complete freezing at 2500 F.

Since the impact values of the steel on heating were 38.0 ft-lb at 2500 F and zero at 2650 F, the similarity of the structures with that of cooling would indicate that the same change would take place physically on cooling.

From this figure it can be seen that the temperature to which a steel is cooled is far more important than the rate of cooling. In fact, the rate of cooling apparently has no effect whatsoever.

After a casting is poured, every section of that casting must cool from the liquid state to room temperature. It, therefore, must go through all existing stages of brittleness and ductility during that cooling cycle. Since steel contracts as it cools, all steel is under stress during this cooling cycle. If the entire casting cooled evenly and there was no resistance from the mold, the casting would never hot tear. However, we cannot avoid complete mold resistance, nor can we avoid unequal cooling. In order to feed a casting and make it sound, we must always have unequal cooling to pro-

duce directional solidification. The more intricate a casting becomes, or the more sectional changes there are, the greater the cooling differential and the greater the stress at the wrong time. The thin sections freeze early and stress other sections which are in the hot tear range.

Hot tearing generally occurs at stress concentration areas, changes of section, etc. Not only is this the logical place for a casting to strain, but it is very likely to be in the brittle temperature stage when stressed. For sound castings, all concentrated stress sections and areas must be in the ductile zone when stresses are generated by thin section cooling contraction.

Naturally, this is more easily said than done, but must be approached from the standpoint of temperature level and temperature distribution. All critical areas must be below the critical temperatures, as explained in this report, when stresses due to contraction take place. This must be accomplished by scientific heading, gating, and chilling, plus temperatures that are no higher than necessary.

The fundamental duty of a chill is not to strengthen the metal in a certain area, but to drop the temperature of the metal into a plastic zone, so that the area will stretch instead of break when stressed.

Although gating is a means of filling up a mold, it is just as important to gate for proper temperature distribution. A chill will only do its job in a critical area if the gating method permits the metal to drop in temperature in this location when it should.

Heads are generally used as reservoirs for feeding. Just as important are their value as a temperature distributor. In many cases heads should be used where feeding is unnecessary.

Naturally, pouring temperature levels control the possibility or impossibility of effective heading, gating, and chilling for the elimination of hot tears. It is the author's opinion that the lowest practical pouring temperature increases the possibility of setting up a hot tear-free foundry. Every steel has a critical temperature below which the steel will not hot tear. It is an obvious conclusion that this temperature can be more easily reached with controlled maximum pouring temperatures.

Conclusions

This report has dealt primarily with the principles of hot tearing, the emphasis being placed upon the fact that it confines itself to definite temperature ranges which are controlled by the carbon content and by the heterogeneity of carbon set up during freezing. The report points to the fact that we are simply determining the melting point of the highest carbon content produced for any given steel during the process of freezing.

The section of the report covering the uniformity of results obtained upon heating to and cooling from the liquid indicates clearly that the range of hot tearing in casting production confines itself to a relatively narrow band of temperatures where the steel is definitely molten around the grains.

An important point shown in this report is the fact that when the steel drops below the relatively narrow temperature band where the last heterogeneous carbon freezes, it instantly becomes non-breakable under a severe dynamic test for many hundreds of degrees. Very seldom does a casting receive more than a small percentage of such a concentrated stress condition.

This investigation was a study of hot tearing under the most severe design conditions. No one will question the fact that severe, concentrated stresses are created when the notched Charpy impact specimen is broken. We believe that brittle and ductile zones established under these conditions are true zones, and unquestionably establish true conditions.

Examples of the conditions brought out in this type of test are shown in the following photographs taken from a 0.30 per cent carbon acid steel heat.

Figure 17 is illustrative of a steel tested near the upper transformation temperature, 1400 F. The impact value here may vary considerably depending upon the composition. Any element that affects the critical for a given carbon content also affects the amount of strength and ductility. It is during this transformation that the last rupturing takes place prior to hot tearing.

Austenite is extremely tough at temperature and displays a marked resistance to notch sensitivity. While the impacts may vary many foot-pounds during transformation, once in austenitic solution all values

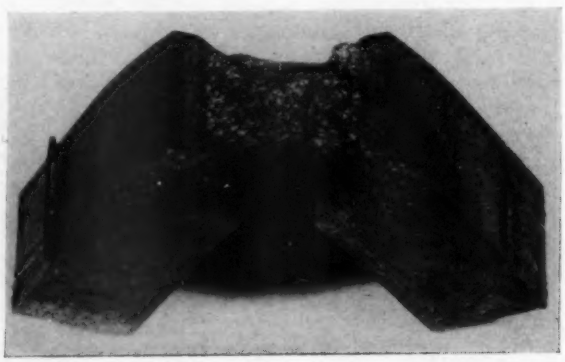


Fig. 17 — Illustration of impact specimen at 1400 F showing type of fracture near transformation temperature. In this range the impact values are extremely high and tearing occurs within the grains.

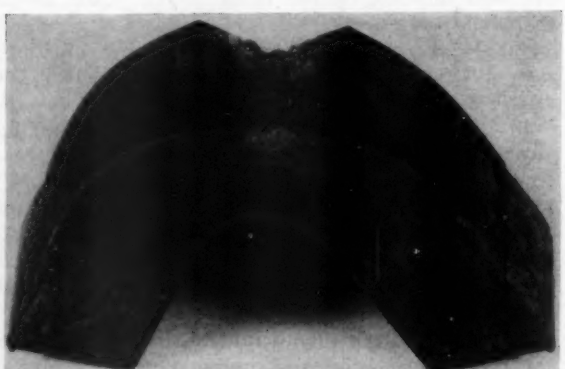


Fig. 18 — Illustration of impact specimen at 1800 F. In this temperature range all steels under 1.30 per cent carbon are resistant to notch sensitivity and give consistent impact values.

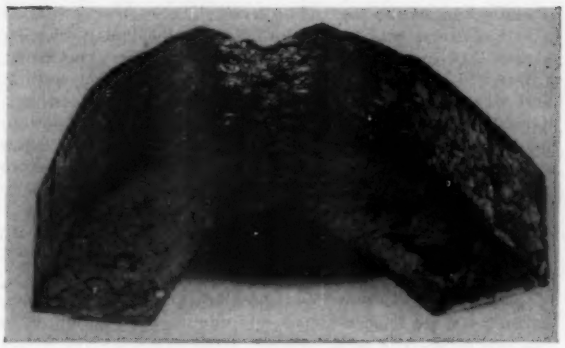


Fig. 19 — This illustration represents the appearance of all steels directly under the hot tear range. In this range the steels are almost completely resistant to rupture.

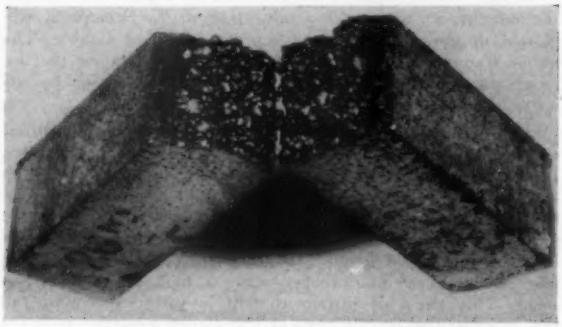


Fig. 20 — This illustration represents the brittle appearance of steels in the hot tear temperature zone. All failures occur around the austenitic grain boundaries. The appearance of the fracture and the size of the grain are influenced by carbon content and by nucleating effect of deoxidizers.

become constant at temperature, regardless of composition.

Figure 18 is a 0.30 per cent carbon acid steel tested at 1800 F. The appearance of this test is common to all steels at this temperature up to 1.30 per cent carbon and, as explained previously in this report, all have practically the same Charpy impact, approximately 100 ft-lb. Realizing the severe shock these tests receive, it displays almost complete resistance to rupture.

The appearance of tests taken immediately under the hot tear range is illustrated in Fig. 19. This particular test was taken at 2500 F on a 0.30 per cent plain carbon steel. It duplicates, both in foot pounds and appearance, a high chrome, a Cr-Ni-Mo, or any other steel of similar carbon content.

This steel, tested right under the hot tear range, is seen to be almost completely resistant to failure, although the stress is all concentrated at a single sharp notch.

Figure 20 shows the sudden change that takes place the instant the grain boundaries lose their strength. This test was taken at 2600 F, approximately 100 degrees higher than that in Fig. 19.

The above-mentioned figures represent the possible physical property conditions that may exist in any single intricate casting soon after it is poured. Even though a wide temperature differential may exist, only that part of the casting which is within approximately 100 degrees of the freezing point is susceptible to tearing.

Basically, the solution to hot tearing is to avoid wide temperature differentials during freezing. With wide temperature differences, high stresses and high stress concentrations exists on intricate designs and section changes. These stresses cause hot tearing in areas not yet past the brittle temperature zones.

Castings must be headed, gated, and chilled with the specific purpose in mind of avoiding wide temperature differentials, thereby minimizing stresses and avoiding brittle temperature zones at sectional changes or stress concentration areas.

The elimination of hot tearing in steel castings is a task that has confronted the foundry industry for many years, and will always be a problem. We hope that the contents of this paper will add to the progress that is being made in this field.

Acknowledgments

No report such as this could have been prepared without the valuable advice and assistance of others. In particular, the author is indebted to Mr. P. F. Cook of our Chicago Works for his assistance in preparing all of the photographs and in carrying out much of the heat treating work which was necessary to obtain data. Then, too, much of the planning and preparing for the actual research work was the result of conferences with our foundry supervisors at Chicago Works, notably Mr. F. Soderstrom and Mr. L. Chick.

DISCUSSION

Chairman: C. K. DONOHO, American Cast Iron Pipe Co., Birmingham, Ala.

Co-Chairman: W. O. IGELMAN, National Malleable & Steel Castings Co., Melrose Park, Ill.

Secretary: D. N. ROSENBLATT, American Foundry & Machine Co., Salt Lake City, Utah.

C. E. SIMS[1] *(Written Discussion):* The work so ably reported by Mr. Christopher is a distinct contribution to our knowledge of the hot tearing defect in cast steel; the causes and possible remedies. His choice of the Charpy bar as a test specimen appears to have been well advised. It is a convenient, rapid, and apparently an accurate means of obtaining data.

Mr. Christopher concludes that hot tears form in cast steel when stress is applied before the steel is completely solidified and while there is still some liquid in the grain boundaries or dendritic interstices. This is in line with previous work and to which the new data give considerable support. Since the excellent work of Briggs and Gezelius* at the Naval Research Laboratory, more than 20 years ago, the bulk of the evidence has indicated that hot tears in castings occur before the steel has reached its true solidus temperature.

During the past several years, some work has been going on at Battelle Institute which is pertinent to this subject and which I will review very briefly. In many respects, the work confirms Christophers observations but differs in others. The work was actually on an investigation of the causes and prevention of hot cracks in high strength weld deposits and has been reported in a paper to the Welding Society**. A weld deposit, however, is a small casting, and it is believed that hot

1. Technical Director, Met. Dept., Battelle Memorial Institute, Columbus, Ohio.

*Briggs, C. W., and Gezelius, R. A., "Studies on the Solidification of Steel Castings Free and Hindered Contraction of Cast Steel," TRANSACTIONS, AFS, v. 42, pp. 449-476 (1934).

**Sopher, R. P., Jacobs, A. J., and Riepple, P. J., "Investigation of Weld Metal Cracking in High-Strength Steel," *Welding Journal*, Research Supplement (November 1955).

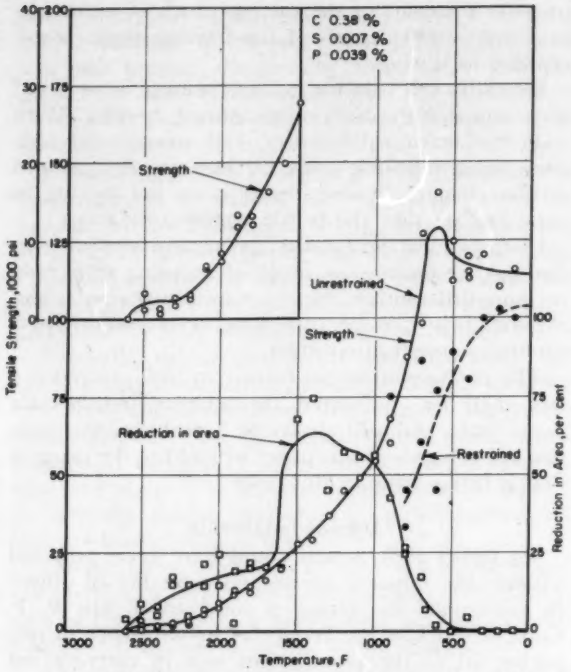


Fig. A — Hot strength and ductility properties of high-phosphorus SAE 4340 steel.

tears in weld deposits and steel castings are very similar and are caused by similar conditions.

The test used was a hot-tension test and the specimens 1½-in.-round bars of the steel to be tested. The bars were encased in a close-fitting tube of fused silica with the ends projecting so that they could be gripped for tensile testing. The center portion of the bar was heated rapidly by induction until a short section was melted, and then the induction current was stopped and the specimen allowed to cool. The molten portion could not flow away because of the restriction of the silica jacket. The cooling was rapid because of the steep temperature gradient. The operation was designed to simulate welding conditions, where metal is deposited in a molten state and quickly chilled. A platinum-platinum-rhodium thermocouple placed in a centrally drilled hole recorded the temperature of the hottest zone.

The heating and cooling cycle to above the melting temperature and cooling to 2500 F took one minute, or 1½ minutes to cool to 1500 F. At a predetermined temperature on the cooling cycle, the bar was loaded by means of an air cylinder and rapidly broken in tension. The breaking load and ductility were measured.

Most of the steels tested were of the SAE 4300 series. Figure A shows the results obtained on a low-sulphur steel for strength and ductility vs. testing temperature. All of the data points are plotted to show the degree of reproducibility obtainable. The tensile strength of the restrained bar is lower than for the unrestrained one. This is believed due to damage caused by the restraint before the breaking load was applied.

Figure B shows the results obtained in steels of four different carbon contents with the rest of the composition similar. It will be noted that both strength and ductility are acquired at a markedly higher temperature for the 0.20 per cent carbon steel than for the one containing 0.59 per cent carbon. The 0.46 per cent carbon steel seems slightly better than the 0.33 per cent carbon steel, but there is little difference between these two and both are midway between the high and low carbon steels. This effect of carbon content is fully expectable and can be explained on the basis of the carbon lowering both the liquidus and solidus temperatures of the steel. They could

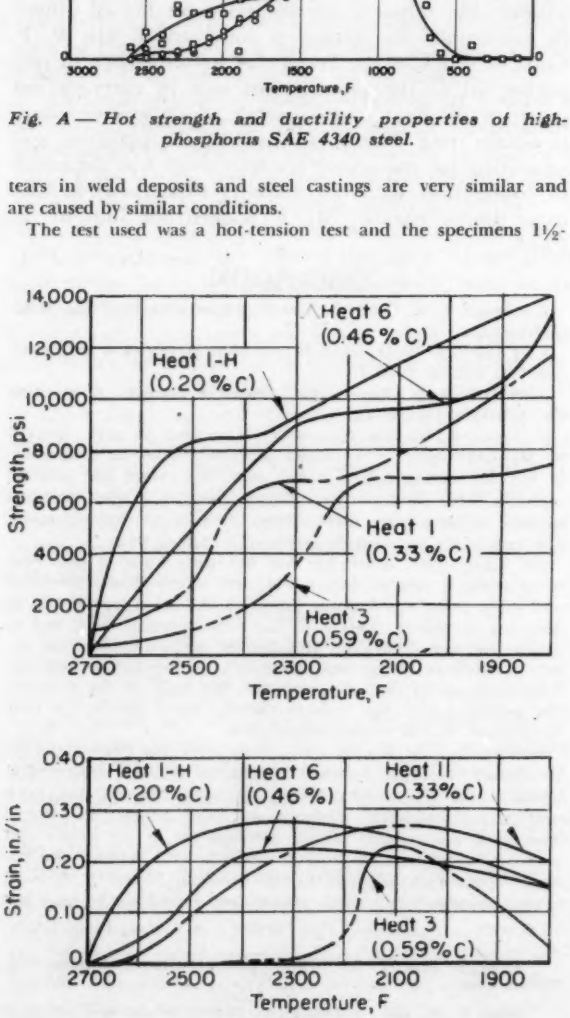


Fig. B — Comparison of hot strength and ductility of experimental SAE 43XX-type steels, showing effect of carbon content (P and S less than 0.010% in all heats).

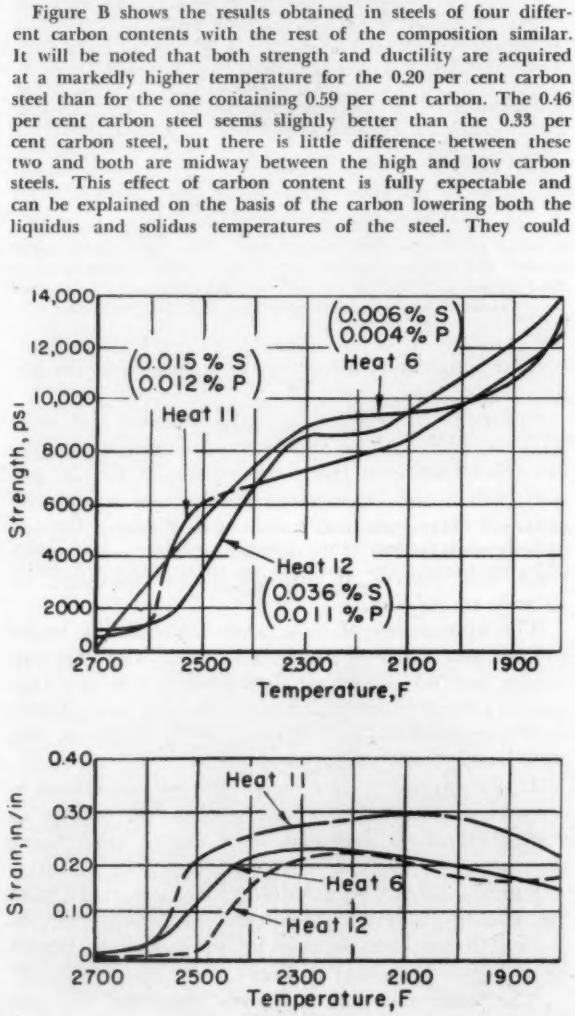


Fig. C — Comparison of hot strength and ductility of experimental SAE 4340 steels, showing effect of sulfur content.

not be expected to gain strength and ductility until fully solidified. This corroborates the observations of Christopher.

All of the steels tested in high temperature tension were also tested for propensity to weld cracking, using a modified Lehigh restraint test. In this test the high-carbon steel showed only slightly greater tendency to form hot cracks than did the low-carbon steel despite the difference in behavior shown in the hot-tension test. The whole solidification sequence was merely moved to a lower temperature level.

Figure C shows the results obtained on steels of three sulfur levels, all of low-phosphorus content. These steels have ostensibly the same liquidus and solidus temperatures. The important effect of sulfur is believed to be on the maximum temperature at which appreciable strength and ductility is attained: It will be noted that this temperature was 100 F higher for the low-sulphur steel. The intermediate-sulphur steel was not much different than the low-sulphur steel, indicating that adverse sulphur effects are obtained only above about 0.015 per cent. In the weld cracking tests, the low-sulphur steel could stand four times as much restraint without cracking as could the highest sulphur steel. The steel containing 0.015 per cent sulphur was intermediate but closer to the low sulphur steel in its cracking behavior.

Figure D shows the effects of three levels of phosphorus in steels of very low sulphur content. The phosphorus has a small effect on the initiation of hot strength but a large effect on the development of hot ductility. Here also, there appears to be little effect of phosphorus up to 0.017 per cent, but 0.039 per cent phosphorus caused a marked retardation of ductility down to 1600 F.

In the weld cracking tests, the low-phosphorus steels showed definite superiority. Heat 2 could stand a restraint of 8 in. and Heat 15 a restraint of 6 in. without cracking, whereas, Heat 13 cracked at a restraint of less than 2 in.

In Fig. E are shown the results obtained on three ingots of a split heat, two of which had misch-metal additions. It may be seen that at 2500 F the addition of misch metal produced a progressive increase in ductility. Inasmuch as the misch-metal also caused a progressive decrease in sulphur, however, any direct effect of misch-metal cannot be distinguished from the indirect effect of decreasing sulphur. In the weld cracking, test 17C, with the lowest sulphur content, required a restraint of 7 in. to cause cracking, while 17A, to which no misch-metal was added, cracked at less than 2 in. restraint.

To obtain more information on the direct effect of misch-metal, two heats of the same sulphur content (0.015 and 0.014 per cent) and only one of which had a misch-metal addition of 6 lb-per ton. were compared. In the hot tension test, Heat A, the control heat, had zero ductility of 2600 F, while Heat B, with the misch-metal, had 50 per cent reduction of area. At 2500 F, both steels were equal at 70 per cent reduction of area. In the weld cracking test, the control Heat A cracked at 5 in. restraint, while the misch-metal heat needed 7 in. restraint to cause cracking.

It should be pointed out that these tests were on a limited range of compositions, but it is believed that they indicate no great effect of carbon in the normal range of composition, but deleterious effects of both sulphur and phosphorus on hot cracking of steel. They also indicate that misch-metal additions are beneficial.

W. K. Bock[2] *(Written Discussion):* The fact that hot tearing occurs at temperatures just under the solidus has been known for some time. The people at the Naval Research Laboratory have been saying that hot tears are associated with molten grain boundaries. It remained for Mr. Christopher to tie the phenomenon to the equilibrium diagram of the iron-carbon system.

Although Mr. Christopher has very ably handled his research, there are certain disquieting results which cause us to look for alternate interpretations. Perhaps the most important question is in connection with the difference between acid and basic melting.

The author shows that the liquid grain boundaries can occur because of a segregation of carbon to the grain boundaries. In Figs. 8 and 9, the lower limit of the hot tearing bands gives the freezing temperature of the segregated, higher-carbon metal.

According to Fig. 9 then, the segregation in steels made by acid practice results in higher-carbon grain boundaries (more

2. Manager of Metallurgical Research, National Malleable and Steel Castings Co., Cleveland.

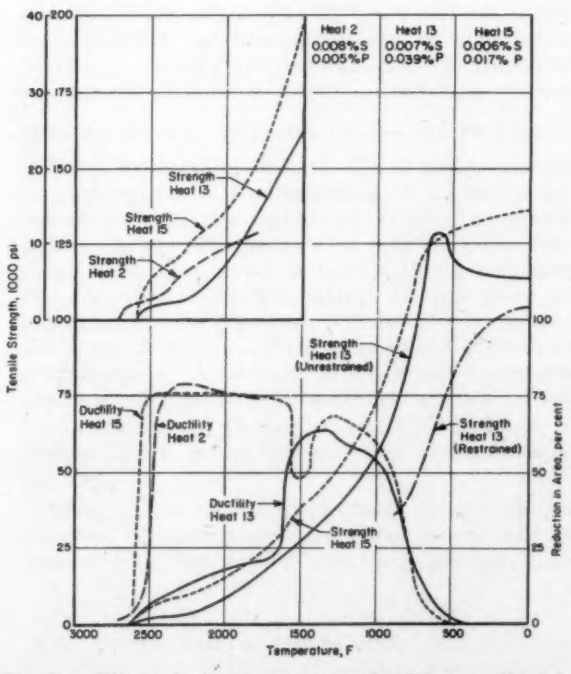


Fig. D — Effect of phosphorus content on hot strength and hot ductility of SAE 4340 steel.

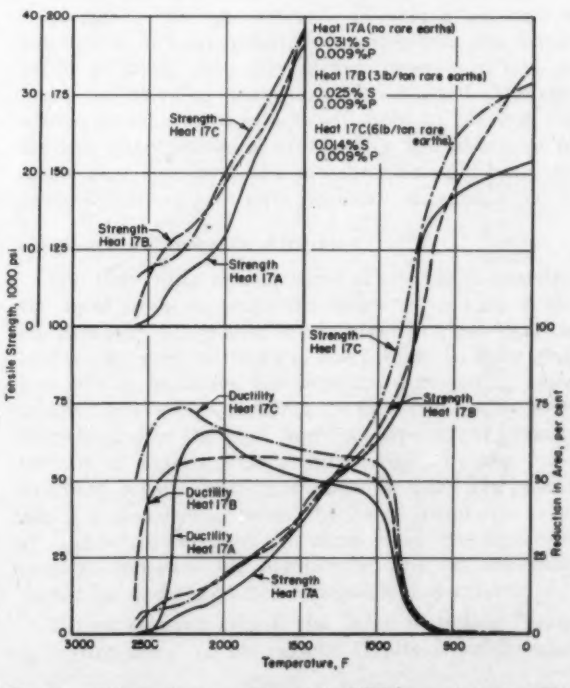


Fig. E — Effect of rare-earth metal additions on hot strength and hot ductility of SAE 4340-type steel.

segregation) than in steels made by basic practice. How does a steel, cooling in the mold, know how to behave unless the melting process has introduced certain influences like nucleation effects, for example?

The heterogeneity explanation presents certain other interesting questions. How does it happen that the segregation in wrought steel J is as great as in cast steels? Why was steel J included with the acid steels when its sulfur and phosphorus contents indicate that it could well have been a basic steel?

Another question arises in connection with heat D. What happened to make a 0.29 carbon steel act like a 0.90 carbon steel?

Mr. Christopher puts more faith in equilibrium diagrams for practical problems than I do. Therefore, I find it easier to regard the lines which form the lower limits of the hot tearing ranges in Fig. 9 as the limits of super-cooling, as determined by the conditions of the experiment. The supercooling idea would make segregation a sufficient but not a necessary condition for the presence of liquid grain boundaries.

Acceptance of the supercooling theory would not divorce the hot tear area from the solidus of the iron-carbon diagram, but it would mean that the lower limit would not be definite and would depend on the conditions of cooling and freezing. Remembering our experience with heat D, there can be a question of whether or not, under more general foundry conditions, the difference in hot tearing temperature of basic and acid steel will always obtain.

Another question of interest concerns the susceptibility to hot tearing. We can compare the results of Mr. Christopher with those of Middleton and Protheroe published in the 1951 *Journal of the Iron and Steel Institute*. The English workers found that alloying elements do not generally change the temperature at which hot tearing occurs. Although they missed the effect of carbon because of a restricted range, their result agrees with Mr. Christopher's.

Middleton and Protheroe show that the force needed to produce a hot tear can be changed by changes in alloy. This point is not apparent in the present work.

There is ample evidence in this paper and others that hot tearing is a problem in mechanics of fracture and not a problem of mere transport of material. The liquid grain boundaries serve to reduce the effective cross section on which rupture occurs and perhaps to supply notches.

As Mr. Christopher points out, hot tearing occurs where multiaxial loading is present. How severe the embrittling condition of this type of loading will be depends on the particular circumstances in a given problem. Experience in testing at room temperature teaches that several materials may act alike in a mild test and behave quite differently in a severe test.

The Charpy test at elevated temperatures is not a severely embrittling test. This point is illustrated by Fig. 19. It is common experience that designs often produce a degree of embrittlement which is difficult to reproduce in laboratory tests.

The Charpy test seems well adapted to determining the temperature at which tearing occurs. Whether or not it is adequate to determine the susceptibility to hot tear in a given casting is open to question, because the susceptibility depends so largely on the response of the solid metal to the embrittling conditions encountered.

PEOPLE — THE OTHER SIDE OF THE AIR POLLUTION PROBLEM

By

E. M. Adams*

A specific air pollution problem invariably appears to plant management in the form of a complaint. Someone living or working in the vicinity of the plant calls or writes a letter expressing objections to some dirt, odors, or fumes. Or, in some localities a public control official calls to report complaints or a violation of regulations. All this is so obvious as to merit no comment except that it does reveal the nature of the air pollution problem.

The Dual Nature of the Problem

A little consideration leads to the realization that it is the complaint against a pollutant which creates the problem; it is not the pollutant alone. Any emission may continue indefinitely without there being a problem — until someone objects. Appreciation of this dual nature in turn leads to the thought that both aspects should be dealt with. Not only should attention be devoted to the emissions and their control, but, also, attention should be devoted to the attitude of the people who have complained or who may complain. Such attention can be expected to prevent or to mitigate problems and to assure that any control equipment actually achieves its purpose.

How the Attitude of People Creates the Problem

The importance of this attitude of people can not be overemphasized. Understanding of the extent and manner in which this attitude enters into the problem is basic in the attainment of a solution. The complaint has always been associated with the nuisances, dirt and odors. Over the country at large there are great variations in the amounts of dirt blowing around in the air. Even in the dirtiest localities there may be no complaints. A nuisance arises, however, and the degree of dirtiness becomes excessive, whenever a sufficient number of persons object. This happens regardless of the actual concentration of materials in the air.

Again, in the case of odors, there are no objective standards of what is excessive. Some people will accept or even like a particular odor, whereas other people will object. Regardless of the substance in the air, and regardless of the nature of the resultant odor, there is no problem until people object.

Even in the case of acknowledged damage to property, animals, and vegetation, it is the objection of people to this damage that creates the problem. The killing of trees on company property or perhaps in uninhabitated areas brings no objection and causes no problem; the killing of such trees in a residential area brings complaint and a problem. The mere thought of injury to human subjects is the cause of many objections, and problems, at the present time.

Even in localities having statutory regulation of emissions the attitude of people determines acceptable levels of contamination. The regulations do not prescribe what the people want, but rather they have the purpose of bringing about the conditions desired.

Treat Both Aspects of the Problem

Accepting that the attitude of people determines whether or not an industrial management has a pollution problem, then this attitude deserves at least as much attention as does the emission itself. Management can take steps which will help to prevent the attitude from becoming unfavorable and also steps to assure that any corrective measures actually have the desired effect of gaining a favorable attitude.

Cultivate a Favorable Attitude

The first thing management should do is examine the local situation and take measures to keep plant emissions at acceptable levels according to location and to changing attitudes of the people. In increasing numbers of localities, the people are becoming more critical and are demanding cleaner conditions than have existed in the past. Such positive action permits control at management's convenience, by the most practical means, and at minimum cost. The plant which is ahead of or with "the field" is not apt to be at "the whipping post." In many cases, the degree of control necessary for acceptance will be less than would be necessary after a problem has arisen.

Of equal importance is the value of telling "management's story" to the people. Efforts at public edu-

*The Dow Chemical Co., Midland, Mich.

cation which show that management is honestly concerned about pollution and which show that emissions are restricted can be expected to gain a considerable good will. Information regarding the general nature of plant emissions, their possible effects, and of what control is practical can be expected to gain understanding of management's position.

Give Attention to Complaints

Closely related to this giving of information is the handling of personal complaints as presented by individuals or by officials. At all costs, these should be handled promptly and with sincere efforts to supply answers. When these matters are answered properly, they also gain good will and understanding. Having these, management can expect to find a higher level of acceptance of emissions, to avoid unjustified complaints and excessive demands, and to be spared highly restrictive control.

It is desirable to have complaints come directly to management rather than through public officials. In the latter case they are likely to reach the newspapers, become a public matter with overemphasis and multiplication of the problem. To encourage direct communication some companies actually canvass surrounding areas, invite questions and comments, and leave the name of a company representative who may be contacted with complaints of any kind which may arise in the future. Such personal contact may be time-consuming, but it can hardly be excelled for building confidence and good will.

Treat the People as Well as the Emission

Finally, keeping the public informed of progress is as important as the actual installation of control equipment in the handling of any pollution problem. People are not analytical machines and of necessity they can judge the state of the air only by what they see, smell, taste, and feel. Changes short of complete elimination may not be recognized or acknowledged.

Persons involved in a prolonged period of complaints and in court actions frequently demand drastic abatement; they may be suspicious and even vindictive. Such persons will not be satisfied simply by the installation of some control machinery. These personal aspects can be handled only by the giving of information, by explaining what has been done, by demonstrating the performance of equipment, by showing that control measures have been taken. *Without this, the installation of equipment may fail to accomplish its purpose – that is, the reduction of an emission to acceptable levels.*

DISCUSSION

Chairman: R. L. McILVAINE, National Engineering Co., Chicago, Ill.

W. C. WAGNER[1] *(Written Discussion):* The author has clearly outlined the fundamentals of the individual problems and has given good advice as to how to proceed where objectionable conditions arise from a given source.

Very probably Dr. Adams should not, in a brief presentation, cover the cumulative effects upon the people. However, experience indicates that most generally, more than one source of objectionable emissions may contribute, and in some situations, even small additions may "break the camel's back" from the viewpoint of the people adversely affected. It, in general, appears necessary for those who may contribute to objectionable levels of contamination, to join together in following Dr. Adam's advice concerning relations with the public.

Locally, the formation of an air pollution control committee by our Chamber of Commerce, with special subject subcommittees, has served splendidly in guiding the progress of proposed City and State legislation, to the mutual advantage of both those who are responsible for causes, those, the public, who are affected, and the regulatory authorities directly concerned.

H. C. BALLMAN[2] *(Written Discussion):* Dr. Adams has keynoted the "new look" for industry toward its neighbors when he suggests that plant managements stay ahead of complaints and tell their story. Twenty or thirty years ago this paper could not have been given or, if delivered, would have been disbelieved for then industry's attitude was—tell them nothing, we will do as we please. Industry has grown up and through management has learned (quite often the hard way) to be a good neighbor.

Few of us—even those in close contact with industry—are fully aware of the great change of attitude by industry toward its potential sources of nuisance. Unfortunately we are all prone to reflect back on the days when industry was pictured as a huge, ponderous, indifferent giant that crushed the little man. Let us therefore take a cue from the paper and accept the change evidenced by industry in recent years in their attempts to be a good neighbor, thereby approaching nuisance problems on the premise that something can and will be done to obtain relief.

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1. Consulting Engineer, Chamber of Commerce of Greater Philadelphia, Philadelphia.
2. Executive Secretary, Air Pollution Control Association, Pittsburgh.

AGE-HARDENING CHARACTERISTICS OF A CAST ALLOY OF COPPER – 5.8 PER CENT TITANIUM

By

N. Hehner*, H. McCurdy*, and R. Edelman*

ABSTRACT

The copper-rich end of the copper-titanium binary phase diagram represents the classical form of an age-hardenable alloy. The maximum solid solubility of titanium in copper is 6.0 per cent at about 1625 F. At room temperature the total amount of titanium held in solution by copper is less than 0.5 per cent. The optimum mechanical properties for a cast copper-5.8 per cent titanium alloy were obtained after solution treating at 1625 F, water quenching, and then age-hardening at 800 F. The average properties obtained are listed below:

Ultimate tensile strength, psi	— 121,400
Yield strength (0.2% offset), psi	— 107,300
Elongation (2-in. gage length), per cent	— 8.0
Reduction in area, per cent	— 18.0

Literature Survey

The most recent work on the determination of the copper-titanium phase diagram has been done by Trzebiatowski[1] and is shown in Fig. 1.

During the early 1930's several investigators studied the copper-rich end of the copper-titanium binary alloy system. Hensel and Larsen[2] made a thorough study of the age-hardening characteristics, resistivity, and mechanical properties of a series of copper-titanium alloys. They reported mechanical properties of alloys containing no more than 2.90 per cent titanium. These alloys were made with titanium of 95.4 per cent purity. Work using purer titanium was carried out by Schumacher and Ellis,[3] but they too did not experiment with alloys containing more than 3 per cent titanium. This investigation was concerned mainly with resistivity measurements. However, the authors did note that the same age-hardening characteristics were obtained in alloys prepared from impure grades of titanium.

Another investigator, Crampton,[4] reporting on the binary system found that an increase in hardness and an improvement in strength was obtained on solution-treated and aged copper-titanium alloys containing up to 4.7 per cent titanium. He also found the maximum alpha solubility to be 3.0 per cent titanium at 1650 F. A patent by Kroll[5] describes

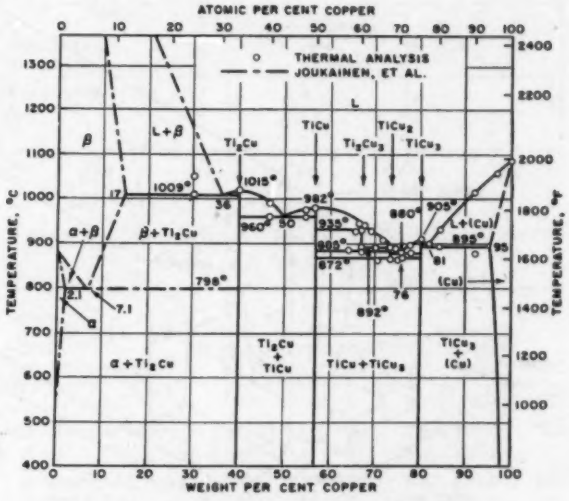


Fig. 1 — System Ti-Cu after Trzebiatowski et al.[1]

the best heat-treating cycles for a 3.0 per cent titanium copper alloy to obtain optimum mechanical properties. These properties were 104,000 psi tensile strength, 30 per cent elongation, and 60 per cent reduction of area.

During this period studies were also made on ternary and quaternary systems. Copper-rich alloys with titanium in combination with beryllium and/or silicon were found by Comstock[6,7] to be age-hardenable. Pilling and Merica[8,9] hold two patents on age-hardenable ternary copper-nickel-titanium alloys. The titanium content covered by these patents ranges from 1 to 10 per cent.

Many of the investigators reported some difficulty in preparing the alloys because of the reactive nature of titanium. Comstock,[10] in an attempt to minimize this difficulty, suggested the use of a protective slag containing calcium fluoride in combination with other alkali and alkaline earth halides.

The most recent discussion of the copper-titanium binary age-hardenable system is in an article by Y. Mishima.[11] This article, reviewed only in ab-

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stract form, mentions the use of titanium scrap and low grade sponge in making a high-strength copper-titanium alloy.

Procedure and Methods

Preparation of the Alloys

All the heats were melted in air using a 3000-kilocycle induction furnace with a graphite crucible. Bar copper of 99.9 per cent purity and titanium in the form of a 50 per cent copper–50 per cent titanium master alloy were employed in making the binary. The 50-50 master alloy was prepared in an arc melting furnace in an argon atmosphere[12] and utilized bar copper of 99.9 per cent purity and commercial purity titanium sponge. Graphite was again used as the mold material.

All test heats of copper-titanium were cast into sand molds having a keel block configuration,[13] Fig. 2. Pouring temperatures were maintained between 2300 and 2350 F. The test specimens were cut from the legs of the keel block to insure the highest soundness. Fluidity appeared to be poorer for the alloy than for pure copper in the molten state. As the pouring temperature was raised to 2500 F, fluidity appeared to become even poorer. Some scattered experiments have been made on the use of fluxes, titanium scrap, and various types of molds. However, more work must be done to determine accurately the influence of these variables.

Heat Treatment

Solution treatments were applied in a vacuum furnace under a pressure of a few microns of mercury. The specimens were quenched into an external water bath rapidly to retain the high temperature structure without any appreciable surface contamination. Aging treatments were done in an argon atmosphere. Again the quenching rate was rapid, and surface contamination was negligible.

Mechanical Testing

The cast bars were solution-treated prior to machining to standard 0.505-in. tensile specimens. They were then aged and tested at a strain rate of 0.05 in./min. Ultimate tensile strength, yield strength (0.2 per cent offset), elongation (2-in. gage length), and per cent reduction in area were the properties that were measured. Duplicate samples were run and there was good agreement between the samples.

Metallographic Procedure

Metallographic specimens were polished and etched either manually or electrolytically. In the latter procedure the surface to be examined was hand-polished up to and including No. 400 grit abrasive paper. The electro-polishing solution consisted of two parts methyl alcohol and one part nitric acid. Fifteen volts for ten seconds gave the best surface with the maximum scratch removal and minimum pitting.

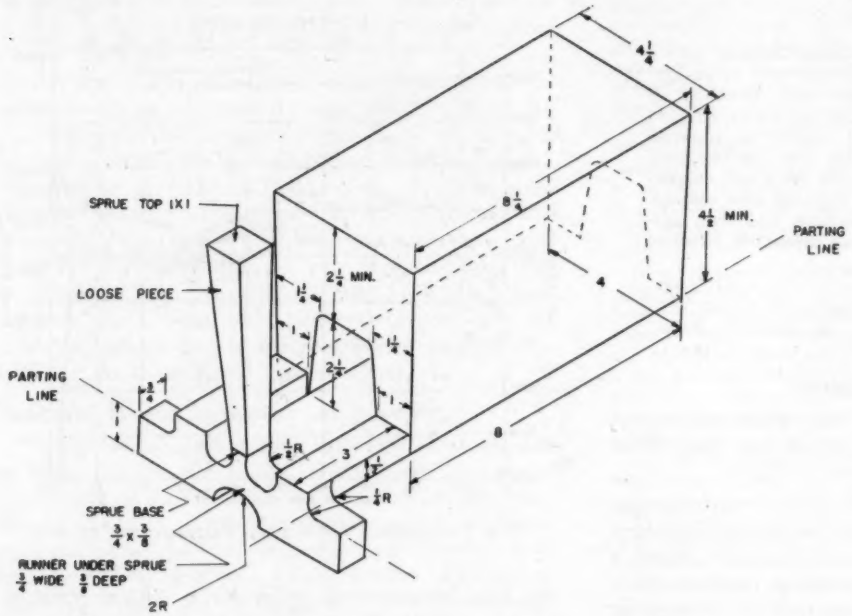


Fig. 2 — Double Keel block ingot.[13]

Results

Heat Treatment and Aging Curves

The alloy used in this investigation contained 5.6-5.8 per cent titanium. This alloy content is slightly below the maximum solid solubility of titanium in copper which is 6.0 per cent. Since 6.0 per cent is the maximum solid solubility, the maximum improvement of properties can be expected with the optimum heat treatment at this composition.

All the samples were solution treated at 1625 F for 16 hours and water quenched. After solution treatment the samples were aged at several temperatures and at various times in order to obtain the best aging cycle.

Aging temperatures chosen were 400, 800, 900, and 1,000 F. Times at the aging temperature varied from 1 to 72 hours. The aging results are shown graphically in Fig. 3.

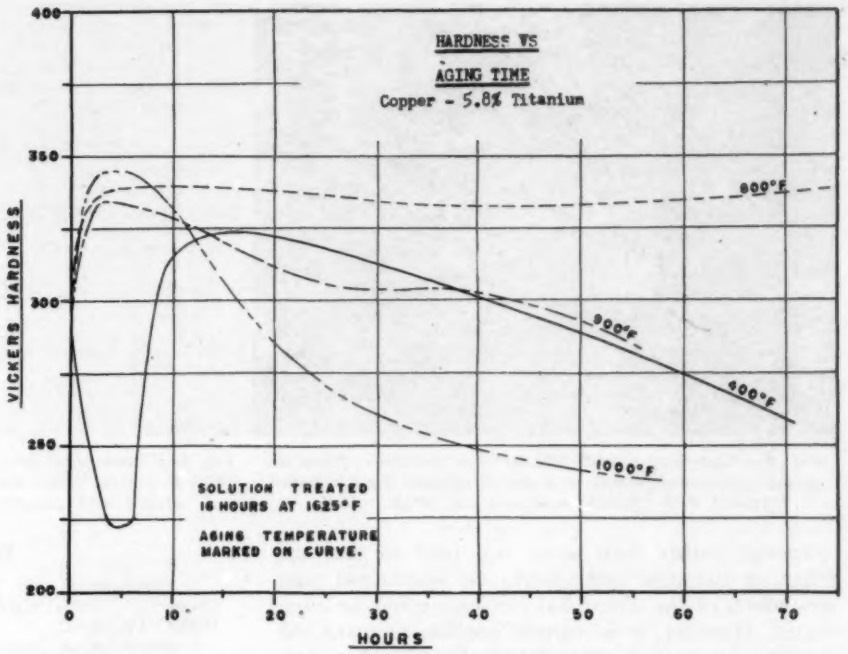


Fig. 3 — Aging curves.

The same maximum hardness (325-340 VHN) is reached at all temperatures, but this maximum is maintained even after being kept at 800 F for 72 hours before being tested at room temperature. At other temperatures, with the exception of 400 F, room temperature hardness drops off after aging times of 10 hours.

There is an initial drop in the curve for 400 F from 275 down to 215 VHN after which the curve rises to a maximum. The reason for this anomaly is not known.

Mechanical Properties

Tensile bars, which were solution treated at 1625 F for 16 hours and aged at 800 F, were tested. These results are tabulated in Table 1.

It is evident from the data that there is little damage in tensile properties with time at the 800 F aging temperature. The average tensile strength was 121,400 psi, the yield strength (0.2 per cent offset) was 107,300 psi, elongation was 8.0 per cent, and reduction of area was 18.0 per cent.

Microstructure

Figure 4 is a photomicrograph of a copper 5.7 per cent titanium cast structure. It is a typical cast structure showing coring.

Figure 5 is the same copper 5.7 per cent titanium alloy after solution treating at 1625 F. for 16 hours and water-quenching. There is only one phase present, the alpha solid solution, except for the small particles which might be impurities.

Discussion

A heat treated cast copper-6 per cent titanium alloy represents a copper alloy of high strength. Its tensile strength is exceeded only by cast beryllium copper. However, its strength does exceed the widely used manganese bronze. For the sake of comparison, the properties of all these alloys are listed below:

Table 1 — Mechanical Properties of a Cast Copper 5.8% Titanium Alloy

Sample No.	*Condition	Yield Strength, psi	Ultimate Tensile Strength, psi	Elong., %	R. of A. %	VHN
1	As-Cast	29,000	70,000	3.0	6 0	185
2	As-quenched	92,400	116,800	17.5	29.0	271
3		91,900	114,800	16.0	28.0	
4	Quenched, Aged	104,800	120,300	9.0	23.0	340
5	800 F, 5 hr	102,500	119,800	11.0	21.0	
6	Quenched, Aged	106,600	122,800	7.0	14.0	332
7	800 F, 41½ hr	113,300	126,300	7.0	21.0	
8	Quenched, Aged	108,500	116,800	3.0	9.0	331
9	800 F, 72 hr	108,300	122,800	9.0	19.0	

*With the exception of the as-cast samples, all the specimens were solution-treated at 1625 F for 16 hr and then quenched into water.

Properties	Beryllium Copper	Manganese Bronze	Copper Titanium
Tensile Strength, psi	155,000 (min)	115,000	121,000 (Average)
Yield Strength (0.2% offset), %	115,000 (min)	70,000	107,000 (Average)
Elongation (2-in. gage length), %	0	15	8 (Average)
Reduction in Area, %	0	15	18 (Average)

Beryllium copper and copper-titanium both depend on an age hardening heat treatment to attain their maximum properties. However, copper-titanium has one advantage over beryllium-copper in that its aging temperature is approximately 200 F higher. Increasing the aging temperature allows the use of a casting at higher operational temperatures without fear of overaging.

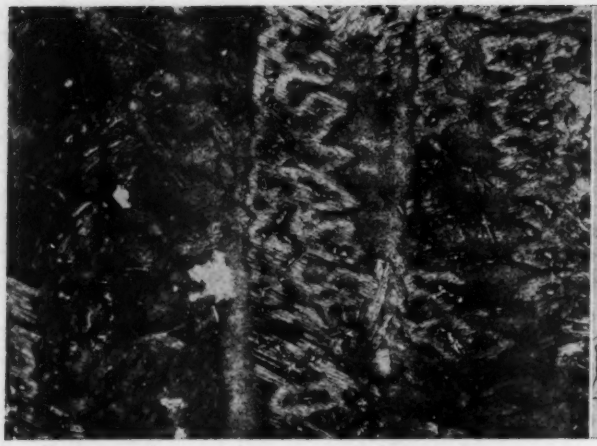


Fig. 4 — Cast copper with 5.7 per cent titanium. Note the coring appearance typical of a cast structure. Hand polished, etched with chromic acid solution. Mag. — 50X.

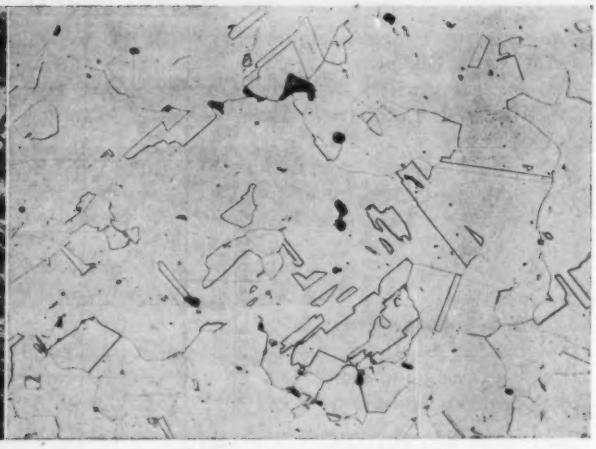


Fig. 5 — Same alloy as in Fig. 4 after solution treatment at 1625 F for 16 hours and water quenched. Hand polished, etched with chromic acid solution. Mag. — 50X.

Sponge, rather than scrap, was used so that the effect of titanium alone could be ascertained, and any effects of the additional elements could be eliminated. However, it is entirely possible to make the master alloy using scrap rather than sponge titanium. Secondly, it was felt that the titanium losses during melting could be minimized using a master alloy rather than contaminated scrap.

Based on data obtained from this investigation work has been started to determine the best methods of using titanium scrap as the alloying addition. A more complete investigation of the relationship between types of titanium additions and methods of making them and the amount of alloy lost is being studied.

All this work is being carried out on a binary of copper and 6 per cent titanium. A third investigation is under way to determine if the addition of a third element could improve the strength characteristics and castibility of the binary alloy.

Summary

1. A heat treated, cast copper-6 per cent titanium alloy has tensile properties exceeded only by those of beryllium copper. However, it does possess strength greater than the commercially used high strength manganese bronze.

2. The best heat treatment for the copper-6 per cent titanium alloy, in order to obtain the optimum tensile properties, is to solution treat at 1625 F and age at 800 F.

3. There was no noticeable change in properties in the test specimens with regard to time (up to 72 hours) at the aging temperature of 800 F.

4. The following average mechanical properties can be obtained from a heat treated, cast 5.7 per cent titanium copper binary alloy:

Tensile strength, psi	121,000
Yield strength (0.2% offset), psi	107,300
Elongation (2-in. gage length), per cent	8
Reduction in area, per cent	18

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DISCUSSION

Chairman: W. H. BAER, Bureau of Ships, Navy Dept., Washington, D. C.

Co-Chairman: G. H. BRADSHAW, Navy Dept., Philadelphia, Pa.

J. J. CONNELLY[1] *(Written Discussion):* The authors are to be congratulated for their progress to date and should be given every encouragement to continue study. Mentioned in this paper is the fact that they and other investigators have found alloys of copper and titanium difficult to cast. We have confirmed this fact at our laboratories and therefore suggest that, if this alloy system is to enjoy the popularity of other high-strength copper-base alloys, the authors concentrate their future studies first upon improving the foundry handling character of this material and then upon improving its properties.

Our interest in this alloy stemmed from a knowledge that the authors were realizing interesting properties with copper-

1. Metallurgist, Research Center American Brake Shoe Co., Mahwah, N. J.

titanium and also from a desire to find an alloy which would prove superior to the conventional beryllium-copper or chromium copper alloys now in use as resistance welding dies and other electrode applications. Relative to this particular application, our results were disappointing since, in addition to casting difficulties, we also found that the conductivity of copper-titanium is considerably lower than that obtained at similar hardness values with the conventional alloys. Our heats were melted in air using induction furnace equipment, poured at 2300 F, and no additions other than titanium were made that could effect conductivity. Hardness and conductivity data after heat treatment for the binary alloys which we produced are tabulated below together with the level of titanium found by analysis in the alloys.

% Titanium	Brinell Hardness 500 kg	Electrical Conductivity % I.A.C.S
0.33	38	25.0
0.48	42	19.0
0.64	54	20.5
1.40	154	19.0

With beryllium-copper (0.25-0.50% Be type), an electrical conductivity of 50% I.A.C.S. is obtained at 200 Brinell and with chromium-copper at 100 Brinell a conductivity of 70% I.A.C.S. is expected. We would appreciate any comments or suggestions from the authors as to how the electrical conductivity of copper-titanium alloys could be improved.

MR. EDELMAN *(Reply To Mr. Connelly):* The authors wish to express their appreciation for the comments and suggestions made by Mr. Connelly.

With regard to Mr. Connelly's suggestions, we would like to answer them point by point. First, it is the intention of the Pitman-Dunn Laboratories of the Frankford Arsenal to concentrate upon improving the foundry characteristics of the cast copper-5.8% titanium alloy. This research would be carried out to alleviate the dross forming tendencies of this particular alloy.

In addition, studies are underway to ascertain the best method of gating and risering.

Second, the prime purpose of this investigation was to develop a copper alloy with good mechanical properties, and no work has been done to improve the electrical conductivity of the alloy. Small additions of titanium cause a sharp decrease in electrical conductivity as shown by Mr. Connelly. To obtain good strength in the binary copper-titanium alloy, an addition of 5.8 per cent titanium is required. Therefore, it is the authors' opinion that this alloy will never develop the electrical conductivity of a comparable high strength beryllium-copper alloy.

CASTING FINISH — TOLERANCE — PRECISION

By

C. A. Sanders*

Casting Finish

Many castings are sold by appearance alone. If the surface finish does not appeal to the buyer, the physical properties of the castings certainly may not appeal to him.

Many machine tool and other castings that appear before the public eye on the selling floor, either attract the buyer or distract his attention toward a competitive product.

Those companies producing die castings, permanent mold castings or investment castings sell castings on appearance and cast almost to a machine finish.

The sand-mold foundryman must be aware of producing better casting finish, as it is an important phase of satisfying customers today. See Fig. 1.

Many items contribute to better casting finish and surface but the sand grain is the most important factor in a sand cast foundry. The finer the base sand selected, the better the texture of the casting. Finer base sand is the "key" to better and closer tolerances.

There is no question but that the aircraft designer has awakened the present steel foundry to an entirely new field. With the coming of larger aircraft, the metals that possess the most desired physical properties are the alloys of steel. The thin-walled castings demanded by the aircraft industry are increasing. The aircraft industry has become one of the largest purchasers of castings in the country. With the seeking of closer tolerances, sand foundries have become aware that they have not yet reached the end of their own ability to produce closer tolerances in conventional molding sand methods. Perhaps it may be necessary to redesign some castings for production operation, but the aircraft industry has recognized that certain alloys of steel perform the job better than other classes of metal.

From time to time a new process claims better finish, closer tolerances, lower costs, better molding conditions, etc. It is not the process that makes for better castings and better castings finish, but the control, supervision, method, and the base materials which are incorporated in the molding mixture.

Every foundryman knows, all too well, how quick the public is to accept a new process with claims unlimited simply because it is "new". If a process states that it, and it alone, is responsible for better casting finish, it is then time to study the raw materials which are required in the recommended mixture and certainly the fineness of the sand which is recommended for that process.

When making a comparison between a present process and a "new" process, always compare "like to like and equal to equal". The economics of the new vs. the old can be evaluated on this basis only.

Closer tolerances, surface finish, dimension size, are directly related to molding sand properties and methods. Foundry progress can be advanced considerably, making the most of what we already have, but utilizing a finer base sand to give this finish and not rely on imaginative claims of new processes.



Fig. 1 — Casting showing rough surface finish. (Courtesy of J. R. Nicholas).

*Vice-President, American Colloid Co., Chicago.

There is no substitute for know-how, supervision, closer control, and the urge to do better.

The foundry industry for years has been trying to improve on old methods, techniques of molding practice, and even the raw materials which are used. This is a healthy condition, as it produces and encourages an advancement for a basic industry. All interested foundrymen recall at least one occasion when a revolutionary idea was advocated to change our productive system. However, there are certain elementary, sound and sensible approaches which must be used and decisions that must be made upon "facts" instead of "opinions", or "the sales approach."

What Molding Method to Use?

There are many processes to select from in operating a foundry, such as green-sand molding, dry-sand molding, skin-dry molding, core molding, loam molding, investment molding, the lost-wax process, frozen mercury process, cement molding, die-casting, permanent molding of various forms, chamotte or ceramic molding, shell molding, and others.

In selecting a process for manufacturing steel castings one must consider the factors involved:

1. What are the initial costs of equipment and materials?
2. What is the cost to make the finished mold ready for pouring?
3. Exactly how much precision is required and demanded (for the casting which is to be manufactured)?
4. What controls have to be established?
5. Can a competitive process manufacture the casting for less cost?
6. Will a few thousandths more of metal actually be more difficult, or more costly to remove?
7. Are the castings to be welded?
8. Are the castings to be annealed?
9. Are the steel castings to be coined, or straightened?
10. How many castings are required per day/week/month or other unit of time?
11. If added to a new foundry, the facilities can be provided for a selected process. If added to an existing foundry the process selected could be governed to a great degree by space available for the proposed castings.

To capture attention rapidly, a process for sale may proclaim that it eliminates machining, or cleaning, or other related costs. When such statements are uttered, a favorable and ready audience reaction is obtained. The phrases "close tolerance" or "better precision" draw attention, particularly from management.

These two phrases are in themselves relative. "Better precision" and "closer tolerance" can be obtained with any of the presently used processes if more control and closer supervision is exercised. Designers, buyers, salesmen and particularly foundrymen should avoid these phrases unless accompanied by the name of the casting process used. Certain foundrymen have been placed on "the spot" by management for having been unsuccessful in achieving publicized expectations.

The past, and even recent publicity that parallels these statements is that which shell molding has received. If this process were as successful as publicity indicates, the majority of our 5,000 plus foundries would be using it in production. As it is, lack of enough public evidence that shell molding is a pronounced production success has discouraged many foundries from initiating it. An under-current of hesitance has developed.

The misconceptions of engineers and designers concerning the advanced properties of certain steel castings and casting processes in general must be cleared. Many of these claims are not always understood by designers, casting architects, etc., who rely chiefly upon literature. This literature must be from a source of production and from a reliable authority.

By giving more attention to control, supervision and equipment with the present time-tested methods, no doubt "closer tolerance" can be obtained, but will the market pay more for the extra costs beyond?

At present, sand-mold casting seems to be the only process which is practically unlimited in the size of parts produced.

Since variations in casting design and the molding method have considerable influence on dimensional variations in castings, it is difficult to set universal standards. Most satisfactory results are obtained when close cooperation exists between designer, patternmaker, and foundryman.

In sand-mold gray iron castings dimensional tolerances can be held under carefully controlled conditions—to the following (steel sand castings are slightly greater):

Dimension	As-Cast Tolerance, ± in./in. (gray iron)*
2 in.	0.010
4 in.	0.020
6 in.	0.030
8 in.	0.040
10 in.	0.050
12 in.	0.060

*AFS TRANSACTIONS, vol. 62, pp. 176-180 (1954).

In some instances, one or two dimensions of a casting may be critical but it frequently is possible to hold those within closer tolerances than normal.

Casting in plaster molds is favored chiefly by the non-ferrous industry to cast within 0.005 in. per in. on small castings and 0.015 in. per in. on larger castings such as rubber tire molds.

Tolerance

It is extremely important that accurate sources of information be consulted in determining tolerances generally achieved by certain foundries, as it is easy to state closer tolerances than are actually obtained. A factual relationship between certain assumptions and practical casting results depends greatly on the casting design, the individual foundry practice and most certainly the nature of the alloy being cast.

Heine[1] and Rosenthal[2] state that "in green sand molding, the dimensions ultimately obtained on the casting are the result of the faithfulness with which shrinkage, changes in dimensions of mold cavity, hardness of mold, stability of molding sand, mechanical

alignment of flask, temperature affixed, etc. are reproduced in the casting process". It is possible for sand castings to aim for the accuracy found in die castings and may be reached. This of course requires time, effort and equipment which affect the price of the casting.

In sand-mold castings, tolerances are as follows:*

Metal	Tolerance, ± in./in.
Aluminum Alloys	1/32
Cast Iron	3/64
Copper-base Alloys	3/32
Cast steels	1/16
Magnesium Alloys	1/32
Malleable Iron	1/32

*From R. W. Bolz, *"Production Processes — Their Influence on Design,"* a series of articles in *Machine Design* from January, 1949, to December, 1949: August, 1949, Sand Casting; September, 1949, Centrifugal Casting; October, 1949, Permanent Mold Casting; November, 1949, Die Casting; December, 1949, Plaster Mold Casting.

Above data are for average foundry sand-mold practice and with attention and effort, these dimensions can be held considerably below the tolerances as given, especially in small castings where other processes offer claims. Usually, one green sand mixture is adopted for an overall foundry practice. It is often found that this one-purpose sand mixture is for all classes of work produced. It is an unfair comparison when other processes show their efforts only toward small castings. The degree of accuracy with green-sand molding processes has never been fully exploited by the steel foundry industry.

Molding Process vs Surface Finish

The cast finish in various processes is as follows:*

	Finish, Micro inches, rms
Green Sand Casting	250 - 1000
Special Sand-Casting Process	250 - down
Die Casting	40 - 100
Precision Casting	30 - 100
Plaster Casting	30

*From R. W. Bolz.

It is incorrect to consider overall dimensional tolerance in the casting. Dimensional tolerance must be considered in sections of a particular casting. This differs from casting to casting design. Certain surfaces may of necessity be machined. Such surfaces certainly demand more attention and control in the dimensional tolerances, and accuracy is closer. The design of any casting influences the molding characteristics, which in turn influences the dimensional tolerances. It is so easy to make statements that a casting is within 0.001 of an inch. This may actually be only 0.001 in. per in. in tolerance on only certain sections of that particular casting, not the overall casting.

What Is 0.001 in. per in.?

An 1/8 in. is 0.125 in.; 1/16 in. is 0.0625 in.; 1/32 in. is 0.03125 in. and 1/64 in. is 0.0156 in. The size openings of certain sieves are as follows:

Sieve No.	Size of Openings In Microns	In Decimals of an inch	In Millimeters
30	590	0.0232	0.590
40	420	0.0164	0.420
100	149	0.0058	0.149
200	74	0.0029	0.074
325	44	0.0017	0.044
400	37	0.0015	0.037

A fine sand may be one with an average AFS grain fineness number of 100. Since the size openings of a number 100 sieve is 0.0058 in., suppose one grain of such sand held apart a shell mold? What then will be the tolerance? It is easy to forget that a No. 400 sieve has an opening larger than 0.001 in. How can tolerance be held with coarser sand?

Parting Line Influence

When parting lines are considered, very close tolerances cannot be obtained. Any parting line absorbs many fractions of inches per inch. A steel foundry is doing well to hold a parting line 0.015 in. per in. Additional measurement is added to the casting tolerance.

The function of designing castings links together the various casting processes to produce the desired dimensions. Certain dimensional characteristics result from the nature of the casting process selected. The question of dimensional accuracy is always involved in the design of any casting.

Precision

During a certain period, "precision castings" were those that were produced by permanent molds, or by dies. They were sometimes associated with the "lost-wax" process.

Casting high temperature steel alloys to accurate dimensions is most difficult. The metallic-mold processes are not likely to be suitable for steel and other alloys which must be poured at high temperatures. Accuracy of + − 0.003 in. per in. may be possible in such castings.

Such intricate casting processes permit castings of less than 10 lb in weight with the majority under 5 lb. Larger ones have been cast but not to extreme accuracy.

There is no question but that the word "precision" came into common usage in comparing castings that were die-cast. Since that time an evolution of claims has been made as to the accuracy of precision which develops from one process to another. This has been based upon a colorful promotion program rather than authentic engineering data.

As Brown[3] states, "Precision expresses a state of exactness, or accuracy or definiteness, or exact conformity with truth or with some standard." It is a relative word and it varies to that standard used to judge it. Some foundrymen prefer to say that tolerance of a particular casting is "a certain figure" and is cast to that "exact figure." This is the degree of precision. It is the ability to cast as exactly as the print states. Precision is not a figure close to, but "exact" to that allowable tolerance.

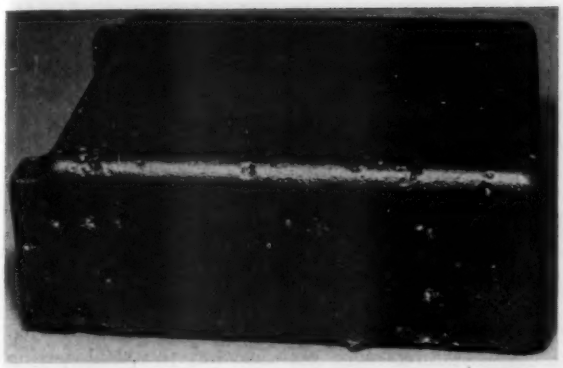


Fig. 2 — Casting with slag inclusion at surface.

Factors That Influence Finish and Tolerance

Men

Sand is only a part of the steel foundry, yet it is often accused of creating rough castings. Sand mixtures are accused so often that some foundrymen consider that production demands influence steel foundry practice. See Fig. 2.

With the gradual passing of the "skilled molder" comes the "mechanized unit operator". The "floor molder" and the "mechanized molder" must not be interchangeably used; one is an individual, the other is part of a team.

A "mechanized unit operator" relies entirely upon the sand conditioning units to prepare proper sand for his use. He relies upon the judgment and ability of the pattern and technical staff to place the gate and sprue in the correct location to warrant a good and smooth casting.

He further relies upon the maintenance crew to insure that the machine is in perfect working condition and that it performs correctly when it is set and gaged by the foreman. This machine often standardizes the number of rams, corrects the air squeeze pressure, corrects the lift, and controls other mechanical details that go into the performance of good mechanized molding practice.

The "mechanized unit molder" is an operator. The tools and the knowledge must certainly be furnished if surface quality is maintained. The complete practice is a formality and for that reason it is termed, "mechanized". It is a "teamwork" function.

Although sand is only part of this "team", too often it is the "goat" for any, and all, bad finish. Poor foundry practice results in a waste of raw materials and encourages trying excessive additions of raw materials to overcome poor finish.

Sand and sand mixtures cannot carry the complete burden as more "team errors" enter into rough surface as production increases. Wasteful habits rise during the periods of high demand.

The significant point is that the experienced oldtimer knew the sand composition, the ramming, the venting, the pouring, and all joined tightly together. If one varied, he compensated as the operation continued. His pride was his craft. Where detail was desired, he took time.

With present day mechanical devices, it is still important to consider finish, however, production makes continual changes impractical. Pride becomes production.

Pride in individual workmanship has gone with the molder; today's emphasis is on the mechanized operator's production record.

Gating

Castings are scrapped and repaired for many reasons, but the habit of thinking indicates that all defects are cataloged either sand, metal, or the night shift.

Figure 3 illustrates a defective malleable casting which was scrapped due to entrapped slag at the ingate. This is not a sand defect, but the result of gating or contaminated metal. Perhaps the direct answer would be to obtain cleaner metal, and no one can disagree with such a statement. No foundry has yet found the technique of completely removing foreign materials such as earthy inclusions, slag, or gas in the metal while maintaining a high production and good yield.



Fig. 3 — Malleable casting with slag inclusion at in-gate.

Best foundry practice observes proven gating methods. There is no subject quite so controversial as proper gating and risering. Gating systems are usually unique to a particular designed casting. Even good systems do not apply to other designs.

In any gating system, it is the yield and soundness of the steel casting that is important, but a defective casting that had a good yield and good soundness is still an added cost to the company.

It is generally accepted that clean metal must pass through a trapped gating system to remove the possibility of inclusions, such as slag, contamination, etc. Slag, sand and most contamination float to the top surface of the metal, but only if there is metal to float upon when in the gating system.

Some foundries prefer to use swirl or whirl gates (Fig. 4) which are strongly encouraged. It is generally accepted that the runner in the cope and the in-gates in the drag work best, but not generally practiced. Gating rough heads is encouraged. A negative pressure system is asking for difficulty.

Gating-Dirt

The steel foundry has accepted the word "dirt" to define any and all inclusions. This is incorrect. Unless the inclusion is properly identified, repair costs may continue.

Many inspection departments have little knowledge of what this "dirt" actually consists, but they realize that placing all inclusions under this heading relieves them of further identification.

Placing the blame on sand, sand mixtures, the molding department may be dangerous to the inspection department. It is still further dangerous to place the inclusion and blame for the scrap casting on the melting department. It is equally as dangerous to place the blame on the pouring department that has the responsibility of properly skimming the metal.

If the repair or maintenance department is accused of faulty ladle linings, etc., that too is dangerous. Management, in general, usually doesn't mind where the blame is deposited, as they are after the cure. "Dirt" then overcomes a multitude of sins.

The word "dirt" must disappear from foundry terminology. If the inclusion is non-metallic and shows ceramic material imbedded in the metal, determine what it is! If it is an oxide and dross inclusion, it must be identified. Most of these inclusions are at the surface and during cleaning they may be removed, but x-rays most certainly show those internally.

Surface inclusions may be the result of the molding sand cutting and washing. They may be the result of molds being soft rammed. They may occur from bonding clays possessing too low fired strength. They may occur from careless molding. They may occur from the cores that collapse too readily and may be due to sand rubbed or broken from the mold and core during setting of the cores or closing of the mold. If they are sand, charge them to sand.



Fig. 4 — Swirl gates promote clean castings. (Courtesy of V. W. Ziemer).

Slag inclusions take their toll but are often placed under the heading "dirt", which they are not! They may be confused with sand holes, or blows if the inclusion is completely removed by cleaning. Identify the inclusion!

Poor ladle practice and poor ladle materials combined with excessive use of water and poor drying can contribute to such a defect. These expensive ladle linings are not "dirt" and should be identified.

Much repair can be avoided by correct gating with a positive pressure gating system. Swirl gates (Fig. 4 are becoming more popular and the steel casting industry has an excellent design.

Slag inclusions may most certainly be broken into several classes. Many low-eutectic slags such as the steel's "snotter" differ entirely in chemical composition than a calcium slag and must not be called "dirt" but must be properly identified. Alloy additions can contribute to this defect and must carry their share of the blame. A gating practice that provides for a good trap may be necessary to overcome this contamination.

A poor gating system may also contribute to this defect if high metal pressure and turbulence is generated in the mold cavity.

Open risers encourage surface inclusions if they are not covered. Bumping the molds prior to pouring generates these defects. There are many, many reasons but this defect is not "dirt". Identify them.

It would be well to photograph in color all surface defects termed, "dirt" and show the inspection department the difference between them before and after cleaning. See Figs. 5 and 6.

What Is a Snotter?

This is one word that should vanish from steel terminology. To the author it is the same as the word, "dirt"; both must disappear. The steel snotter is common to steel castings and the description is solely associated with the steel industry.

The snotter is actually a slag inclusion appearing principally on the surface of steel castings but more frequently on large flat copes. With small castings this condition is more serious than with larger castings.

The snotter has a ceramic composition containing principally silica and alumina with many complex minerals attached. It is a low fusing eutectic with a low density. It is quite fluid and on heavy castings this slag is so fluid that it penetrates between the sand grains. For this reason it does not leave much impression in larger castings, but in lighter castings the slag replaces the metal of the casting. It varies in color from white to dark yellow, to brown and has a strong affinity to cling to silica molding sand. Its color depends on the elements present.

Oft-times this slag floating on top of the steel castings clings so tightly to the molding sand area that when these defects are analyzed much of the molding material is contained in the defect. It appears as a vitrified mass but is strongly different from normal

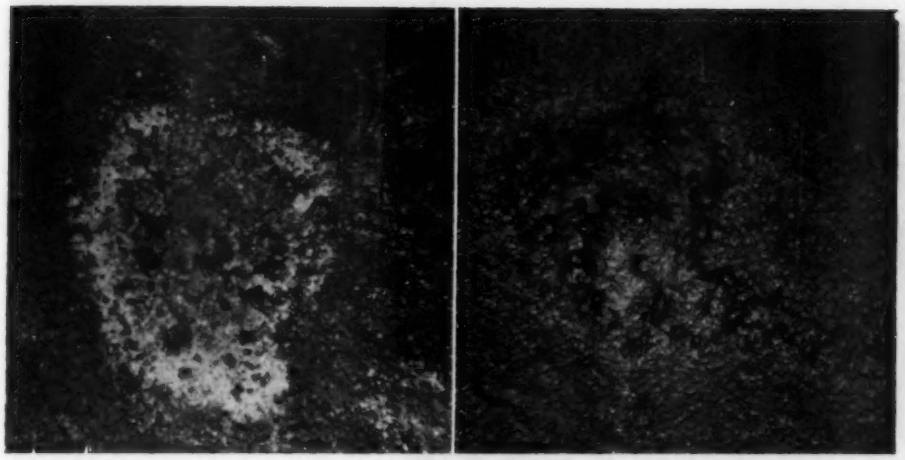


Fig. 5 (Left) — Photo of defect before cleaning and Fig. 6 (Right) — after blasting; hung over inspection bench it helps to identify and eliminate this defect.

furnace slag, as it appears to be thinner on the surface of the metal and rather difficult to remove by customary precautions.

In analyzing this defect by chemistry, if it has penetrated into the sand, it contains about 84 per cent silica content; 5 per cent alumina; FeO, 4-5 per cent; MnO, 4-6 per cent; CaO, less than 1 per cent.

If the steel containing snotters is cast in other than a sand mold, such as a graphite mold, the chemistry of the snotter may then be: silica content approximately 43 per cent; alumina, about 18 per cent; FeO, 2 per cent; MnO, 14 per cent; but the CaO advances to approximately 18 per cent. The chemical composition depends upon where and how the snotter is taken from the casting of the mold. It depends also upon whether or not the slag is fluid enough to penetrate between the sand grains.

Many foundries that have been more successful in controlling this defect use swirl or whirl gates. Figure 4 illustrates such a gate now being used by several steel foundries. There are other steel foundries that have incorporated a double gating system (Fig. 7) in which two runners are used and at the end of the last runner, two heads are placed on a tangent with the runners for feeding.

Skim gates have helped in certain cases but in others they have not been too satisfactory.

Where the gating system passes the runner through blind risers, fewer snotters are likely. A positive pressure gating system is preferred over a negative pressure gating system, thus preventing slag from being sucked into the mold cavity.

Where a siphoning action of the metal occurs, snotters are more prevalent. Pouring into a basin designed to eliminate swirling of the metal in the sprue is strongly encouraged. This swirling should be encouraged in the head or feeder but not in the sprue.

Many steel foundries have proven that snotters exist in the metal, as they used graphite and metal molds to produce castings containing snotters. They have associated this defect with certain deoxidizing procedures and have found that calcium-manganese-silicon, and even aluminum deoxidizers can contribute. Low eutectic slags erode linings also (Fig. 8).

Steel foundries always frown on decreasing yield by changing or increasing the gating system. However, yield may be decreased on some castings with costs



Fig. 7 — Double gating system used to prevent snotters on 150-lb stainless steel casting. (Courtesy B. W. Magalis, Duriron Co., Inc., Dayton, Ohio).



Fig. 8 — Snotters form in copes of graphite molds as well as in sand molds.

going down due to welding costs decreasing by better gating practices.

The argument will always continue as to whether this defect is caused in the molding sand or at the melting furnace. Certain metallurgists use various deoxidizers to obtain better ductility of the steel, others purposely reduce a furnace lining to save alloy costs.

Wherever slag is developed from either the use of alloys or certain furnace conditions, better gating practices are encouraged to overcome contaminated metal.

There is no steel foundry without such defects, but there are those steel foundries who have less snotters due to better melting, alloying, deoxidizing and with better ladle and gating practices (Figs. 9 and 10).

As yield (less gates) generally increases, the snotter defect advances. As carelessness develops in the deoxidizing process, snotter defects advance. It is a time-temperature function and regardless of where or when it occurs, it is bound to enter the mold cavity area unless a good gating system is developed. Foundrymen must always compare welding and chipping costs against yield (not riser) to determine which is the less expensive.

This defect must not hide behind the story that it is a seasonal defect, as it takes a heavy toll of surface defective castings and cleaning costs advance.

Metal Pressure

Head pressure and casting section thickness is directly related to surface finish. As head pressure is increased in certain advocated sand processes, the surface finish is gradually lost.

The pressure produced by any liquid metal in a mold cavity is transmitted in all directions and acts perpendicularly to the mold surfaces. A molding sand must be rammed properly to resist these pressures caused by the liquid metal against the mold. The molding sand mixture must have sufficient flowability to form a rigid and compact surface to resist this action.

The magnitude of the static pressure on any area of the mold surface is directly proportional to the head pressure of liquid metal above that area. The sprue height in inches of metal is one of the most important dimensions in proper gating practices. It is equally important to account for the additional pressure on the mold surface caused by the momentum of the metal as it fills the mold cavity. It is stated that the maximum pressure on the cope surface may be considered as equivalent to two times the static pressure. It is for this reason that it is necessary to have sufficient weight on the cope surface to resist the opposite force.

The total pressure exerted on the inside surface of

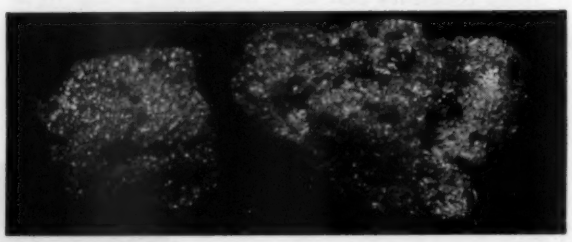


Fig. 9 — Pieces of large snotter which developed in the ladle prior to pouring. Composition is same as of those formed in graphite mold (Fig. 8).



Fig. 10 — Snotter formed in green sand mold.

the cope of a mold may be determined by multiplying the average pressure on the surface by the area of the vertical projection of the cope in contact with the metal.

Flask equipment should certainly be strong enough to overcome any lateral pressure developed in the mold. Therefore, casting finish may be directly related to the type of equipment used.

Pattern and Design

No casting produced is any closer to tolerance than the pattern. The casting is no truer than the pattern unless much "faking" of the pattern is practiced.

The pattern insures dimensional accuracy. Tolerance may then be held to a rigid specification, if the pattern is accurate.

Design is certainly the most important factor that influences finish or tolerance, but in production a patternmaker must explain to the engineers and designers what can be done in practice. The draft, allowance for distortion, and shrinkage allowance are important factors. All allowances for different dimensional areas multiply as the casting dimensions increase in section thickness or size.

A patternmaker works within 1/32 in. per in. on light patterns and slightly less than 3/32 in. on larger patterns. The casting can be no truer to the print than the pattern. While these are pattern production tolerances, aside from minor errors, special green sand patterns may be held to 0.012 in. per in. in accuracy to the print.

Fear of distortion is constantly with the patternmaker. Even pattern coatings for protection effect the dimensional accuracy. A sense of design to balance molding requirements with the purchaser's desire is ever present.

There is a sizable difference between wooden patterns and metal patterns in maintaining dimensional accuracy. Wooden patterns are subject to molding sand and storage conditions which are certainly not stable. Wood patterns vary in dimensional tolerances much greater than metal patterns and many of the larger steel castings use wooden patterns in molding.

Most sand foundries are reluctant to spend additional costs for the green-sand pattern whereas the new process will advocate a much higher priced pattern. Better workmanship of the higher priced pattern holds closer tolerances. Periodic checks on patterns to assure they have not worked loose is an important control point.

Often a larger draft is allowed in order to increase production at the molding machine, thus tolerance and dimensional accuracy is again sacrificed for production. It is unfair to compare casting processes if this major point is not considered.

Effect of Parting Compound

The type and selection of pattern parting compound can easily affect dimensional accuracy. A build up of parting compound greatly affects dimensional accuracy. This is often ignored. A regular dimensional check on flask pins and bushings and all patterns pins should be made.

Molding Equipment

Finish and tolerance cannot be obtained from equipment that has been in use for many years. Long periods of operation take the youth from these molding machines and flask equipment. An appraisal of possible replacement of antiquated machines and flasks is necessary from week to week. Progress can be maintained only if this viewpoint is taken.

A keen foresight into the advancements of molding machines and equipment available to do the job is today being applied by many progressive foundrymen. New processes are compelling green-sand foundrymen to recognize that some of the antiquated molding machines may still be in operation. The machine shop has advanced further in the parade of progress than the foundry.

Molding machines are maintained less than the poorest machine tool in the average machine shop. Molding machines should be purchased on their ability to reproduce molds with certain dimensional tolerances as well as "so many molds per hour."

Cleaning Equipment

Cleaning methods may easily destroy casting finish. Finish is further improved with a decrease in coring. Some cores which are difficult to remove from internal castings remain in the cleaning room as long as it is necessary to remove them. The outside surface may be completely roughened by excess cleaning which is not the fault of the molding sand practice. Again, the variety of designs in a jobbing steel foundry makes it necessary for a central blasting operation and small castings cannot be divorced from the medium and large castings being cleaned. Central shot blasting cleaning units strike an average not favoring the smaller castings.

Moisture

Moisture content has often been termed, "the necessary evil in the foundry." Although water is the least expensive lubricant employed for wetting molding sands, it can be the most expensive if used incorrectly.

Although a number of moisture indicating devices are used in foundry control work, many molders are unconcerned with machine testing methods and still persist in forcing their hand into a sand heap and by various ingenious ways predict the moisture content. This ancient method of moisture determination should not be completely frowned upon as it definitely measures workability by feel, yet it should be verified by the instruments. See Fig. 11.

The only drawback with this crude method is that the molder who claims satisfactory results cannot impart his knowledge to others, as there is no satisfactory rule, and this accuracy can only be acquired by practice. Knowledge from this sense of feel may hastily develop remedies for variations occurring on production lines, but machine verification of this crude test should be made before changes are advocated.

Hot metal cannot lie against water. Molders that use the water too freely may expect rough finish and defective castings. Even the use of a swab for coating

the mold surface with a mold wash can produce cuts, washes and sand inclusions as readily as they can prevent them.

It is too common to find foundries using water freely and to place this "weapon" into the hands of an inexperienced muller operator. He learns his errors of incorrect water application at the expense of the foundry.

High water content is accompanied by weak sand and gives false plasticity. Water in the mold is also one of the chief sources of oxygen and its presence alone can readily oxidize metal and promote pinhole porosity from absorption of hydrogen.

Perhaps the worst influence water has in molding sands is the bulking effect. Bulking effect increases with the surface of the sand. The maximum bulking effect usually occurs with 5 to 6 per cent water by weight. The maximum bulking varies with different sands from deposits; fine sand may be as great as 30 per cent; medium sized sand, 25 per cent; and coarse sand 20 per cent. The addition of 1 per cent to 2 per cent water by weight may increase the volume from 19 per cent to 20 per cent in some instances. This bulking effect creates a mold-metal surface influence and for that reason it is easy for castings to be oversize and overweight which may result in apparent shrinkage. Additional risers or heads may be added simply because the mold moves under metal heat and weight.

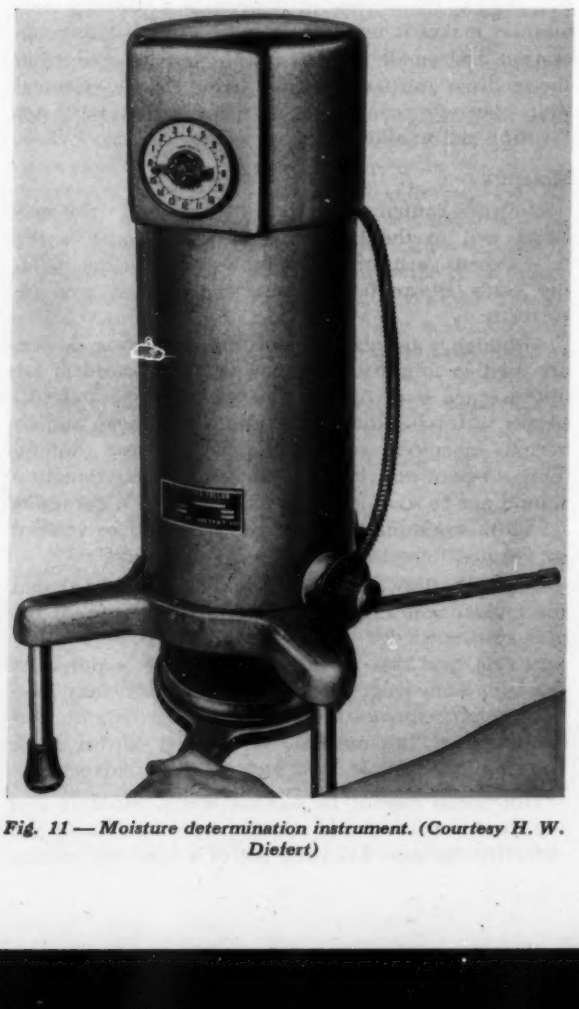


Fig. 11 — Moisture determination instrument. (Courtesy H. W. Diefert)

Since water affects every mechanical property of the molding sand with the exception of the fineness of grain, the control of moisture is of the utmost necessity.

Flowability

Flowability is that property of a sand mixture which enables the sand to fill pattern recesses and move in any direction against the pattern surfaces under pressure. The casting finish is no better than the mold and core surfaces over which the metal flows. If the mold and core surfaces are rough, the casting finish is equally rough.

During ramming, a sand with good flowability can carry a certain tightness of bonding by interlocking its sand grains. It is important that each sand grain be properly bonded and thoroughly anchored into place after ramming.

Although sand grain distribution and moisture have direct bearing on flowability, the greatest factor is the amount and type of bond employed.

A bond that is "too stiff" requires many lost labor minutes and even hours at the machine. If an operator must "force" the stiff sand around the pattern, molding costs are certainly going to increase. If a bond allows the sand to be "too flowable," ram-offs and other obvious defects are certain to occur.

An operator that has a good working sand can forget about controlling the physical properties and can concentrate on mold and core production.

Foundries are wisely becoming acquainted with the importance of decreasing the laborious job of hard filling and unnecessary ramming of the sand in the flask around the pattern.

A good workable sand is one that allows the less skilled molders to make good molds, therefore, more common labor may be employed.

A pattern is manufactured to insure closer dimensional accuracy. Accurate models of all kinds are produced to meet the specific tolerances allowed in making the mold or core. Do not molds and cores which are loosely and poorly rammed nullify the cost and care in producing the pattern?

If a sand has good flowability and properly fits itself around the pattern when rammed or jolted, then tolerances can be held to rigid specifications. A mold that is properly rammed also properly strips from the pattern with good, clean surfaces.

Deformation

Green deformation and hot deformation are of equal importance.

"Deformation" has been defined as a "change in a linear dimension of a sand mixture, which change accompanies a stress". It is the property of a green sand mixture to deform before rupture. It is measured and expressed in inches per inch (Fig 12). To the practical foundryman, the amount which the sand deforms at ultimate green compressive strength is deformation. In general, "deformation" is another way of reporting the ceramic term "plasticity."

If a sand mixture has high "plasticity" the foundryman generally states that the mixture is "fat",

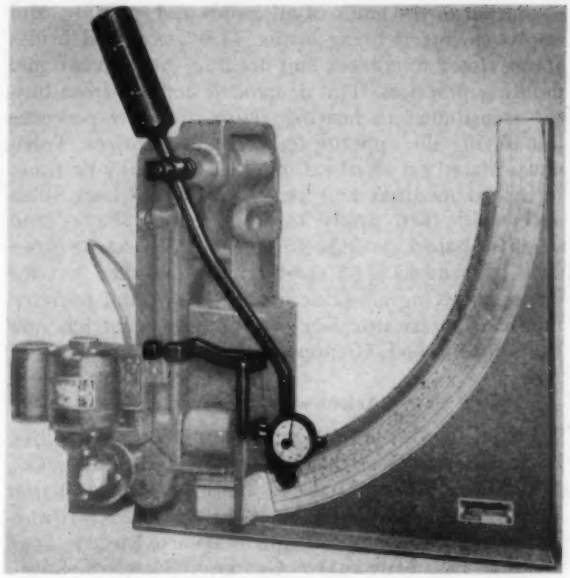


Fig. 12 — Deformation test instrument. (Courtesy H. W. Diefert)

"strong", "gummy", "sticky". etc. This is his verbal way of describing "high green deformation". High deformation is usually defined as being above 0.020 in. per in. Some steel sand mixtures may be above 0.040 in. per in. If the green deformation is exceptionally high, the sand mixture is usually difficult to ram and fill pattern recesses. It is difficult for a "high green deformation" sand mixture to move in any direction against the pattern surfaces under pressure. The foundryman may accuse this sand of "lacking flowability."

A high deformation sand mixture which resists flow and is "rubbery" may be one that encourages metal penetration or other surface conditions which increase cleaning costs. Vertical side walls may "burn-in", or "penetrate". As ferro-static pressure increases, metal penetration increases, if the mold-metal interface is soft and open.

A high green deformation sand may produce soft-rammed molds. When the metal is poured into it, the mold-metal interface at the vertical sidewalls of the soft mold may move. If so, shrink cavities may result from the casting being oversized and overweight. Generally, the sand is not condemned for this characteristic, as the metal is accused of causing the "unusual shrink". Usually more risers are added, or perhaps a chill may be located at this spot.

Many foundries have found that mixtures which produce high hot strengths to resist cutting and washing may promote these same defects if the sand is not rammed properly around the gate area. In such cases, green mold hardness should register the cause but the real cure lies in decreasing the green deformation to allow the sand more flowability around the gate area. This prevents the metal from cutting and washing as it passes through these areas.

One of the dimensional losses in green-sand castings is that too much draw is placed on the pattern to overcome sloppy molding. If a pattern has any deep pockets or deep draws, normally high deformation is encouraged. Even though tolerances are lost, due to production demands, the high green deformation is encouraged, as it prevents the sand mixture from packing too tightly in the pattern and allows for faster draws. Tolerance is intentionally sacrificed to increase these production demands. Bentonite and cereal are excellent to increase deformation in order to aid drawing but surface finish is generally impaired slightly. Straight outside surfaces are not impossible to draw and are practiced in some green-sand foundry work where it is required. Many patterns are drawn without a draft up to depths of 5 in. Although we do not like to advocate increased deformation, it is required in many casting designs.

Deformation increases rapidly with additional moisture content. Swells and rough finish occur more readily as the moisture content increases. Overweight castings and oversized castings may develop. Mold drops may occur as deformation decreases. Crushes may develop with lower deformation.

The percentage of clay in a mixture varies deformation greatly. The moisture content combined with the variation of clay also affects deformation. In general, deformation is least when just sufficient clay is present to bond the sand grains with a low moisture content. As the clay and moisture are both decreased, deformation decreases rapidly.

Mulling has an important part in determining deformation. As mixing effectiveness is improved, deformation increases. As mulling time increases, deformation usually increases.

Mold hardness plays an important part in deformation. As an example, an 80-mold hardness sand carries approximately 150 per cent more metal in the mold without serious deformation than a 40-mold hardness sand mixture.

Group Control

It is most difficult to check any single green, dry or hot sand property to insure against surface defects.

It is the belief of many steel foundrymen that to control green properties, it is necessary to control a group, or unit, such as mold hardness, green compressive strength, permeability and density.

Mold hardness is a "key" to other green sand properties in the mold. It has been neglected at molding stations, as data surrounding it generally come from the laboratory. Permeability is a meaningless word unless the mold hardness is stated. A soft-rammed mold is usually higher in permeability than a hard-rammed mold with the same base sand.

By carefully studying the mold hardness and density, under a given number of rams at the machine, flow characteristics of the sand mixture can be observed. This aids in determining the number of jolts at the machine and the amount of butting and tucking necessary to eliminate the soft spots. It prevents the mold-metal interface from penetration or burning in.

To rely on a single unit of measurement is not correct. Hot properties are perhaps more important

than green properties but the emphasis on control is to overcome surface defects.

Mulling

Proper preparation of compounded molding sand mixtures is fully as important as using good materials; faulty preparation is a frequent cause of casting surface defects even though the best quality ingredients are used.

Methods of mixing depend upon the facilities of the foundry, the type of sand to be bonded and the amount required. The best mechanical mixing produces well over 100 per cent more strength than inadequate mixing; consequently, more bentonite must be used with less efficient methods.

The purpose of mulling sand mixtures is to develop strength. By adopting the incorrect mulling time, or shortening the mulling time, results in many surface casting defects, including many inclusions.

A high speed muller should have a minimum of 1½ minutes mulling time. A slower heavier muller should have a minimum of 5 minutes mulling time. Friable and weak molding sands can cause cuts, washes, inclusions. and other surface defects such as scabs, buckles, etc.

Hot or wet sands in the steel foundry are a continual problem. Many have overcome this problem by using facing sands which are more initially expensive but cheaper in the final overall casting cost. There is no substitute for sand storage to overcome hot sands. Stickers and drops always occur with hot sands. Wet sands in most cases is a sign of neglect.

Re-Use of Reclaimed Sand

Many steel foundries have claimed certain surface defects develop when a preponderance of reclaimed sand is used on particular castings. This does not condemn reclaimed sands, but it points that unless a good job of reclamation is employed, burn-on or burn-in surfaces may occur on many castings.

The unlimited opportunities for the progressive foundryman prevail daily. The ones who become firmly entrenched in the orthodoxy of old-fashioned practices may be unwilling to add progressive and newly discovered green sand molding improvements. Consequently, they pay for their lack of initiative in scrap losses or heavy cleaning and welding costs. Those who lull into a sense of false security soon find the progressive green-sand foundrymen obtaining business from current casting buyers by employing up-to-date practices.

Conclusions

Castings are still ready for the resurrection of improved castings finish, precision and tolerance. There is no question but that the present foundry industry is a fertile field for immense advancement. The consolidation of the gains of all molds and molding sand improvements is being made. This forces and brings about closer tolerances and accuracy by conventional molding practices. The demand is coming from buyers accustomed to hearing claims for new processes but having the appetite for green sand prices. Tolerances placed on small castings will gradually be transferred to medium and larger castings. Closer tolerances will then apply to larger sizes, shapes and weights of steel castings. Limitations of present green sand have never been thoroughly established but the buyer is forcing closer accuracy. Green sand foundrymen can go far in closer control than what has now been approached. Greater tonnages are open to them.

Acknowledgment

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DISCUSSION

Chairman: D. F. SAWTELLE, Malleable Iron Fittings Co., Branford, Conn.

Co-Chairman: B. H. BOOTH, Carpenter Bros., Inc., Milwaukee, Wis.

Secretary: L. E. WILE. Lynchburg Foundry Co., Lynchburg, Va.

NILES KITCHEN[1]: Is the use of strainer cores beneficial in reducing defects resulting from poor gating practices?

MR. SANDERS: In some instances they may help; however, in most cases other solutions, involving new techniques in gating, are better.

GEORGE JAMIESON[2]: In the use of finer sands is it possible to decrease or eliminate venting?

MR. SANDERS: Men with tradition and art vented freely and discriminately. Production shops tend to undervent but they try to build permeability into their sand mixtures through grain distribution, material selection, etc. Venting is an aid against expansion difficulties as it creates voids into which the sand grains can move as well as relieving the gas pressure. Use of "pop offs" is a beneficial venting tool.

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2. Foundry Supt., Canadian Car & Foundry Co., Ltd., Montreal, Que., Canada.

YOU CAN REDUCE NOISE IN YOUR FOUNDRY

By

E. R. Lund* and W. O. Hanson**

The purpose of this paper is to show that the average foundryman can reduce the noise level in his shop effectively without a completely technical approach. Whenever you must escort a man to another section of the shop to carry on a conversation, you know a noise problem exists. We had this situation in our foundry cleaning room.

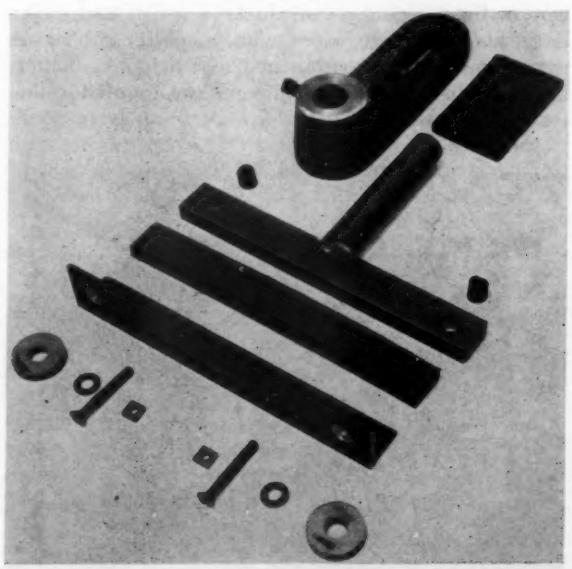
A few years ago we enlarged our foundry. Under the old shop setup, our cleaning room was at one end of the foundry with no separating walls. We had noise but it was not too serious. With the new building program, we placed our new cleaning department in a separate room in the new section of the foundry. Our equipment was well arranged into a compact unit. We soon discovered that we did have a serious noise problem in this boxed-in area and decided to make a study of the noise level.

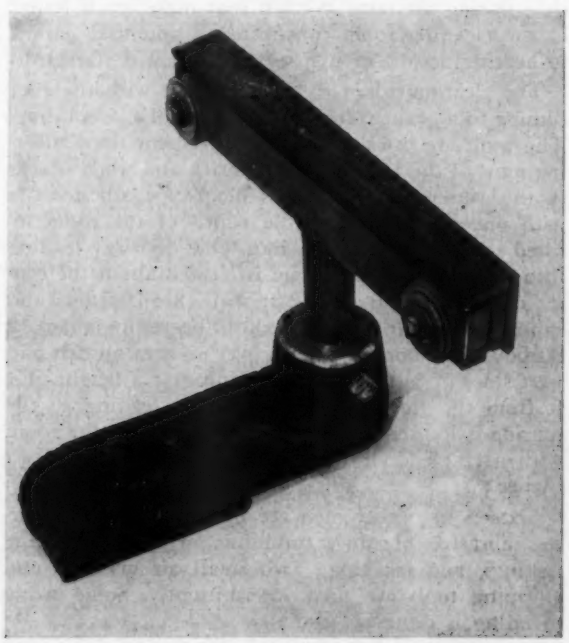
For a year and a half, beginning in 1953, a study of the noise level in the foundry cleaning room of our La Crosse Works had been conducted with reduction of the general noise level in mind. Studies made in 1952 indicated noise levels up to 108 decibels when our cleaning room was operating not nearly at capacity. No analysis was made at that time to determine the frequencies of the noise.

Another study in March 1953 indicated levels up to 104 decibels with no readings falling below 94 decibels. At this time the foundry was again relatively slow productionwise. As was the case in previous checks, no frequency analysis was made of the noise.

Tests conducted during late 1953 and early 1954 repeatedly confirmed exposures to noise levels well above the 100 decibel mark. Realizing that varied sources of noises that develop in a foundry cleaning

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In isolating the footing of tumbling barrels the author used grommets and washers on the nuts and bolts as well as ⅛-in. thick pads to insure complete isolation of noise.



Hoppers and tote boxes were coated with rubber-base material to help reduce the noise level.

room make it impossible to duplicate exactly test conditions each time a check is made, we think it significant that exposures ranging from a minimum of 94 decibels to a high of 115 decibels were obtained by the technician who sampled the room whenever a check was made.

Tests showed noise levels of 93 to 110 decibels in Nov. 1953. An analysis of the noise made with an octave band analyzer indicated noise levels in the 600 to 2400 cps range were constantly up to 96 decibels in the cleaning room. Four cleaning room operators were exposed to noises up to 102 decibels in the high frequencies. These exposures were present for about 5 to 8 hr of the working day. Overall noise levels during these tests ranged from 94 to 106 decibels with intermittent noises of 112 decibels noted from time to time during the tests. We felt that noise levels found in the cleaning room constituted a potential hazard to hearing according to presently accepted standards.

The cleaning room is 40 ft by 70 ft with the roof coming to a peak 30 ft over the center of the building. The walls are brick and tile with cement floor covering most of the area with some brick and wood blocks in the center. The ceiling is block-type asbestos tile with one skylight about the width of the room located in the center of the roof. One doorway leading into the main foundry room is located about the center of the room and is 12 ft wide. Steel girders and supports form a network overhead beginning about 12 ft above the floor. Windows cover an area on east and west sides of 8 ft by 40 ft beginning at a height of 4 ft from the floor. One double door leading to the outside is located at the southeast corner of the room with a single door to the outside on the northeast corner.

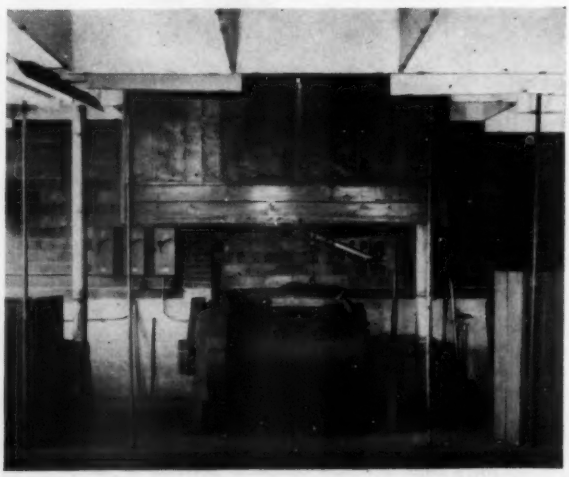
Sixteen to twenty men are usually engaged in airless abrasive blasting, tumbling, grinding, cleaning castings, and trucking. Two small air grinders and chipping tools are used intermittently. Some swing grinding is done occasionally.

There are four tumbling barrels of which two are in use usually for the first 3 hr in the morning. Occasionally three barrels are running simultaneously and sometimes four, but mostly only two. One airless abrasive blasting unit is used steadily all morning and intermittently during the afternoon. Occasionally we use two airless abrasive blasting units simultaneously. Grinding by five to eight men is a steady morning's job. Exposure to noise in high levels begins at 7 a.m. and usually lasts until afternoon, or an exposure of about 6 hr a day for most cleaning room employees. Three to five employees work in the cleaning room during pouring in the afternoon and are subject to noise levels of about 95 decibels for the full 8-hr period.

In the room are four double grinders using 20-in. diameter wheels. Also, one swing grinder is used intermittently and a second airless abrasive blasting unit is used as a standby unit. Throwing of castings into trays and blasting unit hoppers is a regular source of noise.

Studies of noise sources and noise transmission paths were then made over a period of several months. During this time all noise exposures in which men worked in the foundry cleaning room were thoroughly investigated. Identification of noise sources and their effects on the overall loudness were tabulated and necessary abatement steps became clear as the studies progressed.

First source of noise were the four tumbling barrels. Studies indicated noise was being transmitted through a re-enforced cement apron on which the barrels stood and it came out at an airless abrasive blasting unit located across the room on the same apron. Isolation of these barrels by footing changes was indicated. Several types of isolators, mounts, and shock absorbing materials were investigated. Best results were obtained with a commercially manufactured material composed of layers of tightly twisted, closely woven, lightweight, cotton duck of which each layer is thoroughly impregnated with a special rubber or neoprene compound vulcanized together. This material is the most effective dampener we found for this type of isolation.



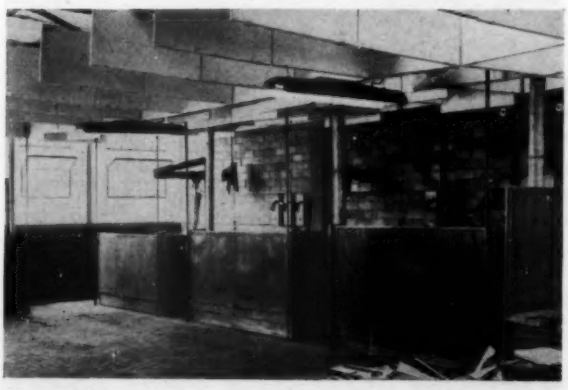
Outside of tumbling barrels treated with rubber-base material reduced noise noticeably.

In isolating the footing of the tumbling barrels we used grommets and washers on the nuts and bolts as well as 5/8-in. thick pads to insure complete isolation. There was no way to estimate the possible effects the isolation of the barrels might have on the overall loudness. We were surprised to achieve a significant reduction in the general noise level with the elimination of these transmission problems.

Another serious noise level was created when employees tossed castings into hoppers of two airless abrasive blasting units. While this noise was intermittent, present only when hoppers were being loaded, the frequency and loudness was sufficient to cause a serious problem. Levels exceeding 112 decibels, when castings were hitting the bottom of the hopper and seldom falling below 100 decibels when the hopper was filling, were constantly recorded. The high frequency and intensity of this noise caused employees to complain of this operation more than any other in the cleaning room. Not having a recording device on which to tape and play back this noise for more accurate analysis we realize there probably can be some question as to the absolute accuracy of our frequency band analysis of this noise source. However, for our purposes we are able to compare results obtained by our abatement measures. From the employees' point of view a serious complaint no longer exists. Reductions in loudness across the frequency band averaged slightly more than 19 decibels with best results in the higher frequency bands.

Cost of the dampener installed in the hoppers was low, amounting to about $150.00 each. Its simplicity of construction and effectiveness had been a constant source of favorable comment by those who have observed their use. After 6 months of hard use, maintenance is no problem.

Throwing castings into steel trays was another source of annoyance that also contributed to the overall high sound level of the cleaning room. Tests were made using a commercially available rubber-base sheet metal deadener applied to outer sides and bottoms of the trays to reduce noise from this source. First a 1/8-in. coating similar to that applied to automobiles was used; while it was an improvement, a 1/4-in. applied in two coats with 48 hr drying time between coats was found more effective. Considerable improvement by both reduction of loudness and frequency was effected in this step. Test trays to which the deadener was applied by an outside body shop would cost about $3.00 per tray. By purchasing our own spraying equipment and buying the material direct, the cost was cut to slightly over $1.00 per tray. Also, the bottoms of all trays into which castings are dropped from hoppers, are covered with 2 in. of sand. The sand covers the corrugations on the bottom of the tray about 1 in., giving a 2-in. layer of sand in one-half of the tray and a 1-in. layer in the remainder.



Sound barriers were installed near tumbling barrels to direct the transmission path of noise away from hearing zone of persons working in the area.



Overhead sound absorbing arrangement at grinding operations. So effective is this set-up that one can see the sparks caused by contact of the wheel and the work piece but cannot hear the results of contact.

Treatment of the outside of tumbling barrels with a 1/4-in. coating of the same rubber-base deadener as used on trays also resulted in a small but noticeable improvement. The hoppers of airless abrasive blasting units were also treated and finally all sheet metal guards were removed, sprayed with deadener and replaced using additional means of stiffening where vibration was indicated. All these steps, while small, actually reduced the overall level in the room about 2 decibels.

Redirecting Transmission Path of Noise

The next step was to direct the transmission path of noise from the tumbling barrels away from the hearing zone of persons working in the area. Costs were obtained from commercial sources for the installation of sound barriers in front or around the barrels. Prices quoted began at $5.00 up to $12.00 per square foot. None of the firms quoting could give assurance what would be accomplished by installing their materials. One firm recommended complete isolation of all barrels on a floating floor at a cost of about $4,500.00. This still would not take care of other difficulties existing in the room. We felt the cost was too high.

Using a plain wood barrier with wings tilted inward toward the barrel about 30 degrees, we found a substantial reduction in noise level was accomplished, more noticeably in front of the machines. Checks throughout the room indicated the air-transmitted path of the noise had been directed upward and some energy was dissipated in the ceiling area. The noise reduction in lesser degrees was reflected throughout the room. This installation was made using low cost No. 3 spruce boards ¾ in. thick, 8 in. wide and attaching them to guards that were already present in front of each tumbling barrel. We also found that spraying a ⅛-in. coating of rubber-base deadener on the inside of the barrier added to its effectiveness. There was an average reduction throughout the room of 3.7 decibels with 4 and 5 decibels reductions found directly in front of the barrels.

Following installation of the barrier in front of the tumbling barrels we still had a noise level ranging from 90 to 98 decibels overall in the room. The next step was to install absorbant materials overhead.

A search for absorbant materials uncovered 7 or 8 makes. Manufacturers provided a method of computing absorption needs, most of them based on Sabine's formula. Knowing the prevailing frequencies we were interested in reducing, we had no difficulty in correlating our problem with data furnished by these people. Tests were conducted on available types of acoustic absorbants to determine which best fitted our needs. Final choice was based on performance data furnished us by impartial sources, our own experience, manufacturers data and price. In calculating our needs, we decided an overall average reduction of 15 to 20 per cent in higher frequencies from our starting point would place the noise level in our cleaning room in a safe level. Reduction in noise level already obtained, indicated we should work toward an additional overall average reduction of about 8 decibels. Actually we were 1.1 decibels under our estimate, achieving a 6.9 decibel reduction. This was satisfactory.

In applying the absorption materials our hanging pattern was determined by the location and intensity of the noise sources. One instance worthy of note is the concentration of baffles over grinding wheels which took care of everyone but the operator. A few feet away from the operator, noise levels of 87 or 88 decibels were found while the operator's hearing zone produced readings of 90 or 91 decibels. To bring this noise level down, a special tool rest was designed. Success of this change was demonstrated in the reduction of 2 to 3 decibels in the overall and as much as 5 decibels in the upper frequencies. Now one experiences an unusual sensation when observing these grinding operations. One can see the sparks caused by the contact of the wheel and work piece but cannot hear the results of contact.

The above steps taken in our noise abatement program in the foundry cleaning room were accomplished in the following order:

1. Reduction of vibration transmission from tumbling barrels by installing isolation materials at footings of these machines.
2. Installation of wood mufflers in bottoms and backs of hoppers on airless abrasive blasting units. These mufflers were covered with sheet metal to insure longer life and to isolate them from the hoppers by use of isolation materials, bushings, and washers on the bolts by which same are attached to hoppers.
3. Treatment of all trays used in the cleaning room with sheet metal deadener, spraying all outside surfaces and bottom of all trays with a coating approximately ¼ in. thick.
4. Installation of sound barriers at the front of the four tumbling barrels. This will tend to absorb sound and change the directivity pattern of sound waves emanating from the tumbling barrels.
5. Installation of fiber glass sound stop baffles. These baffles were hung on wires about 8 ft above the floor in approved patterns.
6. Installation of tool rests and special arbor washers on grinders.

Some further steps are contemplated, i.e., isolating surfaces of benches, quieting ventilation system, etc., but these are only experimental and supplementary to an exposure in this room that is now considered free of hazards to hearing. We believe that noise can be abated at a reasonable cost as shown below from our experience.

1. Isolation of footings on 4 tumbling barrels—	$631.00
2. Installing noise mufflers in 2 airless abrasive blasting units—	300.00
3. Treating trays and barrels and hoppers with sheet metal deadener—	320.80
4. Installing 4 sound barriers—	134.00
5. Applying noise stop baffles to foundry cleaning room—	1,453.05
6. Changing tool rests and applying washers to grinding wheels	41.87
Total cost	$2,611.22

Engineering is important, but the fact remains that you can reduce noise in your foundry with what you have. In the not too distant future industry will include noise abatement studies and engineering as a part of building planning, along with lighting, heating, etc., when considering expansion or new building.

CORE MAKING WITH CO_2 PROCESS

By

Frank M. Scaggs*

ABSTRACT

Exact details of the history of the CO_2 process of making molds and cores are a matter of conjecture. Some sources indicate the English foundrymen had knowledge of the water glass principle before the turn of the century. Others say the French discovered the process only twenty years ago but were unable to produce a workable formula for making cores and molds. Still other sources indicate the carbon dioxide process is but a few years old.

Regardless of true details of a workable formula for the idea of making cores and molds utilizing carbon dioxide and water glass, this is the basic process: A liquid binder with sodium silicate as its base is mixed with the sand so that each grain is coated with the binder. When the carbon dioxide is applied to the sand, a chemical reaction occurs with the sodium silicate forming a firm gel, which cements the sand grains to each other. This much was known when articles by the eminent German foundryman, Dr. Schumacher, by Hans Heine, Roy Carver and others, were presented in foundry publications. Obviously there are other ingredients in the sodium silicate and in the sand, and other details have been omitted, but basically this is the CO_2 process.

It appeared that here was a new tool with which better castings could be produced at less cost. Our company wanted to find out if this is indeed true, and after almost twelve months experience and production, we feel that there are many advantages in the CO_2 process.

Every foundry is searching for methods and tools to raise product quality while lowering cost. In early 1955 published articles on the new carbon dioxide process began to stir imaginations at the author's company. In June active experimentation was begun in making cores with the water glass-carbon dioxide process. Specific results were hoped for in making cores, and larger, eventual goals were in mind for both cores and molds.

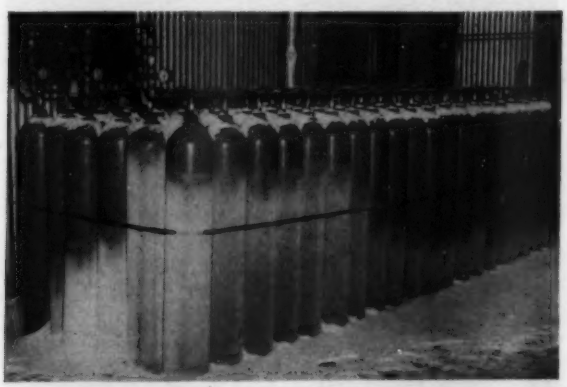
At the outset of experimentation the author's company was using approximately 70 per cent green sand cores and the remainder dry bond cores. Although green sand was economical and produced excellent casting quality, it was felt that handling and breakage costs were excessive. So that became a specific result hoped for in using the CO_2 process, i.e., to reduce handling and breakage costs. Another immediate goal was reduction in overall cost of producing cores by the reduction of man hours, reduction of oven equipment and by cutting indirect labor costs at the paste-up table. The cores made with dry bond material prior to our experiments were producing satisfactory castings, but it was felt that transportation and equipment costs could be cut substantially. So, more specific results were hoped for, including the important and desirable better dimensional tolerances.

First tests with CO_2 were on a very small scale, with one bottle of carbon dioxide gas and sodium silicate purchased from a local chemical firm. First results showed cores that looked very good and acted good in producing good castings with only one undesirable

*General Core Room Foreman, Oklahoma Steel Castings Co., Inc., Tulsa, Oklahoma.



Fig. 1 (Left) — Two-way blast nozzle cores made by CO_2 process. Fig. 2 (Below) — Bank of CO_2 gas tanks.



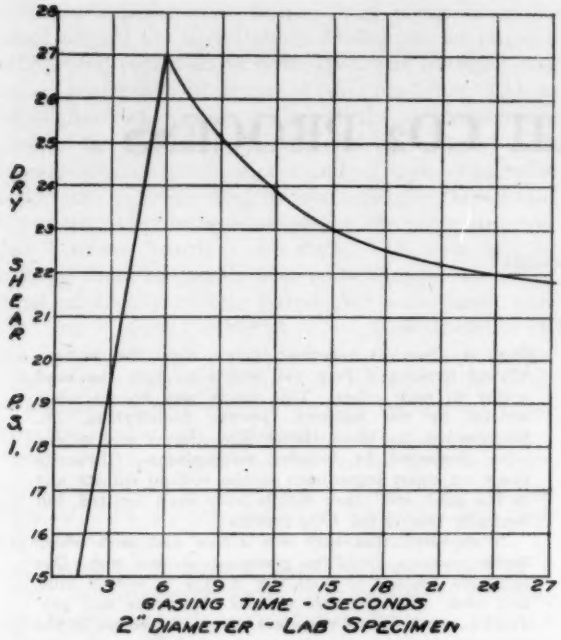


Fig. 3 — Curve shows relationship between gassing time and shear strength of core made by CO_2 process.

result: It was almost impossible to get the cores out of the castings. It was suggested that the cores could be sold for tombstone markers. To compensate for the low level of collapsibility and shakeout, organic and carbonaceous materials were added to the sodium silicate mix. This improved both conditions, but not to a consistent degree that allowed production.

A commercially prepared sodium silicate binder was tried and the desired collapsibility and shakeout were obtained.

CO_2 core production was then begun on three jobs in which driers could be eliminated. These were the two way blast nozzles. See Fig. 1. They have $\frac{5}{16}$-in. metal sections which must take pressure testing at 100 psi of air. As a result of better dimensional accuracy of the CO_2 cores, shop scrap on these jobs was reduced by 4 per cent. No hindered contraction cracks



Fig. 5 — Gassing a core from back side of box.

were encountered. The standards department, after direct time study, indicated an overall savings of 15 per cent by use of CO_2 cores on these three jobs. Other savings, such as elimination of driers, were not considered.

Satisfied that a workable sodium silicate formula had been proved with the nozzle castings, other jobs were tried with the new process. As each new job was attempted, new problems appeared. Foremost of these was hot deformation. Next came breakage due to low dry strength at room temperature.

The author is confident that the problem of deformation can be solved with addition of carbonaceous material or by pocket facing, or overall facing with zirconite sand in the correct sodium silicate mixture. It is the tentative conclusion that the problem of low dry strength was due to clay content in the sand in excess of one-half of 1 per cent.

At this writing the author's company is producing $8\frac{1}{2}$ tons of CO_2 cores per day, or approximately 60 per cent of production. Experience has shown that the foundry's wet sand reclamation system effectively removes sodium silicate residues from the sand, thus no problem has been experienced there. No difficul-

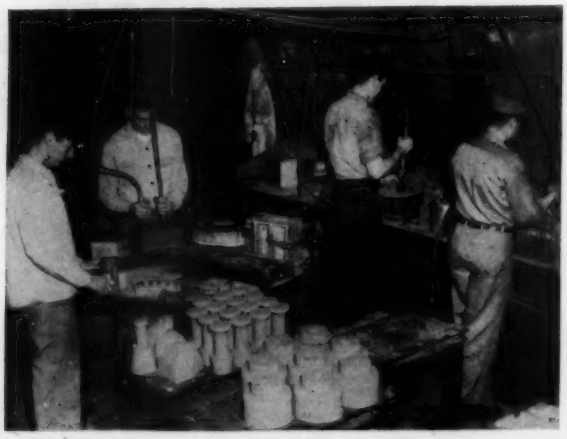


Fig. 4 — Special gassing devices in operation.



Fig. 6 — Gassing fixture for 5-spoke truck wheel core.

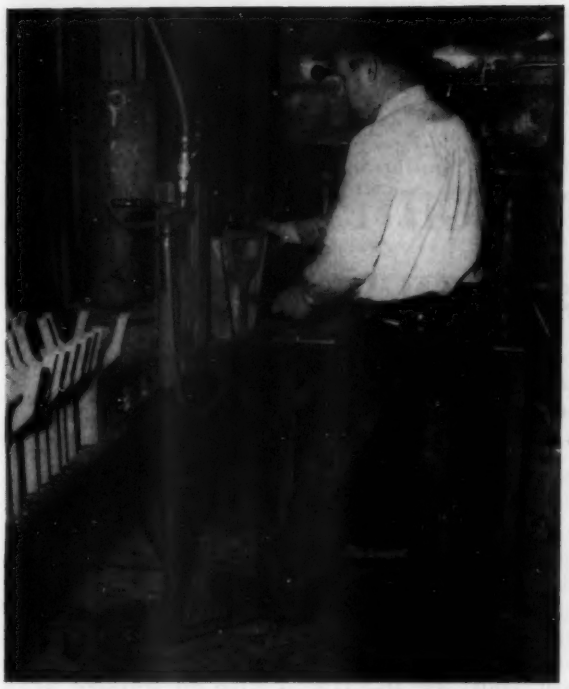


Fig. 7 — Gassing a 2-way blast nozzle corebox with core.



Fig. 8 — Sheave core being gassed.

ties have been encountered in blowing or ramming CO_2 cores. The ramability of the CO_2 mixtures, in fact, is superior to the traditional mixtures used in the past.

Special equipment for the new process consists of two banks of CO_2 bottles, one with 60 bottles and the other 120 bottles. See Fig. 2. The current Tulsa price for CO_2 gas is four cents per pound delivered.

The present system at the author's plant for "curing the cores" is $\frac{1}{8}$ lb of CO_2 gas per pound of sodium silicate sand. This amounts to a gassing cost of four mills per pound of sodium silicate sand. Those directly involved are confident that the cost can be reduced by accumulated information that will permit use of more accurate gassing time and pressure.

A graph was prepared by the sand laboratory technician indicating the desirable gassing time in terms of core hardness. See Fig. 3.

Other equipment for the CO_2 process include special gassing devices such as shown in Fig. 4 and an experimental turn-table which permits efficient gassing of cores by indirect labor. Figure 4 shows two men making bench cores, a third man is gassing cores on a turn-table, and a fourth man removing cores and reclamping the boxes.

Figure 5 shows the gassing of a core from the back side of a box, the rubber cups being adjustable on a four-way fixture.

Figure 6 shows the use of a gassing fixture used to gas a five-spoke truck wheel core. A drag half core is shown on the left, with the fixture and core box being used at the right.

Figure 7 shows a two-way blast nozzle corebox with core being gassed from bottom and two rubber cups a the top. This is a universal fixture to fit the various size two-way blast nozzles which we produce.

Figure 8 shows a sheave core being gassed. Three holes were placed in the outside diameter of the core box and a rubber tube was clamped over the holes and gas induced as shown. Screen vents were placed over the holes to keep the sand out. These are but a few of the methods available for gassing cores. We also use gas lances for deep cores.

The CO_2 core mixture most generally used by the author's company is as follows:

200 lb AFS 55 Fineness Silica Sand
100 lb AFS 100 Fineness Silica Sand
15 lb Collapsible Sodium Silicate
2 lb Cereal
2 lb Iron Oxide

This mixture is prepared at a cost of 73c per 100 lb. Our conventional green sand mixture costs 54c per 100 lb.

Some conclusions are in order after the completion of the first year of operation with the CO_2 process: That our company is enthusiastic about the possibilities as evidenced in the core production figures: Sixty per cent CO_2 cores. It is not the author's contention that the new process is more economical than the conventional—not at this time. But it is certain that modifications of the process and invention of automatic gassing devices and other equipment available soon, will affect the final cost figure of conventional versus CO_2 cores. The highest level of efficiency obtainable with the new process has certainly not been reached by this foundry, but it is felt that when that level is approached, water glass will pay off.

DEVELOPMENTS OF HIGH PRESSURE MOLDING WITH CO_2 PROCESS SANDS

By

T. E. Barlow*

In this work, we are concerned with the utilization of the CO_2 process for making production molds rather than cores. Precision-type castings with unusual finish was a principal objective but many non-precision applications soon became apparent. Above all, the object of this work was to increase the versatility of foundry operations and molding methods.

High-pressure molding includes both equipment and sand in the case of green-sand molding practice. The same combination applies to CO_2 process sands. The equipment aspect is possibly more interesting but we will discuss sand first because the characteristics of the sand explain why high-pressure molding is important in making molds with CO_2 process sand.

In green-sand molding, casting quality is a function of mold strength or resistance to mold-wall movement. This involves green strength (or more accurately, sand compressibility) and sand density. The desire is to produce a rigid mold-cavity and therefore, a precision casting. With CO_2 sand, the curing of the mold provides the mold rigidity. If so, why bother with high pressure? To answer this, we must look at the nature of the bond developed in CO_2 process sands.

Density of mold before gassing is the most important single factor in CO_2 process sand, and possibly the most often overlooked! The bonding action arises from the depositing of colloidal silica between the sand grains of the rammed core or mold. Therefore, only by having grain-to-grain contact can we hope to develop maximum strength per unit of binder. Maximum strength per unit of binder is much more than a matter of economics, in this case. It is extremely important to develop maximum strength in order to use minimum amounts of sodium silicate. Minimum sodium silicate is needed to ensure a minimum amount of soda ash after gassing with CO_2. Minimum soda ash is the only *fundamental* answer to shake-out and collapsibility! Therefore, maximum mold density is the surest and best control over collapsibility and economy in the CO_2 molds or cores.

To illustrate the effect of grain-to-grain contact, we have only to look at the effect of sand fineness. It is true that fine sands require more sodium silicate binder than do coarse sands "in order to develop optimum strength and hardness." But we should not lose sight of the fact that the "optimum" strength of a fine sand is much higher than the "optimum" strength of a coarse sand. Even with the same amount of binder, the fine sand is often the stronger. With the optimum amount of silicate binder, a fine sand mold is definitely stronger and harder than a coarser sand core or mold. Tables 1 and 2 show these effects of sand fineness.

TABLE 1 — EFFECT OF SAND FINENESS ON THE DRY COMPRESSIVE STRENGTH OF CORES MADE WITH 5% SODIUM SILICATE BINDER USING A CONSTANT GASSING TIME WITH CO_2

Sand		Strength, psi
Coarse	(40 GFN)	190
Medium	(70 GFN)	250
Fine	(85 GFN)	270
Very Fine	(120 GFN)	280

TABLE 2 — MAXIMUM STRENGTH POSSIBLE FOR DIFFERENT SANDS (INCLUDING AMOUNT OF BINDER USED TO DEVELOP MAXIMUM STRENGTH)

Sand Fineness		Maximum Strength, psi
Coarse	(3.5% silicate binder)	250
Medium	(4.0% silicate binder)	280
Fine	(4.5% silicate binder)	285
Very Fine	(6.0% silicate binder)	320

One possible reason that this important factor of density has not been stressed before lies in frequent confusion between sodium silicate as a binder and the sodium-silicate CO_2 process. The bonding action in the two cases is quite different. To illustrate this factor, it is necessary only to run a low-temperature bake on a sodium silicate core and a CO_2 core. Using the same amount of proprietary silicate-binder in both cases, the ungassed or straight silicate core

*Sales Manager, International Minerals & Chemical Corp., Chicago.

will be much stronger than the gassed or CO_2 treated core. In one case, the bond is chemical and in the second case, the binder is colloidal silica. The colloidal silica bond is much weaker. This is shown in Table 3.

TABLE 3 — TRANSVERSE STRENGTH AND SCRATCH HARDNESS OF SODIUM SILICATE CORES, WITH AND WITHOUT CO_2 TREATMENT, AFTER BAKING AT 230 F FOR 1 HR

Sand Used	Without CO_2 Gas		With CO_2	
	Strength, psi	Hardness	Strength, psi	Hardness
70 GFN with 2% silicate binder	37	52	20	57
70 GFN with 4% silicate binder	72	85	22	75
70 GFN with 5% silicate binder	68	89	55	87
70 GFN, 5% binder, 1% high pressure additive	68	88	34	90

CO_2-sodium silicate sands have a much better collapse than do sodium-silicate bonded cores. Even so, the collapse problem is frequently the major problem. The poor collapse is a result of the soda-ash by-product action upon the ceramic bond and fusion of the core or mold. The hotter the metal, the greater is the fusion or ceramic bond effect. Hence the shake-out problem is markedly different from that of an oil or resin core. The usual approach is to increase the refractoriness of the core by adding carbon. This is practical in spite of the adverse effect of such additions on the cured properties of the core or mold. Such secondary control should be used only after every precaution has been taken to ensure minimum amounts of soda-ash. In using high-pressure molding additives for this purpose, we add carbon for this purpose in addition to controlling sand flowability. Of the two, the sand flowability effect is the most important because it helps obtain high density, with minimum silicate and therefore, minimum soda-ash by-product.

To make proper use of the process as it applies to high-pressure molds, we must evaluate the usual limitations to see if they apply. For example, it has been stated that most additions such as clay and cellulose detract from the strength and hardness of CO_2 process sands. Closer examination of the data shows that the conclusions are based on holding the core or mold for a period of time. It is true that clay will cause a low strength after seven days storage but in practice, we are pouring the molds only minutes or hours after making. There is no reason for holding a production mold for days before using. In this case we find that from 0.25 per cent to 0.50 per cent of western or southern bentonite usually *increases* the cured strength and hardness. As much as 2.0 per cent to 3.0 per cent of fine clay type bond is frequently desirable in the same manner. These effects are shown in Table 4.

In the same way, we find that treated cellulose may add strength and hardness by increasing the flowability and controlling moisture. In fact, in this process, we find that moisture is a most important factor. Unlike most core work, however, we do not strive for zero moisture. We want sufficient clay, moisture and additives to add compressibility to the sand. We want some green strength in order to develop mold hardness even before gassing. Again, grain-to-grain contact is the important factor. At all times, we keep in mind that only the "immediate" properties are of interest and we want to use a minimum of silicate to develop the finest, smoothest possible mold.

Equipment and Rigging

There are two methods of producing density in a mold. One is sand flowability as described above. The other is through the use of high-pressure squeeze. The high-pressure work has used the diaphragm principle for two reasons. In the first place, we wanted to use facings when desired. Diaphragm squeeze makes this simple because it does not move the facing from its original position on the pattern. There is no tendency for the facing sand to slide or be jarred from the position of importance in the final casting. Secondly, we used diaphragm in order to produce a thin shell of sand rather than a full mold. The shell has the advantage of permitting faster gassing because of its thickness in addition to its obvious economy by using less sand. In its present form the process can be used in three different forms, (1) contoured shell, (2) total facing or, (3) spot facing. The essential equipment is the same for all three.

To use the principle efficiently in the foundry, we first must replace the ordinary pattern stool with a universal gas-stool. This is either a hollow box with holes to let the gas out or a labyrinth or maze design which will connect with any pattern plate used. The pattern plate is vented with standard core-box vents. The underside of the pattern plate may be routed out to connect the vents with the gas chamber. The CO_2 is passed up through the sand from the plate. This provides maximum ease and speed of gassing because gas flow is not restricted after it hits the sand. The process is much simpler than trying to gas the surface of the mold and expecting the CO_2 to find its way out through the vents. Furthermore, this technique adapts itself to production because the CO_2 valve may be mounted on the machine. After squeezing, the molder hits the CO_2 valve before he strips the mold. In a few seconds, he starts the pin lift and picks off a completely cured mold.

The three methods are simple in concept. The choice depends on the application and foundry.

TABLE 4 — EFFECT OF CLAY ON CO_2 PROCESS CORES AS USED FOR HIGH-PRESSURE MOLDING AND AFTER SEVEN DAYS STORAGE

Clay Added	Compressive Strength as Normally Used, psi	Compressive Strength if Held 7 Days, psi
No Clay	250	500
0.25% bentonite (western)	280	220
0.50% bentonite	240	180
1% fireclay bond	290	380
2% fireclay bond	300	400
3% fireclay bond	280	200

In the countour-shell method, the entire mold is made of CO_2 sand. Tight flasks have been used thus far because we were after precision in this process. The flasks are made as shallow as possible. After the high-pressure diaphragm-squeeze, a shell ranging from ½ in. to 1½ in. in thickness is left to be gassed. After gassing the cured mold is stripped, clamped, and poured. This process applies to the foundry specializing in such molds and particularly one capable of scrubbing and reclaiming the sand from the shake out. No green sand or backing is used. Normal flasks and equipment are used throughout.

In many cases, the foundry will wish to use a minimum of CO_2 sand or will wish to incorporate the process on selected castings in their regular green-sand molding line. In this case, they riddle a thin layer of CO_2 facing over the pattern and then fill the mold with green sand. The gassing is very fast because of the thin shell. After gassing, the mold is stripped, clamped, and poured. At the shake-out, the cured shell has either disintegrated (at the pattern face) or goes across the shake-out as a core-butt. In either case, the soda ash pick-up is at a minimum. The CO_2 shell and the green-sand backing are actually knit together so that the shell does not fall out before pouring. After all the two sands are squeezed together *before* gassing. They must cling to each other.

We believe one of the most interesting possibilities is the use of CO_2 sand for spot facing to eliminate a ram-up or set core. There are a large number of castings which are essentially green-sand jobs, but which could be improved if cured areas or cores could be set at low cost. For example, a cylinder head might be improved by a cured CO_2 facing on the combustion chambers. Again, an air-cooled cylinder could be made in green sand except for the fins. Or we may wish to stop an erosion problem near the ingate. The possibilities seem endless.

The practice is simple. The pattern is mounted on the usual gas stool with a vent or vents located only at the spots to be faced. A handful of sand is placed where desired and the rest of the sand is regular backing or system sand. The diaphragm molding method prevents the "spot" from shifting and at the same time provides the squeeze pressure needed for density. After pouring, the "spot" comes across the shake-out as a core-butt. Soda ash infiltration is not a factor.

Except when "spot" facing is used solely to stop an erosion or scabbing condition, we have worked with rather fine sands for maximum finish. The results were exactly as predicted. We found it possible to produce any part (or the entirety) of a casting with "super" finish and unusual casting tolerances. We retained all of the desirable effects of the CO_2-cured mold from the standpoint of dimensional stability, scabbing resistance, a speed of curing. Because we were working with much finer sands than normally used for a blown core, we developed peel and finish equal to the finest obtainable in any other casting technique, including high-pressure green-sand molding with fine sands.

Although we were using a shell of cured sand, we were able to handle the mold in a perfectly normal fashion. Because the CO_2 process facing and the normal green sand were actually molded together before introducing the CO_2 gas, there is no tendency for the two materials to separate or part. This cannot be compared with backing up a cured shell with green sand. It is more like a routine facing operation except that the facing in this case has a high scratch hardness together with high compressive strength.

The shell facing should be kept thin for minimum gassing time. Incidentally, passing of CO_2 through the green-sand portion of the mold has no effect on the properties of the mold. Only the special CO_2 facing-sand hardens during gassing.

Up to now, we have worked with gray iron, aluminum and bronze. Other metals will be studied in the future program.

Summary

To summarize our present position, we should emphasize that this paper is a progress report and not a final summary. What has been done has commercial possibility but further improvements in the sand practice should not be too far away as we will delve more deeply into the various additives such as clay, water, cellulose and high-pressure proprietary products. It is already obvious that "super-castings" are practical by combining the known effects of high-pressure molding on density with the fast cure properties modified CO_2 process sands. The process shows tremendous versatility because the effect can be confined to a specified part of a casting or to the entire casting. To the user of high-pressure molding equipment, it offers the possibility of specialized castings on the same molding line and equipment as now used for green sand. As such, we believe it warrants continued investigation.

DISCUSSION

Chairman: J. O. OCHSNER, Crouse-Hinds Co., Syracuse, N. Y.
Co-Chairman: E. C. ZIRZOW, Werner G. Smith, Inc., Cleveland.
Secretary: D. J. JONES, New Jersey Silica Sand Co., Millville.

M. H. HORTON[1]: Are you certain that some of the properties in the cope are not coming from the baking of the cope by radiant heat during the pour? Could this be related to the ratio of alkalinity to acidity and, thereby, effect collapsibility? Does this addition of sodium silicate harm the sand system?

MR. BARLOW: Actually, we are curing the mold with CO_2 before we strip it. If you could draw the pattern without curing, you would have no reason to use the gas. The baked core or mold is stronger than the gassed core or mold.

There is no way to determine just what is going to happen to the sand system since no foundry has run enough of the material, letting it get back into their system sand, to give this answer.

W. E. GRUVER, JR.[2]: I question Table 3 in that the cores were also baked after gassing and, therefore, the strengths shown are a result of unhardened silicate left after gassing and not silicate gel as mentioned by the author.

MR. BARLOW: If we gas the core and then bake it, the strengths are always higher. In practice it is always safer to undergas than overgas. There is a difference between CO_2 silicate core and just a silicate core. If we could always gas to optimum conditions, results might be higher in the gassed core.

1. Deere & Co., Moline, Ill.
2. Meehanite Metal Corp., New Rochelle, N. Y.

PRESSURE TIGHTNESS IN 85-5-5-5 BRONZE CASTINGS

AFS Research Progress Report

By

Paul K. Trojan and Richard A. Flinn*

ABSTRACT

The purpose of this investigation was to develop quantitative data relating the pressure tightness of 85-5-5-5 castings to mold characteristics and metal quality. Both of these groups of foundry variables are important since either centerline shrinkage or gas porosity can lead to failures in pressure tests. In the long range planning it is anticipated that a variety of mold characteristics such as mold materials and risering practice will be explored and that metal quality will be varied by using a number of different melting furnaces and practices.

For these initial stages of the work, however, it was suggested by the AFS Brass and Bronze Research Committee that the effort be concentrated on the development of a satisfactory test bar and pressure test method.

Data are given for pressure tightness at various distances from the riser in a 2-in. x 2-in. x 8-in. bar. The important features of this section are the charts showing a linear relation of the rate of leakage to increasing pressure. There is also a simple relation of the thickness of the section tested to the leak rate. A decreased leak rate is observed for specimens prepared with an abrasive cut-off wheel due to metal "smearing" while an increased or decreased rate of leakage can be developed for etched sections depending upon the etchants. The pressure test apparatus is of novel design permitting the use of specimens of various thicknesses, testing of localized regions ½ in. in diameter, as well as microscopic observation of the leaking areas.

Preliminary data are given showing the greater leakage for green versus dry sand castings, the vastly improved properties obtained with tapered chills, and the deleterious effect of gas resulting from poor melting practice.

Suggestions for future work include the further evaluation of the effects of mold material and design and of melting practice on test bar pressure tightness. Following this work the application of test bar data to the solution of carefully selected casting problems is recommended.

After a survey by the AFS Brass and Bronze Research Committee† of the problems of the small and large foundries producing brass and bronze castings, it was evident that the highest tonnage alloy was 85-5-5-5 bronze and furthermore that the principal foundry losses in the use of this material centered about the leakage of castings under pressure.

Guided by this information, it appeared that the most useful expenditure of research effort to benefit the small foundry as well as the large producer would be to develop quantitative information dealing with the effect of various molding and melting practices upon pressure tightness.

A review of the literature as well as many discussions of the subject revealed that although a number of castings are routinely tested for pressure tightness, relatively few test bar data were available. The principal test bar results were developed by the Naval Research Laboratory.[1,2,3]

To broaden the test procedures for this investigation it was decided to attempt to measure the leakage rate and to locate porous regions accurately, in place of using the pressure test as a go-no go gage. With this in mind the following procedures were developed.

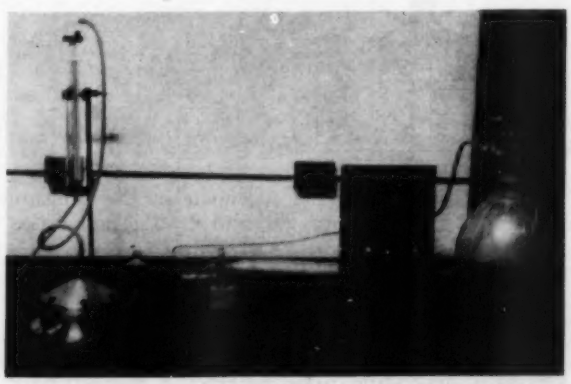


Fig. 1 — An overall view of U-M pressure test apparatus showing the source of nitrogen, pressure gage, surge tank, test cell, and gas burette.

*College of Engineering, University of Michigan, Ann Arbor, Mich.

†This is a Progress Report on an AFS-sponsored Research Project under the guidance of the AFS Brass and Bronze Division Research Committee. Personnel of this Committee are as follows: F. L. Riddell, *Chairman;* R. B. Fischer, *Vice-Chairman;* Mac Blackman; R. A. Colton; J. E. Fries, Jr.; G. J. LeBrasse; and E. F. Tibbetts.

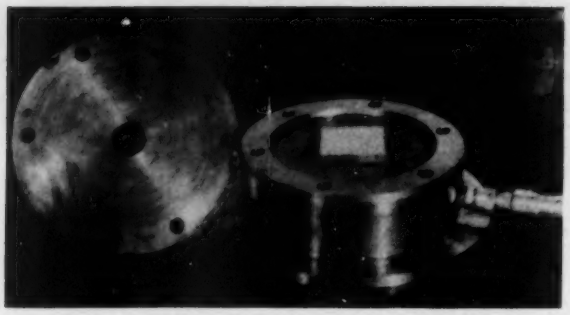


Fig. 2 — Disassembled view of the pressure test cell with a ¼-in. specimen and spacer in place. Note the "O" ring seated in the cell.

Procedure

U-M Pressure Test Equipment

An overall view of the equipment is shown in Fig. 1. Compressed nitrogen at 2200 psi from a supply tank at the right enters a surge tank and gage to deliver gas at the desired pressure level, from 0-1000 psi, to the test cell. When the desired pressure is reached in the surge tank, the supply tank is shut off from the system by the cut-off valve.

The test specimen is clamped between two "O" rings in the pressure test cell (Figs. 2,3) and the surge tank pressure applied to the underside of the specimen. The gas leaking through the specimen is collected from the pressure-tight head over the specimen is a gas burette (Fig. 1) at atmospheric pressure, and the leak rate (volume per unit time) is calculated.

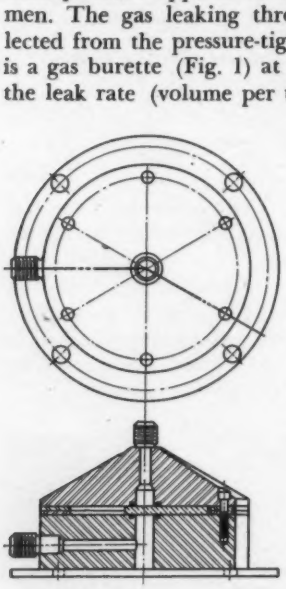


Fig. 3 — Detailed drawing of pressure test cell.

A close-up view of the specimen in the pressure test cell is given in Figs. 2,3, and 4. The apparatus will accommodate up to a 5-in. diameter test slice but permits examination of local areas of ½-in. in diameter. In this way the variation of leak rate across a section can be evaluated.

A choice of heads is available for placing over the specimen depending upon whether a measurement of leak rate or whether visual or miscroscopic examination is desired. The head for the latter use is shown in Fig. 4.

Specimens of almost any thickness can be tested by

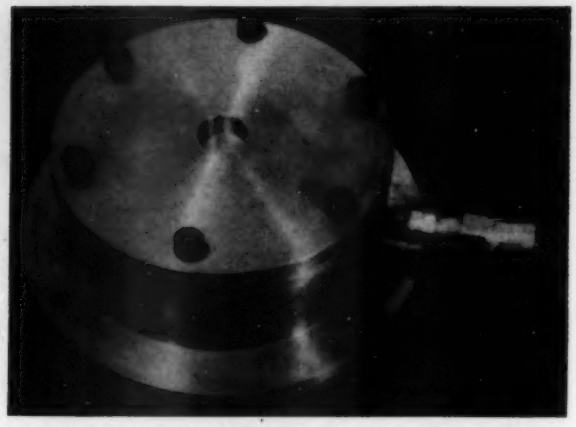


Fig. 4 — A different head assembly for the pressure test cell which can be used for a macro and micro examination of bubble formation using a soap or oil film over the specimen.

selecting the proper spacer ring to match the specimen. At present $\frac{1}{16}$-in., ⅛-in. and ¼-in. thick specimens have been tested.

Before using any cast specimens a piece of steel sheet was placed between the "O" rings *overnight* at 400 psi and no decrease in pressure was obtained. The leak rate of a "blank" therefore is less than the limit of error in reading the gas burette in ml/sec ± 2 ml). The time period for leak determinations is from 2-250 sec depending upon the leak rate of the particular casting. The time is determined with a stop watch graduated to $\frac{1}{10}$ sec.

Testing is begun at low pressures proceeding to high pressures to avoid opening leak passages. To the present time, however, no hysteresis effects of this type have been obtained.

Test Castings

The test casting (Figs. 5,6) is 2-in. x 2-in. x 8-in. risered with a 4-in. diameter, 6-in. high riser at one end. Patterns for longer castings are being prepared. Tapered chills, ⅝-in. thick and tapering from 2-in. at the far end to a point at the riser of the casting were machined and used as indicated in the data. Preliminary thermal gradient data have been ob-

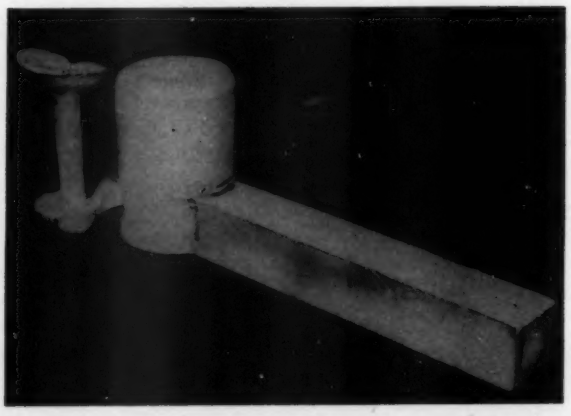


Fig. 5 — Test casting used with the riser and gating system still attached. Mold material was a green-sand mixture.

tained (but are not reported) using thermocouples at the positions indicated in Fig. 6. Test specimen slices 1/16-in., 1/8-in., and 1/4-in. thick were machined from the positions shown.

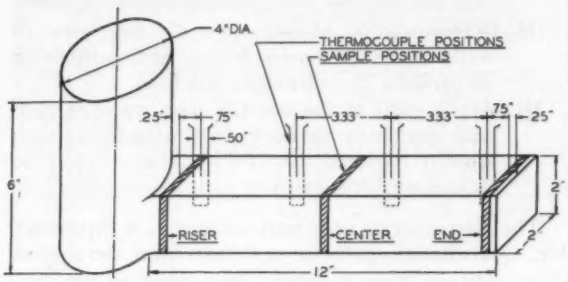


Fig. 6 — Drawing of test casting.

Melting Practice

To insure uniformity of charge in the initial experiments, one lot of 4000 lb of 85-5-5-5 ingot, all from the same ladle, was procured from the Michigan Smelting and Refining Co. through the courtesy of the Brass and Bronze Ingot Institute. All charges were 100 per cent ingot unless otherwise indicated.

Melts were made in either a 200-lb gas-fired furnace donated by Federal Mogul Co. or in a high frequency (3000-cycle) life coil induction furnace. The crucibles were supplied through the kindness of the Vesuvius Crucible Co. The details of the melting practice are given on the appropriate graphs in the data.

Data and Discussion

Test Methods

Before evaluating the effects of mold or metal variables on pressure tightness it was important to establish details of specimen selection and preparation and of pressure testing. These data in themselves are quite interesting and will be discussed under the headings:

1. Surface preparation of the specimen,
2. Specimen thickness,
3. Effect of pressure.

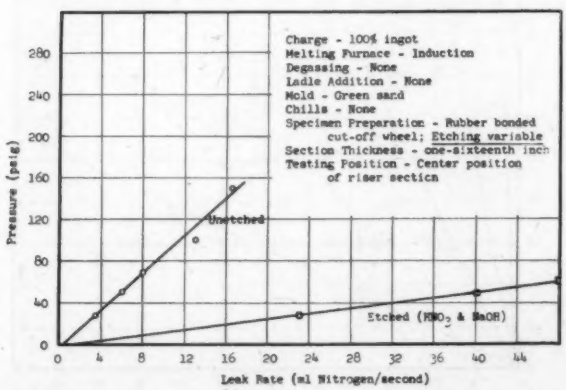


Fig. 7 — Effect of etching on surfaces produced by abrasive cut-off wheel.

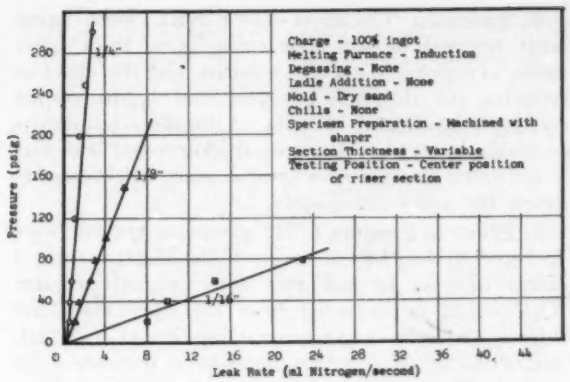


Fig. 8 — Effect of specimen thickness on leak rate.

1. Surface Preparation — It was feared that plastic flow or smearing of the metal during machining might affect results. Four types of surface preparation were employed; (1) abrasive cut-off wheel, (2) cut-off plus etching (HNO_3 and NaOH), (3) shaping, (4) shaping plus HNO_3 etching.

Figure 7 indicates that the combination etch (conc. HNO_3 followed by 20 per cent NaOH) removes the flowed layer from the cut-off wheel and therefore increases leakage.

Figure 9 illustrates the effect of leaving the products of etching in the specimen pores. This was an early experiment in which a shaped surface was etched with concentrated HNO_3 only. Apparently the solution products sealed the pores and gave a reduced leakage rate.

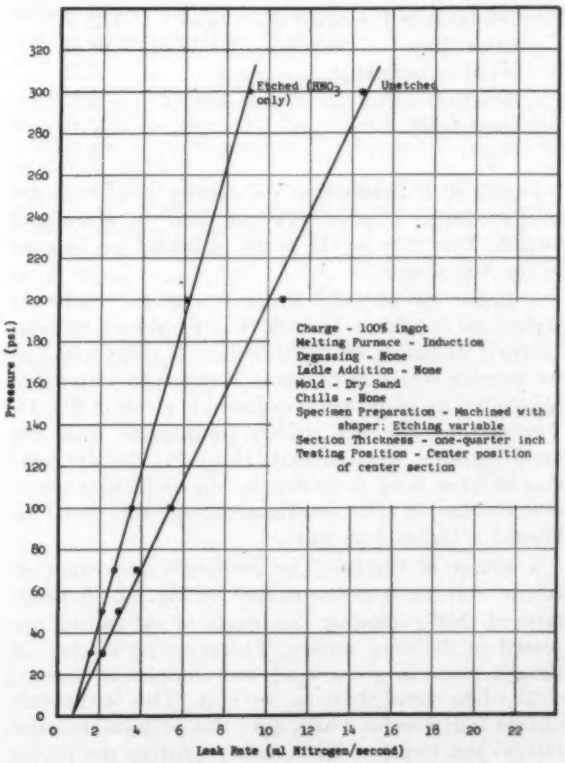


Fig. 9 — Sealing of shaper surfaces with etchant products.

2. **Specimen Thickness** – The NRL work previously referred to was done using 1/4-in. thick specimens. To provide a more sensitive test the effect of reducing the thickness to 1/8-in. and 1/16-in. respectively was evaluated (Fig. 8). A simple relationship, probably hyperbolic, between thickness and leak rate is indicated but requires several other thicknesses to define the curve adequately.

3. **Effect of Pressure** – The preceding graphs (Figs. 7, 8 and 9) and others to be discussed later show a linear increase in leak rate with applied pressure. This would be expected from the Ideal Gas Law. This is contrary to the general impression that leakage occurs when a certain breakdown pressure is exceeded. This apparent disagreement may result from the fact that when a liquid is used as the test fluid a certain base pressure is required to overcome capillary restraint–which is not encountered when using gas as the test fluid.

Mold and Metal Variables

While the principal effort has been directed toward establishing testing practice, some preliminary experiments to evaluate mold and metal variables have been conducted:

1. Leak rates in castings poured in green sand and dry sand molds,
2. Effect of melting practice,
3. Effect of chills.

1. **Effect of Mold Material** – The following mixture was used for both green- and dry-sand molds. The dry-sand molds were baked at 360 F for 8 hours.

140-mesh N.J. washed silica sand ..	172 lb
Silica flour	18 lb
Western bentonite	8 lb
Dextrine	1 lb
Corn flour	1/2 lb
Water	8 lb

Figure 10 indicates that the casting poured in dry sand exhibited a lower leak rate than the green-sand casting. This may be the result of higher gas content in the latter case.

2. **Effect of Melting Practice** – In the early experiments induction melting was employed without nitrogen degassing or deoxidation. A comparison of the inferior structure of metal of this type with crucible melted metal properly degassed is given in Fig. 11. The castings are not strictly comparable since the mold material was changed. However, the dry sand should have been favorable to the induction-heated material but in spite of this advantage this metal exhibited a higher leak rate.

3. **Effect of Chills** – The extremely important effect of chilling is demonstrated by Fig. 12. A single tapered chill extending the length of the casting was placed in the drag surface. The resulting casting, although made in green sand, was completely pressure tight when tested in 1/8-in. sections. This is the only casting tested so far which gave this performance and follows the trend of NRL data regarding the importance of chills in gun metal.

Conclusions and Suggestions for Future Work

In long range terms this program can be outlined in three principal stages:

I. Development of satisfactory test equipment and test bars for evaluating pressure tightness.
II. Determination of the controls necessary in mold materials, mold design, and in melting to produce pressure-tight test bars.
III. Application of the test bar data, such as thermal gradients necessary for soundness, melt quality required to avoid harmful porosity, to actual problem castings such as valves.

The first stage is now well advanced. A reproducible, quantitative pressure test has been developed. The addition of microscopic observation to the present technique should be of additional value in determining the types of porosity which are most harmful.

The second stage has been started and quantitative results from preliminary experiments have corroborated the NRL data for gun metal regarding the use of chills. The effects noted for green- and dry-sand molds are also in the generally accepted direction. Further evaluation of the effects of mold material will be of interest.

A quantitative evaluation of the effect of thermal gradient from the riser and chills by thermocouple readings during cooling is in process. The application of the effects of chilling upon reduced center-line feeding resistance should be of great significance.

Since no leakage was encountered in properly

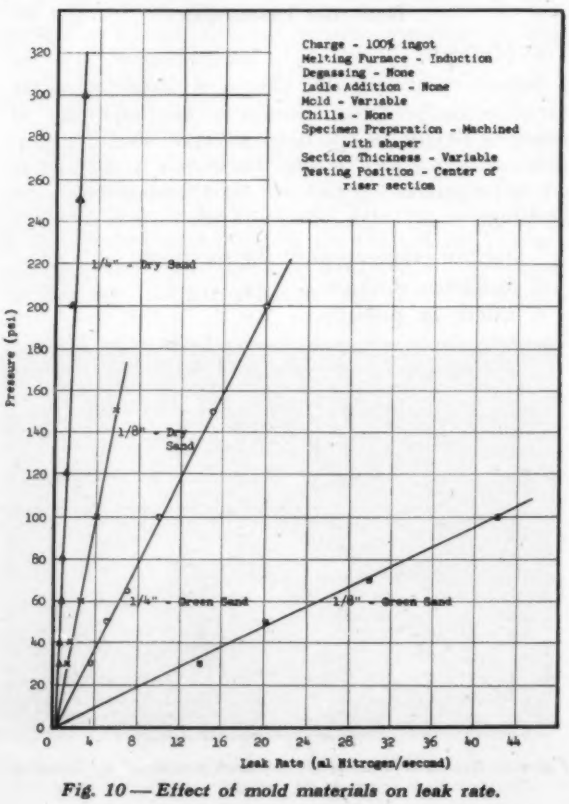
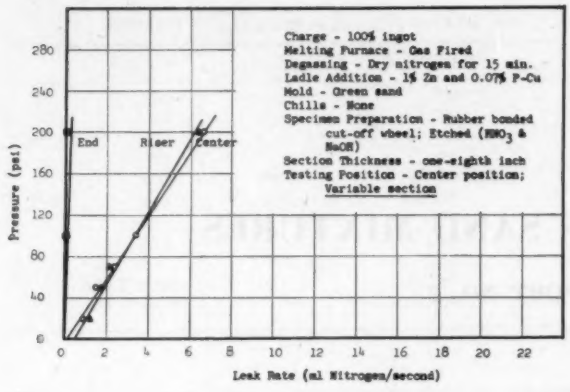


Fig. 10 — Effect of mold materials on leak rate.



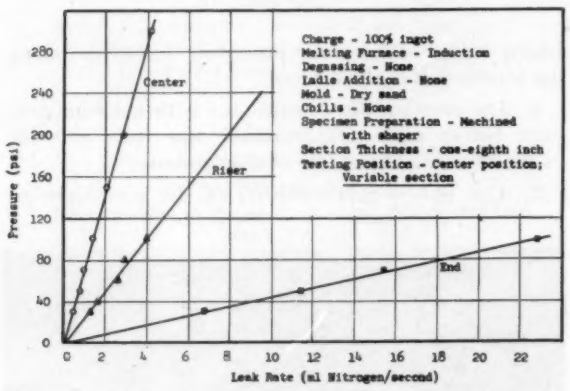
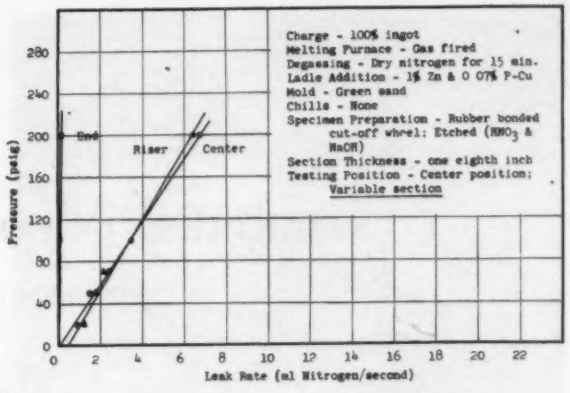


Fig. 11A (Top) and Fig. 11B (Below) — Effect of melting practice on casting soundness.



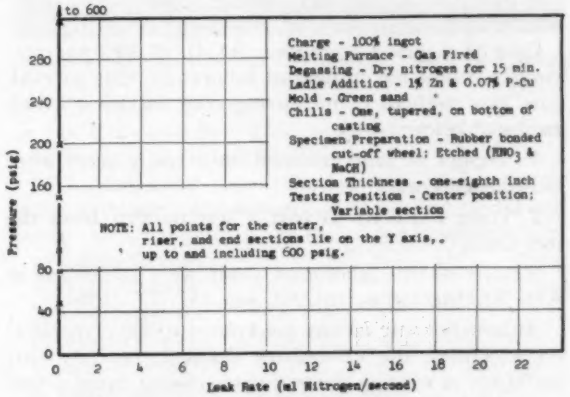


Fig. 12A (Top) and Fig. 12B (Below) — Effect of chills on leak rate.

chilled 2-in. square bars, exploration of heavier sections is indicated. The effect of various melting furnaces, type of charge, nitrogen degassing and vacuum degassing, deoxidation practice should also be evaluated.

The third stage, the application of test bar data to difficult castings is in the planning phase. It is generally agreed that a moderate sized valve be made in important quantities and offering leakage problems would be of the most general utility. The solidification pattern encountered with production rigging would be surveyed first, including thermocouple data and local pressure observations. Then the minimum mold and metal conditions for pressure tightness developed from test bar data will be applied and the effects upon soundness evaluated.

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STICKINESS IN CORE SAND MIXTURES

PROGRESS REPORT NO. 2

By

D. S. Mills*

Core Stickiness Committee (8-C-f) of AFS has previously conducted numerous laboratory tests to evaluate two methods of measuring core stickiness which are listed below:

1. Weight of sand retained on a test pattern after blowing the core.

2. Force required to part a test pattern from the core sand.

Results of this laboratory study were published in AFS TRANSACTIONS, vol. 62, pp. 131-137 (1954).

Although many factors are known to have an effect on stickiness, the Committee decided to study first the effect of varying the moisture content using actual cores blown in production equipment. At the same time the two methods of measuring stickiness listed above were evaluated to determine the correlation, if any, between stickiness occurring in actual production cores and the stickiness tests themselves.

Production equipment and laboratory facilities of the experimental foundry of the Producers Development Foundry of General Motors Corp. were made available to the Committee for this study. It is the purpose of this paper to report on the results of the work completed to date by the Committee on stickiness.

Test Procedure

Table 1 gives the sieve analysis of the core sand used and Fig. 1 illustrates the valve tappet core which was produced in a production core blower for this study. Table 2 lists the core sand mixtures used to produce the tappet core which were mixed in the muller. A dummy mix was run through the muller initially to help reduce any variables from one mix to another.

The mixed sand was delivered immediately to a hopper above a production coreblower (Fig. 2). Listed below is the exact procedure followed during the production of these cores:

1. The corebox was conditioned with kerosene only once before each sand mixture was used and the excess kerosene blown free of the corebox.

2. The blown characteristics of the core were re-



Fig. 1 — Valve tappet cores produced in production coreblower from Mixtures 1, 2 and 3.

*Chairman of Core Stickiness Committee (8-C-f) and Project Engineer, Foundry Process Development Section, General Motors Corp., Detroit. Other members of this Committee are J. R. Young, Cadillac Motor Car Div., GMC, Detroit; A. L. Bryant, General Motors Corp., Detroit; W. H. Buell, Aristo Corp., Detroit; A. L. Graham, Harry W. Dietert Co., Detroit; and T. V. Linabury, Ford Motor Co., Dearborn, Mich.

TABLE 1 — LAKE SAND SIEVE ANALYSIS

Sieve No.	% Retained
30	.568
40	4.570
50	23.136
70	47.120
100	20.780
140	2.554
200	.590
270	.216
Pan	.164
AFS Fineness No.	53.31
AFS Clay, %	0.308

TABLE 2

Mix No. 1

1200 lb Lake Sand	— 100	per cent
2.5 lb Southern Bentonite	— .21	per cent
15 lb Light Weight Flour	— 1.25	per cent
18 lb Water	— 1.50	per cent
5 qt Core Oil	— .81	per cent

Mix No. 2

Same as above except 30 lb of water (2.5%) was used

Mix No. 3

Same as above except 42 lb of water (3.5%) was used

Mixing Cycle

1 Minute Dry
3 Minutes Wet
3 Minutes Oil

corded to indicate the density, observed stickiness, and behavior of the sand in hopper.

3. After every 5th core the drag half of the corebox was removed from the coreblower and any sand adhering to the corebox was carefully brushed into a pan and weighed. At least 20 cores were blown from each core mix.

4. Five cores from each mix were weighed in the green state to determine the average blown weight of each core.

5. Cores from each mix were baked at 450 F for approximately 1 hour.

6. The surface condition of each baked core was observed and a rotary scratch hardness test performed on the flat surfaces. Figure 1 illustrates the observed core density and surface condition.

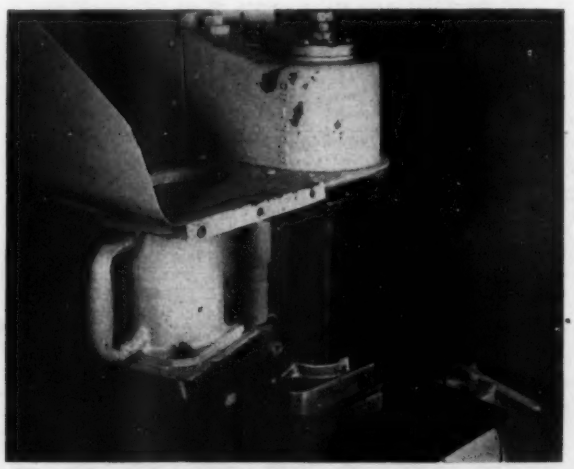


Fig. 2 — Coreblower used in this study.



Fig. 3 — Gelometer used in this study (Courtesy of Lauhoff Grain Co.).

7. At the same time that the tappet cores were being produced, the following tests were performed in the sand laboratory on each core sand mixture:

a. Green Strength
b. Sag resistance
c. Overhang ability
d. Moisture
e. Specific weight
f. Baked tensile and scratch hardness
g. Weight of retained sand in a stickiness test mold was determined. The blowing set-up is illustrated in Fig. 2. The cumulative weight of retained sand on the pattern after five successive blows was recorded.

8. Figure 3 illustrates the instrument used to measure the force required to pull a 5° and 15° truncated cone from a rammed 2-in. standard specimen. This instrument is a gelometer.

Test Results

The Stickiness Committee was concerned mainly with the effect of varying moisture contents on a production core sand mixture. Three core sand mixtures were evaluated with every thing held constant except for the moisture added to the sand which was 1½ per cent, 2½ per cent, and 3½ per cent respectively.

Results of Production Tests

An aluminum corebox was conditioned initially with kerosene and the sand was mulled and delivered to the production coreblower as described in the *Test Procedure* Section. As each core was blown the blowing characteristics, apparent stickiness, density, and overall quality of the core was observed. After each five cores were blown, the drag half of the corebox was brushed free of any adhering sand and this sand was weighed. Five cores from each mix were baked in a core oven at 450 F for approximately 1 hour. The total weight of five green cores was also determined as a measure of density. Listed on the following page are the visual results of each test:

Mix No. 1 (1.5 Per Cent Added Moisture)

1. Cores blew well and were dense and appeared to be hard.
2. Sand did not hang up in hopper.
3. No sign of sticking at any time during the blowing of all 20 cores.
4. Necks of all the cores cracked or were broken while placing in the oven, a sign of low green strength.
5. Five green cores weighed 99 lb or 19.8 lb each.

Mix No. 2 (2.5 Per Cent Added Moisture)

1. First core looked perfect and no visible sticking.
2. Sand started to hang up in hopper and partially blown cores resulted from Core No. 1 to No. 10. This did not appear to contribute to observed stickiness.
3. Slight stickiness appeared to be concentrated at the steel inserts in the drag corebox directly under the blowholes in the cope.
4. After the 10th core was blown the stickiness appeared to diminish and was negligible by the time the 20th core was blown.
5. Some porosity was noticeable in the cores but not excessive.
6. Five cores weighed 103.2 lb or 20.6 lb each.

Mix No. 3 (3.5 Per Cent Added Moisture)

1. Slight sticking occurred immediately around the steel inserts in the drag corebox.
2. At core No. 4 sections of the core pulled away from the core at the steel inserts.
3. A man was required full time to push the sand down into the sand magazine because of excessive sticking or bridging in the sand storage hopper. A vibrator was tried in the cycle of the coreblower but was not sufficient to cause the sand to flow down the hopper.
4. Cores No. 6, 7, 8, 9, and 10 sticking excessively.
5. Cores also had weak blow areas and excessive porosity.
6. At cores No. 17 and 18 stickiness was nearly gone and at 19 and 20 not very noticeable. Porosity condition and weak blow areas remained.
7. Five cores weighed 104 lb or 20.8 lb each.

Baked Production Core Results:

Figure 1 illustrates the condition of the baked cores for mixes No. 1, 2, and 3. The scratch hardness and baked tensile properties are also given in Table 3. The scratch hardness and tensile increased with moisture but is still a little low for some production foundries.

Apparent Stickiness and Green Property Test Results:

Tables 1 and 2 list the sieve analysis and the three core mixes tested in this evaluation. Tables 4 through 8 list the green and baked properties the sand mixtures plus the retained sand test results and the gelometer test data. Note all mulling times given do not include the 1-minute dry mixing period.

TABLE 3 — BAKED PROPERTIES OF THE THREE CORE MIXTURES TESTED (Average of 5 Tests)

Mix No.	Tensile Strength, psi	Hardness	Hardness on Production Valve Tappet Cores (Average of 12 tests)
1	128	63	59
2	187	75	67
3	188	76	74

TABLE 4 — GREEN STRENGTH TEST RESULTS

Sand Mix No. 1 (1.5% H_2O)			
1A	1B	1C	1D
0.62	0.95	1.10	1.55
0.62	0.93	1.14	1.56
0.59	0.96	1.14	1.54
0.59	0.91	1.10	1.52
0.56	0.90	1.09	1.51
Average 0.596 psi	0.930 psi	1.114 psi	1.536 psi
Sand Mix No. 2 (2.5% H_2O)			
2A	2B	2C	2D
0.98	1.13	1.14	1.30
0.92	1.10	1.14	1.26
0.94	1.12	1.15	1.27
0.94	1.10	1.16	1.26
0.91	1.14	1.16	1.26
Average 0.938 psi	1.118 psi	1.150 psi	1.270 psi
Sand Mix No. 3 (3.5% H_2O)			
3A	3B	3C	3D
1.00	1.10	1.15	1.19
.97	1.12	1.10	1.17
.96	1.12	1.14	1.18
.96	1.11	1.10	1.19
.97	1.14	1.12	1.19
Average 0.972 psi	1.118 psi	1.122 psi	1.184 psi

A = 3 Minutes mull water, cereal and clay
B = 4 Minutes mull water, cereal and clay and oil
C = 5 Minutes mull water, cereal, clay and oil
D = 6 Minutes mull water, cereal, clay and oil

TABLE 5 — GREEN PROPERTIES OF EACH SAND MIXTURE

Overhang	Sag Resistance 5 Jolts	10 Jolts	15 Jolts	20 Jolts	25 Jolts	Moisture, %	Sp. Wt., grams
			Mix No. 1 (1.5% H_2O)				
16	.0060	.0080	.0090	.0100	.0110	.95	
15	.0055	.0065	.0075	.0850	.0095	1.05	167
16	.0060	.0070	.0080	.009	.010		
15							
15							
Avg. 15	.006	.007	.008	.009	.010	1.0	167
			Mix No. 2 (2.5% H_2O)				
34	.0070	.0100	.0125	.0150	.0170	2.3	168
30	.0050	.0750	.0100	.0120	.0135	2.2	
33	.0065	.0100	.0125	.0145	.0165		
30	.0065	.0100	.0125	.0145	.0160		
31							
Avg. 32	.0065	.0095	.012	.014	.016	2.25	168
			Mix No. 3 (3.5% H_2O)				
31	.0075	.0110	.0130	.0150	.0170	3.05	
30	.0090	.0135	.0175	.0210	.0230	2.95	
33	.0090	.0135	.0165	.0185	.0200		171
29	.0080	.0115	.0140	.0165	.0180		
30	.0085	.0115	.0140	.0165	.0185		
Avg. 31	.0085	.012	.0150	.0175	.0195	3.0	171

TABLE 6 — SAND RETAINED ON THE DRAG HALF OF A PRODUCTION VALVE TAPPET COREBOX IN GRAMS

Core No.	Weight Collected
Mix No. 1 (1.5% H_2O)	
5th Core	0.3245 grams
10th Core	0.3052 grams
15th Core	0.2617 grams
20th Core	0.2782 grams
Mix No. 2 (2.5% H_2O)	
5th Core	0.3970 grams
10th Core	1.5010 grams
15th Core	0.1087 grams
20th Core	0.2068 grams
Mix No. 3 (3.5% H_2O)	
5th Core	0.3245 grams
10th Core	0.3052 grams
15th Core	0.2617 grams
20th Core	0.2782 grams

TABLE 7 — SAND RETAINED IN A SPECIAL DIETERT STICKINESS TEST COREBOX, CUMULATIVE AFTER FIVE BLOWS

Mix No.	Weight Retained, grams Cope	Drag	Total, grams
1	0.0441	0.0498	0.0939
2	0.0496	0.0512	0.1008
3	0.0255	0.2945	0.3200

1. Green Strength

Figure 4 illustrates the effect of mulling time on the green strength and the resultant effect of moisture also. Samples of sand were removed from the muller after mulling the sand cereal, clay and water for 3 minutes and the 4, 5, and 6-minute cycle with the core oil in the sand. Figure 5 compares the effect of mulling time on the green strength for each of the three core mixtures.

TABLE 8 — FORCE REQUIRED TO PULL 5° AND 15° TRUNCATED TEST PATTERNS FROM A STANDARD AFS TWO INCH SPECIMEN

Cores were rammed five times and cones cleaned with Tricloralethylene before each group of five. Results are the average of five tests.

Mix No.	5° Cone	15° Cone
1	224.8 grams	128.8 grams
2	176.0 grams	111.2 grams
3	163.2 grams	125.6 grams

These two curves indicate the following:

A. The highest green strength was obtained with the sand mix of the lowest moisture.

B. Peak green strength was obtained after 6 minutes mull on all mixes.

C. It is apparent that maximum green strengths are reached at the 3.5 per cent moisture levels at all mulling periods except the 6-minute mulling periods where a peak is reached at 1.5 per cent moisture. (See Fig. 4)

Figures 6 and 7 compare the effect of moisture on the amount of retained sand in the drag half of the valve tappet corebox at different times of the production cycle.

A. Mix No. 1 (1.5 per cent moisture) which exhibited little or no stickiness, left very little sand in the drag half of the corebox.

B. Mix No. 3 which exhibited excessive stickiness in the corebox also retained the largest amount of sand in the corebox.

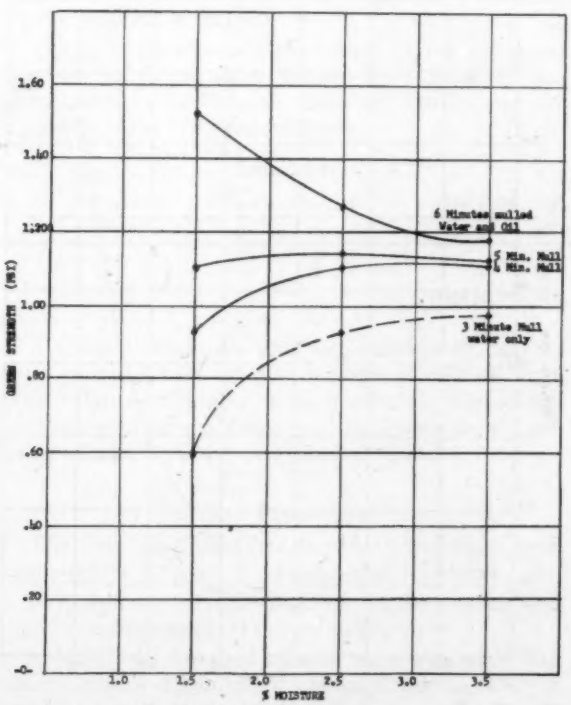


Fig. 4 — Green strength vs moisture and mulling time.

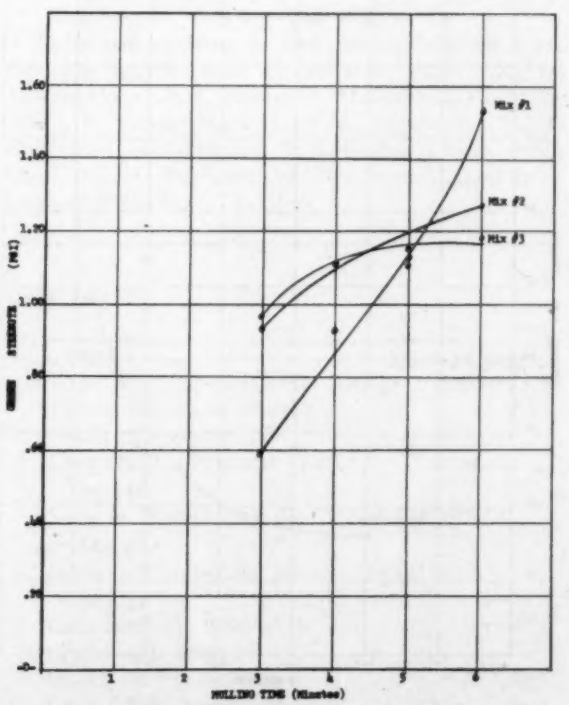


Fig. 5 — Green strength vs mulling time.

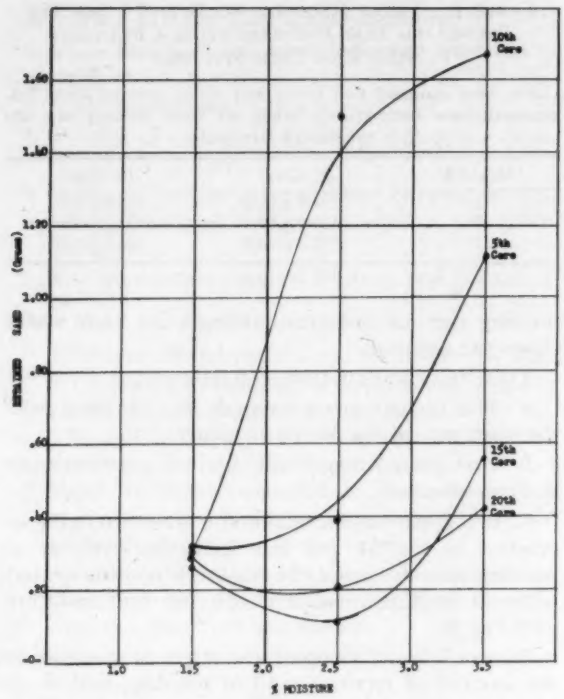


Fig. 6 — Per cent moisture vs retained sand and number of cores blown (production cores).

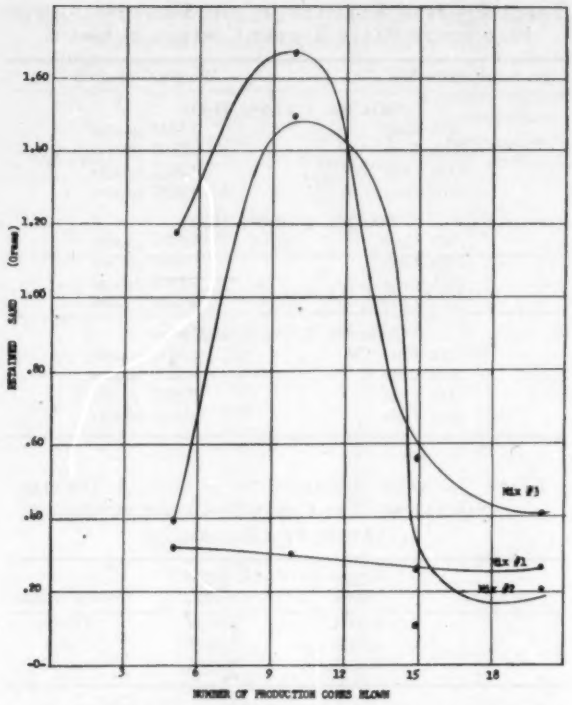


Fig. 7 — Number of cores blown vs retained sand for the three core mixtures (production cores).

C. Both mixes 2 and 3 had very little retained sand after 15 cores were blown and also exhibited very little stickiness in the corebox after the 15th core was blown.

D. In general, as the moisture increased the retained sand increased, particularly during the first ten blows of the coreblower. This correlates very closely with the observed stickiness in the corebox.

Figures 8 and 9 are comparisons between the relative amount of sand retained in the production

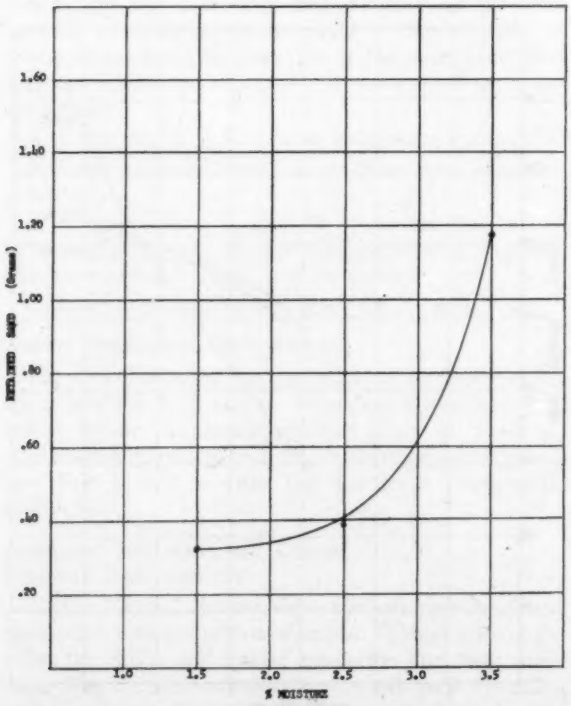


Fig. 8 — Retained sand vs moisture after blowing five production cores.

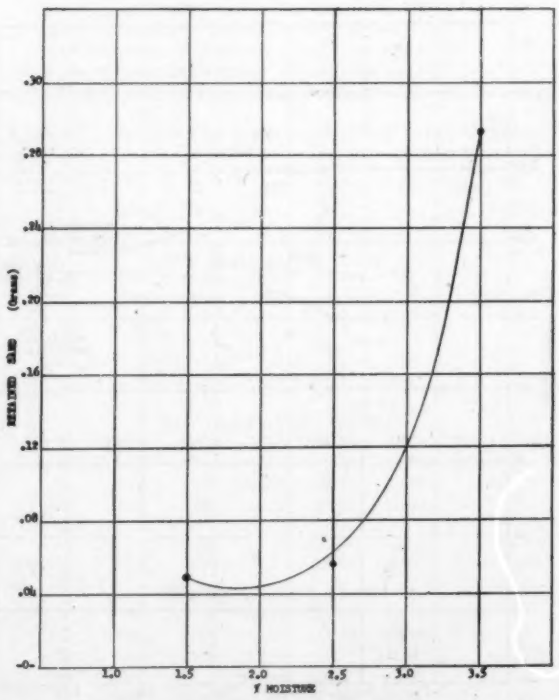


Fig. 9 — Retained sand in drag half of Dietert stickiness test pattern vs moisture after blowing five cores.

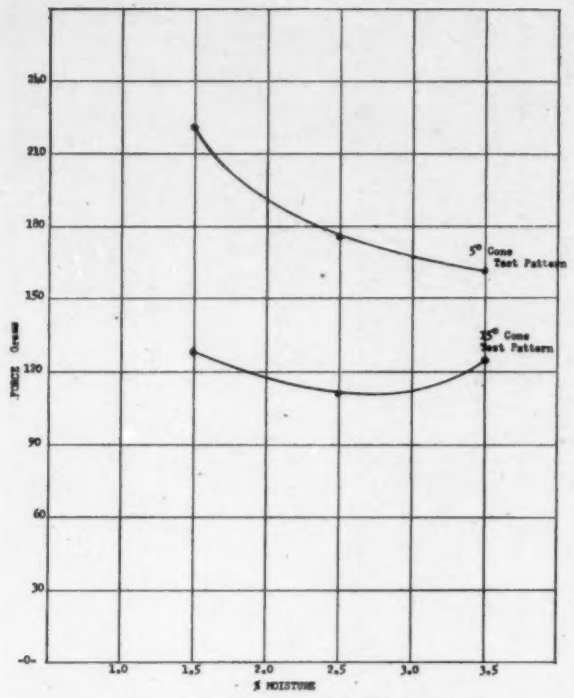


Fig. 10 — Force required to pull 5° and 15° cone from standard 2-in. specimen vs moisture using Lauhoff Grain Co. gelometer test machine.

corebox after five blows with that of the Dietert stickiness test corebox. The amount of sand retained of course, is large with the production corebox but the general slope of both curves are the same. This indicates that both tests are comparable and they both correlate to a degree with the observed stickiness in the production corebox.

Figure 10 summarized the results of the gelometer test on all three core mixes. Very little if any correlation exists between these test results and the moisture of the observed stickiness.

Summary

1. The lower moisture core mixes exhibited less stickiness and appeared to have more blowability and flowability.

2. The retained sand test both in the laboratory and in the production corebox correlated very closely with each other and also with the stickiness.

3. The results of this work were applied in an automotive foundry and all moisture contents were reduced on all core mixtures on the average approximately 30 per cent with a large improvement in stickiness, blowability, flowability and density.

4. Very little correlation exists between the gelometer test results and apparent stickiness in the production corebox.

Conclusions

1. Results of this study indicate that the lower moisture levels the less the apparent and real stickiness in a standard core oil mixture.

2. The retained sand test both in the laboratory and performed in a production corebox correlate very closely with each other and also with the actual stickiness as observed in a production operation.

3. No correlation appears to exist between the force required to pull a pattern from the core sand and the observed stickiness as found in a production corebox.

Recommendations

Results of this study indicate that the retained sand test method has some correlation to the actual stickiness found in a core mixture, therefore it is recommended that more work be done to further establish this correlation by varying other important stickiness factors and producing actual cores on a production basis.

Acknowledgment

Those participating in and carrying out the work reported herein were as follows: From General Motors Corp., A. L. Bryant, S. McComb, D. S. Mills, and J. R. Young; from Harry W. Dietert Co., R. T. Daksiewicz, R. Glass and A. L. Graham; from Aristo Corp., W. H. Buell and M. Dombrowski; and from Lauhoff Grain Co., J. E. Huss.

PROGRESS REPORT NO. 3

A series of stickiness tests were performed on core sand mixtures varying only the type of core oil used. The relative stickiness of each core oil was known and the stickiness test was used to determine the relative stickiness of the core sand mixture.

Test Procedure

The core sand mixtures shown to the right were evaluated in this study of core and stickiness.

1. A dummy mixture was run in the muller and each ingredient weighed out carefully.

2. Each mix was used as soon as it was made and approximately 35 minutes was required to blow 50 cores.

Mix A	Mixing Cycle
Lake Sand (50 AFS) 35 lb (100%)	1 minute
Cereal 159 grams (1.0%)	
Water 325 grams (2.0%)	2 minutes
Core Oil A 159 grams (1.0%)	3 minutes
Mix B	
Same as 'A' replacing the 'A' core oil with 'B'	
Mix C	
Same as 'A' replacing the 'A' core oil with 'C'	
Mix D	
Lake Sand (50 AFS) 35 lb	
Cereal 159 grams	
Water 302 grams	
Resin 318 grams	



Fig. 11 — Test corebox used in this study (Courtesy of Harry W. Dietert Co.).

3. The test cores were blown in the corebox illustrated in Fig. 11.

4. Before blowing the first core for each sand mix the corebox was sprayed with kerosene and the excess blown off with compressed air. The box was not sprayed again with kerosene until a new core mix was used.

5. The sand adhering to the corebox was weighed on an analytical balance to the nearest milligram after each five cores were blown. A total of 50 cores were blown for each mixture.

6. The adhering sand was brushed off carefully with a soft-haired brush from the cope and drag half of the test corebox.

Test Results

Basic core oil mixtures A, B and C were selected for a series of tests using the retained sand stickiness test patterns. The relative stickiness of each of the three different core oils was known by only one member of the Committee. The difference in stickiness was reported small between two of the core oils and one core oil had proven to be a good non-sticking core oil in actual production. It was felt that if the retained sand test could identify these three core oils and pick out the small difference in stickiness between them, then the test would be valid at least for this type of a core sand mix.

Listed below are the main characteristics of the three core oils tested:

Core Oil A and B

These oils are almost identical except for the boiling or evaporating range of the solvents used.

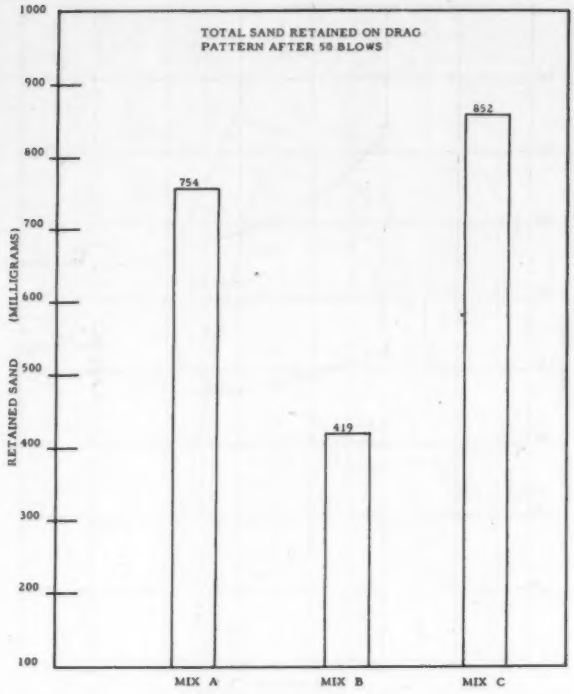


Fig. 12 — Sand in milligrams retained on test pattern for each core sand mixture after 5 and 10 blows.

Core oil A contains the more volatile solvent and core oil B contains the less volatile solvent. The viscosity of each oil was the same and was 300 seconds (Saybolt viscosity at 100 F). Core oil B was reported as the least stickiest oil and core oil A the next least sticky.

Core Oil C

This oil has the same type of solvent as core oil A but has a higher Saybolt viscosity of 450 seconds. This oil was reported as the stickiest core oil of all three.

All of the core oils can be expected to produce approximately the same range of tensile strengths.

Each core sand mixture was mixed and handled using the procedure described under *Test Procedure*.

TABLE 9 — RETAINED SAND IN GRAMS FOUND ON THE STICKINESS TEST PATTERN FOR EACH CORE SAND MIXTURE

No. of Blows	Mix A			Mix B			Mix C			Mix D
	Drag Pattern	Cope Pattern	Total	Drag Pattern	Cope Pattern	Total	Drag Pattern	Cope Pattern	Total	Drag
5	0.039	0.149	0.188	0.027	0.143	0.170	0.100	0.173	0.273	4.670 grams
10	0.065	0.194	0.259	0.052	0.175	0.227	0.104	0.230	0.334	
15	0.051	0.265	0.316	0.032	0.191	0.223	0.043	0.245	0.288	
20	0.057	0.242	0.299	0.037	0.180	0.217	0.052	0.248	0.300	
25	0.082	0.245	0.327	0.036	0.237	0.273	0.106	0.295	0.401	
30	0.095	0.284	0.379	0.042	0.220	0.262	0.085	0.140*	0.225*	
35	0.112	0.239	0.361	0.052	0.191	0.243	0.115	0.299	0.414	
40	0.091	0.232	0.323	0.056	0.229	0.285	0.127	0.368	0.495	
45	0.096	0.273	0.369	0.058	0.186	0.244	0.065	0.264	0.329	
50	0.066	0.249	0.315	0.027	0.172	0.199	0.057	0.306	0.363	
Totals	0.754	2.372	3.126 grams	0.419	1.924	2.343 grams	0.852	2.568	3.423 grams	4.670 grams

*Note: Some of the sand adhering to the corebox was spilled during handling. (Approximately 200 grams).

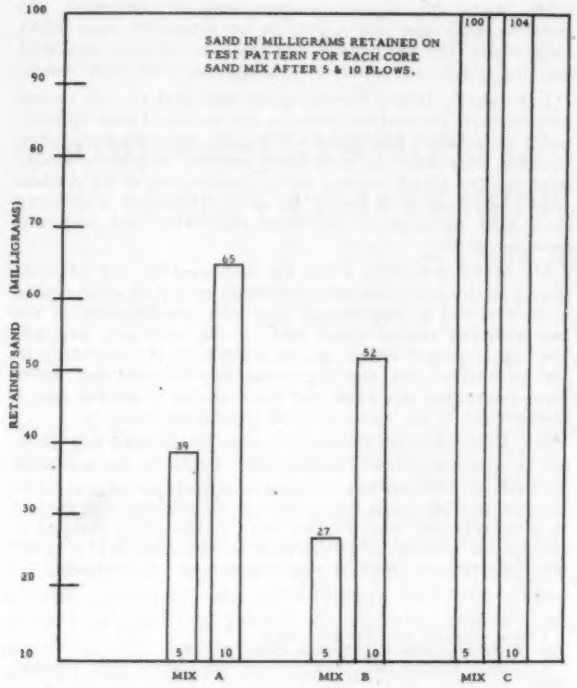


Fig. 13 — Total sand retained on drag pattern after 50 blows.

Fifty test cores for each mix were blown in the test box illustrated in Fig. 11, and the retained sand on both the cope and drag were weighed after each five cores were blown. The results of each test are given in Table 9, as well as the sand retained on the drag pattern for the very sticky resin core mix 'D'.

The sand shown adhering to the drag plate in Fig. 11, illustrates the general pattern left by the very sticky resin-bonded core sand mix 'D' made without a release agent. The core oil-bonded sand covered as wide an area on the test pattern, but only a few grains of sand were retained compared to the illustrated resin mix. Figure 11 also illustrates that the most critical stickiness area in a corebox is directly beneath the blowholes. This was verified during the production tests covered in Report No. 2.

Figure 12 is a bar chart illustrating the weight of sand in milligrams retained on the drag plate of the stickiness corebox after five and ten cores were blown from mixtures A, B, and C.

It is evident that mix 'B' left the least amount of sand on the corebox with 'A' next and mix 'C' leaving almost four times as much sand in the corebox as mix 'A'. This places mix 'B' as the least sticky, mix 'A' next and mix 'C' as the stickiest core oil-sand mix. This correlated exactly with the performance expected of core sand mixtures made with these core oils in actual practice. This test was able to differentiate easily between the three different core oils used to bond the core mixtures A, B, and C.

Figure 13 is a bar chart reporting the total sand retained on the drag half of the test corebox blowing 50 cores and brushing out the adhering sand after each five cores were blown. Mix 'C' still appears to be the stickiest mix and mix 'B' as the least sticky. The total sand retained for mix 'C' would be approximately 200 milligrams higher if some of the adhering sand had not been spilled during handling of the sample. Regardless of which total weight of adhering sand is used to determine the stickiness of either of the three mixtures, it appears 'B' is the least sticky and 'C' is the most sticky of the three core oil-bonded sand mixtures.

Table 9 also illustrates the irregular pattern of retained sand left by each core mix as each successive series of five cores were blown up to 50. The corebox was sprayed with kerosene at the beginning of this test, blown clean, and never touched again with kerosene. After approximately ten cores were blown the retained sand reached a peak and then dropped until the 25th or 35th core where the amount of retained sand peaked again only to fall off sharply by the time the last or 50th core was blown. Because of this irregular pattern it appears that the retained sand adhering to the test pattern should be brushed off and weighed at the 5th or 10th blown core.

The decrease in stickiness or amount of retained sand appearing after 50 blows might be caused by the release action of the solvent in the core oil which gradually forms a film in the corebox over a period of time. This may also indicate that the application of repeated sprays of kerosene or a release agent before blowing each core is detrimental to stickiness rather then helpful. This was also borne out in Report No. 2 where a high moisture and sticky production mix lost a considerable amount of its stickiness after blowing approximately 15 cores without applying any kerosene release agent after the initial application.

Conclusions

1. The retained sand test as used in this study was sensitive enough to determine the relative stickiness between three different core sand mixtures, two of which would have only a slight stickiness variation in actual practice.

2. The degree of stickiness as measured by the amount of sand retained on the drag half of the corebox correlates closely with the stickiness expected from these core sand mixtures in actual practice.

The core sand mix containing core oil 'B' (a known non-sticking core oil) left the least amount of sand adhering to the test corebox and the mix bonded with core oil 'C' (an oil which is known to be sticky in practice) left the most sand adhering to the test pattern.

3. More consistent or accurate results are possible if only the sand retained on the drag half of the corebox is weighed and used as a measure of the relative stickiness. This is true mainly because the majority of the sand adhering to a blown corebox is concentrated directly under the blowholes. This area under the blowholes appears to be sensitive to stickiness whenever a sticky mix is used (Fig. 11).

4. All test cores should be blown and not rammed and at least five cores should be blown in a clean metal corebox before carefully brushing out and weighing the adhering sand. If possible the retained

sand must be weighed out to the nearest milligram. It appears that the larger the test corebox and the more blowholes used the more accurate the stickiness test will be.

Recommendations

Results of this series of tests indicate that the retained sand method of measuring stickiness is sensitive even to small differences in stickiness, therefore it is recommended that further work be done to substantiate the validity of this method of measuring stickiness.

DISCUSSION

Chairman: J. O. OCHSNER, Crouse-Hinds Co., Syracuse, N. Y.
Co-Chairman: E. C. ZIZRZOW, Werner G. Smith, Inc., Cleveland.
Secretary: D. J. JONES, New Jersey Silica Sand Co., Millville, N. J.

J. G. HOUSE[1]: Has any thought been given to making the Stickiness Test applicable to resin and sodium-silicate-CO_2 cores?

MR. MILLS: Preliminary work indicates that the test is applicable to resin binders also. Figure 11 illustrates the pattern left by a resin-bonded sand mix without a release agent. This test also appears to have some merit with green or uncured sodium silicate mixes but very little correlation when the core is cured by CO_2 in the corebox.

C. L. BOWMAN[2]: It is our experience that a change from one oil to a different oil develops core stickiness in the corebox. What type of metal was used in the corebox used in test and also in production?

MR. MILLS: Our experience agrees with Mr. Bowman's, and certainly some core oils appear to be inherently more sticky than others. The test corebox was made out of cold-rolled steel and the production corebox is aluminum with steel inserts.

J. E. BOLT[3]: It was interesting to note that on the various core mixtures the sticking tendency was decreased after approximately ten blows. Significantly the same thing happens when changing over from an oil-bonded mixture to a resin-bonded mixture. The author's theory that a conditioning of the corebox takes place appears to be in line with our finding in running resin sand mixtures on coreboxes previously used with oil-sand mixtures.

MR. MILLS: A corebox which has been used for core oil sands should be cleaned thoroughly with steam or a good solvent prior to its use with a resin-bonded sand mix. Conditioning of the box with any release agents used in the resin-sand mix will also insure proper release of the corebox at the beginning of the production cycle. Our experience also indicates that under some production conditions too much release is sprayed into a corebox and at too many intervals promoting stickiness.

MR. MILLS *(Author's Closure):* In order to provide a tool which can be used economically and quickly in most of the foundries' laboratories, the Stickiness Committee is going to try and adapt the present blow-tensile test corebox to the retained sand method of measuring stickiness. Further work is also being planned to change one variable at a time in a conventional mix to determine the relative effect of that variable on the stickiness.

---

1. Dow Chemical Co., Bay City, Mich.
2. Dalton Foundries Inc., Warsaw, Ind.
3. General Electric Co., Chemical and Metallurgical Div., Pittsfield, Mass.

SOME OBSERVATIONS ON THE TRANSVERSE TEST AT ELEVATED TEMPERATURES FOR MOLDED SAND MIXTURES

By

D. C. Williams*

The transverse test for determining and evaluating properties of molded sand mixtures is seldom, if at all, used in the foundry industry. The test is simple to make as various pieces of equipment are available for this purpose. Knowing the breaking strength and the dimensions of the test piece, one can readily, if desired, compute the modulus of rupture. The amount of test data presented herein is limited yet there may be readers who find this exploration of sufficient interest that the study would be continued.

The equipment used for the tests reported herein was the transverse test accessory for the Dietert sand strength machine.

All theories concerning plastic flow in soil have been developed using the concept of an ideal soil and the assumption is made in this paper that the plastic flow in sand mixtures, being molded, has sufficient similarity to the ideal condition to permit the use of the appropriate theories. Terzaghi† describes the movement of soil particles as follows for a particular condition; "If the soil support of a continuous footing yields, all the soil particles move parallel to a plane which is perpendicular to the center of the footing."

For our purpose we have assumed that the one-piece rammer head of the transverse accessory when placed upon a mass of sand mixture to be molded can be likened unto a continuous footing. When the ramming force is applied, the rammer head can be said to yield by downward movement. During the yielding of the rammer head, the sand grains, while molding, would move in a plane perpendicular to center line of the rammer head. This action pictures the grains moving along planes of sliding toward the direction of the long sides of the core box. It is further conceived that the planes of sliding grains nearer the rammer head will move sooner against the sides of the core box with tighter compaction.

Such compaction will build up frictional resistance which the ramming energy must overcome before any energy is available to cause grain movement along additional planes of sliding deeper in the core box. Therefore, as additional rams are applied, a continuously decreasing quantity of ramming energy is available to cause plastic flow during ramming. Therefore, in the molded mass, the compacted density decreases as the distance from the rammer head increases. The properties of the molded mass will change as the density changes and, consequently, any test value derived from such a molded mass is thought to be some average value for the mass.

It is difficult to separate a given test specimen into parts for investigation and be reasonably sure that tests on the parts can be interpreted. Because of the above described plastic flow of the sand mass during molding, it appeared that making the transverse test by setting various faces of the test specimen on the transverse test supports, it would be possible to make some assessment of the effects of the plastic flow.

The core box and hopper were hand filled and every effort was made to carry out this operation uniformly for all test specimens. After inspection of the test results, the conclusion was reached that this operation would bear much further investigation.

All test specimens (1 in. x 1 in. x 8 in.) were molded using three rams employing the standard sand rammer fitted with the transverse accessory rammer head. The sand mixtures tested in the molded form and their designations in Tables 1 and 2 are as follows:

Sand Mixture	*Designation*
4 per cent western bentonite 96 per cent N. J. No. 60 sand	Bentonite
4 per cent western bentonite 10 per cent silica flour 86 per cent N. J. No. 60 sand	Silica Flour
4 per cent western bentonite 1 per cent corn flour 95 per cent N. J. No. 60 sand	Corn Flour

*Department of Metallurgical Engineering, Ohio State University, Columbus, Ohio.

† Terzaghi, K., *Theoretical Soil Mechanics*, Wiley and Sons, 1944, page 133.

TABLE 1 — MODULE OF RUPTURE OBTAINED BY PLACING VARIOUS FACES ON THE SUPPORT

Mixture:	Bentonite			Silica Flour			Corn Flour		
Supported Face:	Top	Bottom	Side	Top	Bottom	Side	Top	Bottom	Side
	33.3	26.3	36.4	63.0	60.4	60.4	32.7	28.8	21.6
	27.2	38.0	33.1	52.2	52.2	55.0	23.5	20.2	21.0
	37.1	14.4	31.5	63.4	55.0	50.4	16.9	27.8	11.9
	37.8	25.4	29.7	53.6	41.7	48.1			
	40.8	30.6	27.0	58.3	49.0	38.7			
	33.3	31.5	31.8	42.0	36.8	34.0			
	35.2	38.7	38.6	61.9	38.7	37.8			
	37.2	38.7	39.2	46.5	50.7	47.7			
	35.4	29.8	30.4	55.6	38.0	51.3			
Average	35.3	30.4	33.1	55.2	46.9	47.0	24.4	25.4	21.5

The moisture content of all sand mixtures was 4 per cent and all test specimens were dried at 212 F for two hours. The test specimens were broken using the transverse accessory affixed to the sand strength machine.

Tests were made placing various faces of the test specimens upon the two transverse supports. The top face designation indicates the face against the rammer head was placed upon the supports and the bottom face designation indicates that the face opposite that adjacent to the rammer head was placed upon the supports. The side face designation indicates that one of the side faces was placed upon the supports.

An inspection of the data of Table 1 shows for the test specimens molded from the bentonite and silica flour mixtures that the average top face value for the modulus of rupture was slightly higher than the average values for the other faces. However, the individual modulus values for each face condition overlap so much that only average values can be considered if one wishes to state there is a difference. For the corn flour mixture certain low values for the modulus affect the average values.

One possible conclusion to be drawn from Table 1 is the rigidity of a molded mass is related to the direction in which such a mass is subjected to transverse stresses. A core which fails when placed in a mold in one position might prove satisfactory if rotated 90 or 180 degrees. Dimensional tolerances for internal dimensions might be closer controlled if the most rigid position of the core was ascertained. More rigid cores might be produced by studying the core ramming methods and parting surfaces for assembled cores.

The data of Table 1 also point up that whenever the transverse test is used as a laboratory tool for evaluating molded sand mixtures that core face placed against the transverse test, supports should always be the same. Results in Table 1 also indicate that more than three individual test values are necessary in order to obtain an adequate average value. Further studies may reveal how to mold more adequate test specimens.

In addition to the irregularities of density created within the molded mass due to the relation between plastic flow and ramming energy there is the possibility of irregularities caused by the drying of the core. Whether a test core is dried or baked, it will be found that at some time during the application of heat that the core plate becomes quite wet with water.

During the heating, water is moved about within the core in two methods. First, water that passes into the vapor state can be lost through pores on surfaces exposed to furnace atmosphere. Water vapor can also condense on sand grains in portions of the core which have not attained the temperature level for evaporation. A manifestation of such condensed water is clearly observed when a core not fully dried or baked is removed from the core plate and a wet surface is found on the plate.

The water vapor condensed within the core will provide a water content considerable in excess of that originally distributed through the molded mass. It is not difficult to conceive that certain portions of a core could approach or become saturated with water. Such very high moisture contents could have a marked effect upon the bond characteristics in portions of cores which were the locales of condensed water. The bond could migrate with the water or the water could react with the bond and either strengthen or weaken that portion of the core. Future investigation may show that these considerations may be of as much significance as plastic flow characteristics in producing cores of uniform strength. It was not determined if the movement of condensed water during drying had any effect in producing the differences of results shown in Table 1.

Elevated Temperature Tests

Following the studies for the transverse properties of dried test specimens, it was deemed desirable to investigate the transverse properties of test specimens at elevated temperatures. Tests are conducted at elevated temperature levels for the hot compressive strength and expansion characteristics of molded sand mixtures and it was thought that the transverse characteristics would be a contribution to the problem of evaluating molded sand mixtures. It was not possible to obtain transverse test values at temperature levels as low as 1600 F for many test specimens ruptured or deflected due to their own weight before the completion of the 20-minute heat soaking period. The 20-minute heat soaking period was chosen to insure as well as possible the same attained temperature levels

TABLE 2 — TIME IN MINUTES AND SECONDS FOR COMPLETE RUPTURE DUE TO OWN WEIGHT
(Minutes — Seconds as 0-40)

Mixture: Supported Face:	Bentonite			Silica Flour			Corn Flour		
	Top	Bottom	Side	Top	Bottom	Side	Top	Bottom	Side
				Furnace at 1600 F					
	0-40	1-19	1-20	10-0	>20-0	1-20	7-20	9-42	>20-0
	0-43	0-40	5-43	2-05	10-52	1-43	>20-0	7-52	11-30
	0-44	1-40	5-25	9-30	>20-0	1-11	12-0	>20-0	>20-0
	1-55	0-21	0-20	1-57	10-15	>20-0	>20-0	9-16	>20-0
	>20-0	0-28	1-19	1.03	>20-0	>20-0		>20-0	>20-0
	0-48	0-18	2-06	>20-0		>20-0			
	>20-0	0-40	12-20						
				Furnace at 2000 F					
	0-38	1-32	1-23	0-42	1-05	0-32			
	0-42	1-23	1-42	0-41	0-33	0-20			
	1-08	1-14	0-54	1-20	0-58	0-59			

at the surface and center of the test specimen.*

Test specimens molded from the bentonite and silica sand mixture would, upon introduction into the test furnace, maintain an excellent rigidity up to the time of rupture when due to their own weight, they would break into approximately two equal parts. Test specimens molded from the corn flour mixture would exhibit a marked deflection prior to breaking into two parts. Table 2 gives the time, in minutes and seconds, for complete rupture of test specimens supported on various faces when subjected to shock heating at temperature levels of 1600 and 2000 F.

From an inspection of Table 2 it will first be noted that the times to produce failure vary over considerable lengths of time. From the variation of module of rupture shown in Table 1, the data of Table 2 is not unexpected. Those test specimens which appeared to maintain the rigidity for at least 20 minutes would fail when the support carrying the test specimen was given the slightest jar when loading contact was brought into contact with the test specimen.

Data of Table 2 indicate that the rigidity behavior of a core subjected to thermal shock is unpredictable and variable. That is, failure due to its own weight will vary from a few seconds to several minutes at a temperature level of 1600 F. At 2000 F, the time required for failure was very short; less than two minutes. This short life for rigidity coupled with contact forces of turbulent metal flow could warp if not break a core subjected to appropriate transverse stresses.

One core sand mixture was made using materials supplied by the Utica Radiator Co., and was compounded as follows: 1.5 per cent core oil and 98.5 per cent sand. The cores were baked at 450 F for 1.5 hours. Table 3 gives the rupture times due to own weight for various faces at two temperature levels.

Upon subjection to thermal shock the organic material started to burn and continued to do so for two minutes at 1600 F; while at 2000 F, burning gases lasted only one minute. This is additional evidence that considerable time is required before the organic material in a core becomes completely gas forming. Immediately upon extinction of the flame, very irregular longitudinal cracks appeared on the side faces of the test specimens. Such cracks were thought to result from the method of filling the core box before ramming.

Several transverse test specimens of the oil-sand mixture were submitted to thermal shock at a temperature level of 2500 F. The modulus of rupture values for baked test specimens were 297, 274 and 355 psi, when the bottom face of the test specimens was placed upon the supports.

Thirty seconds after insertion into the furnace, the test specimen began to deflect and continued such action until contact was made with furnace floor. The distance deflected was one inch. The surface of the test specimen showed little or no cracking. By observation, it was noted that no cracking of the core on the lower face became visible until the deflection approached one inch.

Many times it is expedient to produce castings in which cores are placed in a horizontal position in the mold. Upon shakeout, the holes are not straight and horizontal, but bowed downward as if the core had sagged. The above reported observations might give a clue to a mechanism to explain the sagging of cores in castings.

If the core is not too large and the filling of the mold cavity was not rapid enough, a core would deflect downward upon being subjected to radiant heat

TABLE 3 — TIME IN MINUTES AND SECONDS FOR COMPLETE RUPTURE DUE TO OWN WEIGHT
(Minutes — Seconds as 7-0)

Supported Face:	Top	Bottom	Side
		Furnace at 1600 F	
	7-00	5-20	5-34
	7-10	7-13	3-03
	7-13	6-03	6-03
		9-00	
		Furnace at 2000 F	
	3-30	3-17	3-15
	2-53	2-01	3-17
	3-05	2-45	3-33

*Williams, D. C., and Kyle, P. E., "Seventh Annual Report on Investigation of Properties of Steel Sands at Elevated Temperatures," AFS TRANSACTIONS, vol. 55, p. 607 (1947).

thermal shock. The degree of deflection would depend upon how soon the molten alloy made contact and supported the core.

The test specimens observed in the furnace were not reinforced by rods. If rods are used, they expand as they become heated and can deflect in any direction. Noting that an unreinforced core will deflect in 30 seconds, it appears that such a core, if reinforced, would offer little resistance to moving along with any deflection of a core rod. A reinforced core could deflect as long as it was surrounded by liquid alloy. Therefore, with large castings, a core which is in position at the completion of the pouring operation would not necessarily be so at the time of final solidification. This kind of deflection is applicable to horizontally or vertically placed cores.

COMPENSATION TRENDS IN LOSS OF HEARING

By

Floyd Frazier*

Background of the Problem

The human race has been subject to loss of hearing in some degree throughout the history of mankind. To understand why loss of hearing has within recent years become a workmen's compensation problem it is necessary to review some of the historical background of compensation laws.

Before the advent of the workmen's compensation laws in the United States the employee's legal remedy for an industrial injury was governed by common law. The employer had available the three common law defenses of (1) assumption of risk, (2) acts of a fellow servant, and (3) contributory negligence. Even if these hurdles were surmounted, the worker had no assurance that damages could be collected from employers financially unable to make payment. As a result of this situation great hardship resulted to injured employees and their dependents.

To remedy these weaknesses inherent in the common law system, workmen's compensation acts were adopted. These acts were developments from long established principles in several European countries. As early as 1854 Prussia required employers to contribute to sickness association funds. In 1884 Germany adopted the first workmen's compensation system and in 1897 Great Britain adopted similar laws. The first compensation law in the United States was passed in 1908 and covered civil service employees of the Federal Government. By 1910 the first state law had been adopted and since that time all states have passed similar acts, the last being in Mississippi in 1949.

As originally conceived, compensation laws covered only accidental injuries but eventually the trend was toward amendment of the laws to include both accidents and occupational diseases.

Athough the compensation laws of the various states are not all alike, nor are they interpreted uniformly by the courts, the general qualifying condition is that the accident must arise out of and in the course of the occupation.

The coverage for occupational diseases may be more limited than for accidental injuries. In some states, only certain diseases which are listed in a schedule are covered, all others being excluded. Other states use varying language stating that the disease must be due to a specific industrial hazard, leaving the courts to determine whether the disease actually arose out of the employment.

The Wage Loss Concept

The original philosophy behind and purpose of the workmen's compensation laws has been to guarantee to the worker a portion of the wage loss caused by accidental injury or occupational disease. It has, therefore, been generally required that before the employee can establish a claim for compensation he must have suffered an actual wage loss or impairment of earning capacity. Most states provide a waiting period, requiring the worker to be absent from the job for a specified number of days before compensation payments may be started. However, many states list in a schedule certain types of impairments for which a fixed amount of compensation is paid regardless of the time actually absent from the job. These are usually referred to as schedule losses. This practice has been described as simply a method of estimating in advance the probable eventual reduction in earning capacity.

Factors in Compensability of Loss of Hearing

Severe cases of hearing impairment due to noise exposure are normally developed only by workers in extremely noisy occupations such as boiler making, drop forging, rock drilling, etc. In these occupations, the very noise which creates the hearing impairment is usually so intense that communication must be by means other than speech.

Thus, the hearing impairment seldom occasions any lack of ability to continue in the normal occupation and seldom does it result in inability to continue to earn full wages. The same type of hearing loss which results from noise exposure may, however, occur as a result of other causes having no connection with the occupation. These will be discussed later.

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Developments Affecting Legislation

Within recent years legislative action and court precedents in a few states have placed the loss of hearing problem in a different light from that described above for compensation cases.

New York Precedents

In May, 1948 the New York State Court of Appeals ruling on the case of Slawinski v. J. H. Williams & Co. held that compensation was payable for hearing impairment under the schedule even though no loss of earnings was involved. Subsequently, the courts rendered a similar decision in the case of Rosatti v. Despatch Shops.

Since the above decisions several hundreds of cases have been filed in New York but no awards have been made because of a ruling of the Workmen's Compensation Board that the worker must be absent from the injurious noise for at least six months before the extent of permanent hearing loss can be determined. This ruling was based on medical opinion that some portion of the hearing loss will be recovered after removal from the noise but that until further research is completed the exact period of time necessary cannot be stated. The ruling is also supported as an expedient to forestall a flood of claims until solutions which are economically feasible and fair to both employers and employees can be devised.

Wisconsin Precedents

In Wisconsin a similar decision was made by the Supreme Court in the case of Wojcik v. Green Bay Drop Forge Co. To prevent a possible surge of claims which it was recognized that industry was not in a position to handle, the Legislature passed a temporary law, effective July 1, 1953, requiring that in such cases the worker must have sustained an actual wage loss. A new law was passed, effective July 1, 1955, which enables payments for loss of hearing under a schedule without a loss of earnings, but which requires the worker, among other conditions, to show the extent of permanent hearing loss after six months absence from the noisy environment. The effect of this provision is to delay the filing of claims until retirement or job severance for other reasons, thus enabling industry to build up some reserves for claims settlements.

Trends in Other States

Numerous legislative proposals have within the past year been introduced in various states and others are expected to be advanced. In general however, the tendency in most states has been to await the results of legislative action and test litigation in New York and Wisconsin. It is to be expected that the patterns established in these states may affect developments in others since both states have long been leaders in labor legislation of all types.

Due to lack of test litigation it has not been determined whether loss of hearing due to noise is compensable in numerous states. In others compensation is barred, under the conditions previously described, by requirements that the loss be total, or that there be a demonstrated loss of wages or of earning capacity. Numerous legislative proposals, designed to clarify or change the intent of the compensation laws are being advanced. Thus, any statement concerning the specific coverage conditions in individual states would have little meaning. For information on this subject, inquiry should be made to competent legal advisers in the particular state involved.

Causative Factors in Loss of Hearing

Much of the difficulty of fixing the responsibility and determining the remedies arises from the facts that noises which damage hearing are not restricted to industry and that noise is not the only cause of the specific type of hearing loss to which industrial workers are subject. Some of these factors are indicated below.

Causes Not Restricted to Industry

Standards of audiometry are discussed in another section. For the purpose of this discussion, it is sufficient to state that in general, hearing losses of less than 15 or 20 decibels are not considered significant. In fact, there is medical evidence that a hearing loss of 15 decibels or less actually constitutes no loss whatever but is simply the limit of variability in hearing of normal persons. With these factors in mind, the following examples of widespread existence of hearing losses become significant.

1. School Surveys. In hearing tests conducted in elementary schools throughout the United States, it was found that an average of five per cent of all school children had hearing losses of 15 or 20 decibels in one or more frequencies.

2. General Population. Dr. Howard House[1] and other authorities have made the statement that 10 per cent of the total population of the United States has significant hearing losses.

3. Job Applicants. Numerous published statements[2] have indicated that in pre-employment audiometric examinations it has been found that at least 25 per cent of all job applicants have significant hearing losses.

Types of Hearing Losses

The medical aspects of the various types of hearing losses are discussed in another section. It is, however, necessary to mention them briefly in this section to indicate their importance in the compensation problem.

In general, hearing losses are of two types (1) conduction, and (2) nerve impairment types.

1. Conductive Impairment. These types of losses are not of industrial origin. They occur when there is interference in the passage of sound through the external ear, the ear drum, or the middle ear region. These conditions may result from foreign bodies in the outer ear canal, infections in the middle ear area, trauma to the ear drum, or blockage of the eustachian tube. Fortunately, this type of hearing loss can readily be identified by the otologist.

2. Inner Ear or Nerve Impairment. The hearing losses caused by noise are of this type. The mere fact that a nerve impairment hearing loss exists does not

mean that it was due to industrial noise. Unfortunately, in the absence of supporting historical data, it is difficult or impossible for the otologist to distinguish between these causes. Among non-occupational causes are some of the common childhood diseases, sensitivity to certain drugs, and presbycusis, which is the loss of hearing which accompanies the aging process in every individual. In New York and Wisconsin, in recognition of this latter factor, deductions are made from the percentage hearing loss for each year over 50 years of age. Recent research indicates that the effects of presbycusis may be significant at an earlier age, possibly around 40.

Even though the hearing loss is conceded to be due to noise there remain unsolved problems. Noise induced hearing losses have two components. The first of these is auditory fatigue, which is a temporary effect from which the worker recovers upon absence from the noise, probably within 48 hours or less. The remaining loss is believed to be due to nerve degeneration. It is commonly believed that some portion of this remaining loss may be recovered after prolonged absence from the noise but research groups have not yet determined the exact period of absence necessary.

Sources of Exposure

The exact type and duration of noise exposure which will produce hearing loss remains to be established through research. Although a number of factors including the overall intensity of the noise, the frequency of the sounds which make up the overall spectrum, the length of the exposure and difference in individual susceptibility remain to be determined, several authoritative medical groups have tentatively estimated that the level at which possible damage starts is in the range of 90 decibels. A report by a committee of the American Standards Association[3] indicates the possibility, however, that hearing losses at some frequencies may develop from exposure to noises with decibel intensities in the low eighties. This poses the problem that noises which have a harmful effect upon hearing are encountered off the job as well as on and that the general public, as well as industrial workers, is exposed in some degree. The following list of typical exposures on and off the job demonstrates this point and indicates that full responsibility for all hearing losses even though due to noise cannot logically be assessed against industry.

Source	*Decibel Level*
Boiler Factory	107
Inside 4 engine airplane	100 and up
Punch presses	96-110
Subway train passing station	105
Loud radio or phonograph	90-100
Wood saw	100
Automobile	92
In a modern (quiet) streetcar	90
Noisy Factory	90
Piano	80-100
Printing presses	80

Economic Impact

From the foregoing discussion it becomes apparent that one of the major problems is to determine the degree of responsibility which should be assessed against industry for an impairment which is not limited to industrial workers. Furthermore, since in the past industry has kept records of neither noise levels in the plants nor audiometric records of workers' hearing it is impossible to determine during what periods of past employment hearing losses actually developed.

To estimate the probable cost of indiscriminate payment of claims for loss of hearing is an impossible task. As has been indicated, probably 10 per cent of the total population has hearing impairments. Based on recent records of pre-employment tests it seems obvious that the percentage among industrial workers is higher, probably about 25 per cent. The number of potential claimants could, therefore, run well into the millions.

Because widespread claims have in most states been retarded by the legislative developments there are few precedents for determining the probably average value of claims. The maximum benefits in the schedules of most of the states for loss of hearing were established primarily to deal with total loss of hearing resulting from accidental injuries. Total loss seldom occurs from noise. With these facts in mind, it is of interest to note the wide variation in possible maximum benefits in various states or jurisdictions.

Under the Federal Employees Compensation Act, which covers civil employees of the Federal Government, an employee may receive as much as $6,300 for total loss in one ear, and $24,200 for both ears.

The Longshoremen's and Harbor Workers' Compensation Act allows $1,820 for total loss in one ear, and $7,000 for both ears. The variations in allowances in different states for loss of hearing due to traumatic injury range from a high of $12,333 in Wisconsin to a low of $1,560 in Maine for total loss in both ears. For loss in a single ear the high is $2,700 in Oregon and the low $625 in Kansas.

From the above facts it becomes obvious that if the laws were amended or interpreted to permit widespread payment for losses of hearing regardless of cause, the total cost could run into billions of dollars.

The Problem of Accrued Liability

Spokesmen for industry have asserted that to require industry at this late stage to assume liability for all hearing losses which have developed in the past would impose a financial burden which could not be borne. Unlike the problems which arose with other types of occupational diseases, such as silicosis and lead poisoning, the impairment is not limited to relatively small groups of workers in certain industries or occupations. Every industry is involved to some extent. Because the problem was not recognized as one calling for compensation, self-insured industry has not, in the past, accumulated reserves with which to pay such claims. Insurance carriers have collected no premiums for this purpose because claims experience,

upon which rates are based, has been lacking. Thus, the sudden imposition of a possible tremendous accrued liability cost could make it impossible for insurance carriers to insure many types of industrial concerns unless some relief is provided through legislation.

The situation could be similar to that which existed in the state of New York some 20 years ago when silicosis suddenly became a similar problem. Because small businesses could not withstand the financial burden which was threatened, and could not secure insurance to relieve them of the liability, many of them were threatened with bankruptcy and it has been estimated that from 35,000 to 50,000 workers became unemployed. To forestall the flight of industry from the state and to preserve the financial stability of industry, the Legislature passed emergency legislation which drastically limited liability as of that date and gave industry and the insurance carriers time to accumulate reserves before claims could develop substantial costs.

Solutions — Proposed and in Effect

To provide solutions which may be fair to workers, to industry, and to the public numerous solutions have been considered. Some of these are indicated below.

Restoration of Wage Loss Concept

Because of the many factors involved numerous industry groups feel that the only fair solution to the problem is to limit compensation to those cases in which there is a demonstrated loss of wages or of earning capacity depending upon the intent of the laws of the state involved. This opinion is supported by the fact that the basic intent of workmen's compensation is reimbursement for wages lost as a result of occupational injury or disease and that in the absence of such loss there is no justification for compensation payments.

Limitations on Liability

Partly as a means of keeping potential costs within limits which industry can assume, various limitations on liability have either been enacted or proposed. In Wisconsin, the maximum benefit payable for loss of hearing due to noise has been reduced to approximately one half that payable for sudden loss of hearing due to traumatic injury. Such a measure also has some measure of medical support since it is conceded that hearing loss due to noise occurs gradually over a long period of years thus giving the worker time to adjust himself to the change whereas a loss due to traumatic injury occurs suddenly and causes a drastic readjustment of the worker's life.

Another form of limitation which has been considered is the escalator type of law such as was adopted in New York State to deal with the silicosis problem. In that case, liability for disability existing as of the date of the law was limited to $500 and this maximum was permitted to increase at the rate of $50 per month until a maximum compensation of $3,500 was reached. Payments for partial disability were eliminated.

Schedules With Restrictive Conditions

In some states measures are either in effect or have been proposed whereby payments for loss of hearing are made under a schedule which limits the total number of weeks of compensation which may be paid. The Wisconsin provision has been mentioned. It not only limits the amount of compensation but requires that the worker be removed from the noisy exposure for at least six months as a result of job severence, retirement, layoff or other specified cause.

Methods for Determining Compensation Payable

Although the methods for measuring the hearing loss at various frequencies are fairly well standardized there has been no general agreement concerning the methods by which these records should be translated into percentage disability for compensation purposes. A few of the significant measures are indicated below:

1. Limited Frequencies. Although audiometric measurements of hearing loss cover a range of from 125 to 8000 cycles per second, authoritative medical groups[4] state that for compensation purposes only the losses in the three frequencies of 500, 1000 and 2000 cycles should be considered. One reason for this opinion is the belief that the primary function of hearing is to understand speech and that these three frequencies are all that is necessary for this purpose. This measure eliminates consideration of losses in the high frequencies which develop first with presbycusis and from other causes and which ordinarily have no disabling effect.

2. Elimination of Minor Losses. A method which is in use in the state of Wisconsin and which is being considered in other states consists of eliminating payment for hearing losses which average less than approximately 15 decibels. Thus compensation is eliminated for minor hearing losses which are seldom disclosed except through audiometric tests and which have no disabling effect whatever.

At the other end of the scale, any hearing loss of 80 decibels or more is considered total loss of hearing. Although such losses seldom occur from industrial noise it is possible to measure hearing losses on the order of 100 decibels.

3. Deductions for Presbycusis. Although in some states deductions are made for that portion of the hearing loss which is the natural accompaniment of age there are exceptions. Presbycusis is an established factor and several recognized tables are available for placing definite values on it. The losses from this cause are most noticeable in the higher frequencies but with advancing age, extend into the speech frequencies of 500, 1000 and 2000 cycles. This is indicated in Table 1.

4. Age Reduction Factors. In some states (Wisconsin is an example) a deduction is made because of advancing age from compensation paid for any form of schedule disability. The justification for such deductions is that compensation is intended to pay either for loss of wages or of earning capacity and that the older the worker, the less potential earning capacity can be lost through disability.

A further extension of the age reduction principle

TABLE 1 — PRESBYCULIS TABLE[5]
Hearing Loss in Decibels

Men, Age	500 Cycles	1000 Cycles	2000 Cycles	3000 Cycles	4000 Cycles
25	0	0	0	0	0
30	0	0	1	2	4
35	0	0	2	5	7
40	2	2	4	8	12
45	2	3	8	12	15
50	3	4	11	16	21
55	5	6	14	21	26
60	6	8	18	25	32
65	7	10	22	30	37

is being considered in some localities. In view of the wage loss basis for compensation, it is contended that if the worker remains employed until retirement age without having suffered any disability or wage loss there can be no justification for payment, at retirement age, of compensation for hearing loss which has existed in the past. On this basis, proposals have been advanced to reduce compensation payable by a specified percentage with each year of advancing age with the result that benefits would be largely eliminated if a claim is not established prior to retirement.

Eventual Basis of Solution

There can be no definite solutions to many of the problems presented above until much more has been learned through research. Since hearing losses due to noise develop over long years of exposure, it is necessary to make repeated hearing tests, over long periods of time, upon workers exposed to known types of noises.

Because, in the absence of adequate past records, it is impossible to determine what portions of hearing loss are due to present and past employments and to occupational and non-occupational factors, it is essential that industry not only take steps to control noise as far as possible but also to maintain adequate medical records including pre-employment and periodic audiometric retests so that the origin of hearing losses may be established.

Once adequate medical and scientific data have been accumulated the eventual solution will lie in compromise considering on the one hand what is desirable from the standpoint of the worker who suffers an actual impairment and on the other hand, the extent of the financial burden which can be imposed upon the employer and enable him to continue to operate and remain solvent.

References

1. Dr. Howard P. House, Chairman, Subcommittee on Noise in Industry, American Academy of Ophthalmology and Otolaryngology.

2. Scholz, W. F., "Industrial Hearing Loss," April 1955 *Best's Fire & Casualty News;* also,

House, Howard P.; "Problems Encountered in Attempts to Standardize Methods of Calculating Hearing Disability"; 41st Annual Convention of International Association of Industrial Accident Boards and Commissions, Chicago, Illinois, September 20, 1955.

3. "The Relations of Hearing Loss of Noise Exposure," a report from Subcommittees Z24-X-2 of American Standards Association.

4. See the following:

a. "Extension 8, Schedule for Rating Disabilities, 1945 Ratings for hearing impairments," U. S. Veterans Administration.

b. "A scale of social competence for hearing," Committee on Hearing, National Research Council.

c. "Report of Medical Advisory Committee of Advisory Committee on Workmen's Compensation on Legislation," Wisconsin Industrial Commission.

5. This table is an average, compiled by the Research Staff on the Subcommittee on Noise in Industry, and is based on the averages of result of three studies shown in the report of Subcommittee Z24-X-2 (See Reference 3).

DISCUSSION

Chairman: ALLEN BRANDT, Bethlehem Steel Co., Bethlehem, Pa.

Co-Chairman: K. M. SMITH, Caterpillar Tractor Co., Peoria, Ill.

N. S. SYMONS* *(Written Discussion):* The whole problem is a complicated one due to our lack of acceptable criteria for determining not only what are hazardous industrial noise levels but how to determine what part, if any, of an employee's hearing impairment is due to his occupation.

Mr. Frazier has mentioned two matters which deserve particular emphasis.

1) Departure from Wage-Loss Concept

Our Workmen's Compensation system was established, not as a damage statute, but as a form of partial protection against income loss[1]. The schedules referred to by Mr. Frazier were, historically, intended to apply only to "accident" cases, and the current tendency to merge the concept of the schedule (in which the wage loss is presumed rather than having to be proved) into our "occupational disease" statutes is fraught with grave danger.

Accidents (which depend on the factor of chance) and occupational diseases are two entirely different matters because of such questions of how and when they are caused, the incidence of potential claims which can develop (this affects cost), and the matter of the accrued liability. If in the occupational disease field we are going to compensate for purely physiologic loss unaccompanied by any economic loss there is no telling where the process will stop[2].

Having in mind the many organs and senses of the body that may deteriorate due wholly or in part to our complex industrial environments, the carrying of this principle to its ultimate conclusion could put a monumental and perhaps impossible cost burden on American Industry. This matter should be seriously considered by those interested in finding a proper solution for the problem. The plain fact of the matter is that the partial impairment of hearing caused by industrial noise does *not* cause any loss of wages during the life time of the employee.

2) Problem of Accrued Liability

Equally serious is the problem of accrued liability for which no provision has been made either directly or in the insurance rates. The difficulty in suddenly being asked to pay a large bill for *past* liability neither provided for nor legally anticipated is obvious. It is vastly different from the matter of accidents where the insurance rates are set up to cover events happening *in futuro*. Mr. Frazier mentioned the silicosis problem in New York in the thirties which involved this principle.

The present problem of accrued liability is more serious than it was in the silicosis situation because industrial noise is far more prevalent than was the silicosis hazard, and the exposure is as unquestionably more difficult to control. When, either by legislation or by Court rulings, we create obligations stemming from past exposures which, during their existence, were not regarded as creating either legal or economic rights, we are headed for genuine trouble.

To quote Henry D. Sayer, the recently retired General Manager of the New York Compensation Insurance Rating Board, "In assuming now to make provision out of the moneys of industry, for the effect of years of work in an environment of noise, we are assuming an ability to do the impossible."[3]

*Past Chairman, Insurance Law Section of New York State Bar Association, Special Legal Consultant on Industrial Noise to Associated Industries of New York State; member of latter's Technical Committee on Noise in Industry, and member of law firm of Brown, Kelly, Turner & Symons, Buffalo, N. Y.

If the wage-loss principle is not adhered to in these cases it is imperative that some proper solution be found for this aspect of the problem either in the form of a law creating an escalator type of liability or in some other way.

Current Developments in New York State

Since Mr. Frazier remarked that the tendency in most states has been to await the results of further action in Wisconsin and New York I thought it might be appropriate to conclude my comments by summarizing the present status of the problem in my home state of New York.

The *Slawinski* case[4] was decided in 1948. This resulted in a substantial increase in the number of claims filed although a completely ruinous avalanche of costly awards was prevented by the adoption by the Workmen's Compensation Board of the six-month waiting period which prevented the granting of any award as long as a claimant continued to work in the injurious exposure.[5] This ruling, still in effect, does not *bar* but merely *postpones* the award.

In the meantime Industry became increasingly aware of the grave potentialities of the problem. It knew that the six-month waiting period was purely an administrative ruling which could be repealed over night. Consequently, acting through the leadership of Associated Industries of New York State, it initiated vigorous efforts to find a proper legislative solution which would protect the competitive economy of the state.

These efforts culminated in the introduction in the 1955 session of the Legislature of a bill, known as the Morgan-Milmoe bill, to restore the wage-loss principle and to require actual proof of a causally related hearing impairment before an award could be granted.[6]

A public hearing was held on this bill in Albany on March 2, 1955 before the Joint Legislative Committee on Industrial and Labor Conditions, at which time its merits were debated by representatives of Industry and Labor.[7] The bill was actively supported by 31 other co-sponsoring employer groups including Inter-Allied Foundries of New York State.[8]

In the closing days of the legislative session this bill was debated on the floor of the Assembly where it was passed by a vote of 76 to 68. However, subsequent intensified opposition to the bill by Labor and by the Workmen's Compensation Board made it necessary for the bill to be recommitted without a vote in the Senate.[9]

Thus, while no law was enacted, these proceedings accomplished much good by bringing the problem out in the open and by educating the members of the Legislature and the public as to its implications and potential gravity. It was not without significance that at least one branch of the Legislature, after full debate, reaffirmed the legislative intent that the wage-loss concept should be applied to this type of case.

The next development was that in the early fall of 1955 the Joint Legislative Committee on Industrial and Labor Conditions suggested that representatives of Industry and Labor confer to see if some satisfactory legislative solution of the problem could be agreed upon. This suggestion was accepted and representatives of Industry and Labor started a series of conferences to discuss the problem.

At one of the early meetings an agenda of items for discussion was agreed upon and I see no impropriety in stating, without expressing any opinion upon them, that those items included (a) the problem of accrued liability, (b) definition of what constitutes an injurious noise level, (c) the question of loss of hearing due to aging (presbycusis and what allowance should be made for this factor, (d) the six-month waiting period, (e) the law of presumptions applicable to these cases, (f) length of exposure prerequisite to the granting of any award, (g) question of pre-placement examinations and relieving the employer from liability for any hearing impairment existing at the time of hiring, (h) what frequencies to use in determining the percentage loss of hearing, (i) use of a binaural method of determining percentage loss of hearing, (j) elimination of minor losses not causing any social or conversational inconvenience or disability, (k) use of an age reduction factor to reduce award as the employee nears retirement age, (l) use of best of several audiograms which give different readings, (m) amount of schedule if schedule is used, and (n) the question of allocating the liability where there are several successive employments each contributing to the hearing loss.[10]

The nature of these items which indicate the extent and complexity of the problem, are discussed, directly or indirectly, in Mr. Frazier's paper. If the problem cannot be solved in any state by the retention or restoration of the basic wage-loss principle these matters should all be given careful consideration and study. Without intelligent correlation of the interlocking acoustical, medical, actuarial, social, economic and legal aspects of the problem it will not be possible to find a satisfactory solution which will be fair to all segments of our economy.

Noise Control

The ideal solution would be the elimination of the hazard and Associated Industries of New York State has, along with its efforts in the legislative field, not neglected this important phase of the problem. On the contrary, and in recognition of Industry's obligation to furnish its employees with the safest possible working environment, it has created a Technical Committee of Noise in Industry[11] which, beginning in 1953, has sponsored three state-wide conferences dealing with various technical phases of the problem including practical methods of reducing and isolating noise, plant conservation of hearing programs, personal ear protection for employees, etc. These conferences have been extremely well attended. Much more can and will be done in this virgin field.

In Conclusion

In a recent speech Richard B. Cooper, Assistant Vice President of the American Mutual Liability Insurance Company, said:[12]

> "That solution, whatever it may be, will be acceptable and permanent only to the extent that it is based on *humanitarian considerations* as well as *dollar values*. It will have to be a *constructive* answer, not a *defensive* one. It will be found largely in the area of *prevention*, not simply *compensation*. And it will have to recognize the desire of modern management to provide healthful working conditions for its greatest asset—the men and women in its plants."

With this fine expression of the proper method of seeking a satisfactory solution I cannot disagree. However, I should like to amplify what Mr. Cooper said. In the first place if occupational deafness is to be compensated where there is no wage-loss the cost should be within Industry's capacity to pay not only now but in periods of depression when claim filing under our compensation system often increases substantially.

In the second place it should never be forgotten that under the basic pattern of our compensation laws the public ultimately pays the cost of all awards in the price of the manufactured product.[13] Their interest in the problem—and many members of the public have a loss of hearing not due to noise—cannot, in fairness, be disregarded.

Lastly I should like to emphasize the imperative need of cooperation on the part of all concerned in finding the right solution. The problem involves not only socio-economic considerations but what the American Standards Association has referred to as "the disciplines of psychology, biology and physics, and their practical application in audiology, otology and acoustical engineering."[14] In determining how much money should be paid for a given hearing loss all of these many factors have to be considered and weighed. It behooves everybody, therefore, including Industry, Labor and the medical and other professions, to approach the problem with fairness and with what the Chinese philosopher, Lin Yutang, has termed the spirit of "sweet reasonableness". Only in this way can the problem be solved in a way that is fair to the individual employee, to his employer, and to the public.

NOTES AND REFERENCES

(1) See report of the Wainwright Commission appointed by Governor Hughes in 1909 to study the problem in New York (Report to the Legislature of the State of New York by the Commission Appointed under Chapter 518 of the Laws of 1909 to Inquire into the Question of Employers' Liability and Other Matters, First Report, March 10, 1910, New York Senate Documents, 133rd Session, 1910, Vol. 25, 38, N. Y. State Library). Explaining to the Commission the English law which was the prototype of the New York law later

adopted. A. H. Gill a Labor Leader and a member of the English Parliament, said (page 84):

"The basis of compensation is disability to earn wages. A man is not paid for the loss of a limb; he is not paid for the pain he endures; he is not paid for any of the domestic stresses, and so on, that arise from what happened to him; he is paid on the basis of the wages he has lost through being unable to continue his work."

(2) Under Secretary of Labor Arthur Larson stated that ". . . in the case of partial loss of hearing, there is room for plenty of argument that I do not propose to open up at this point. I only want to stress that, as a matter of principle, we must never cut loose from the loss-of-earning-capacity concept, for, if we do, there is no telling where the process will stop". (Larson. A.: *The Future of Workmen's Compensation,* paper presented at the Annual Covention of the International Association of Industrial Accident Boards and Commissions, Chateau Frontenac, Quebec, October 4, 1954). In view of Mr. Larson's subsequent sponsorship of the Discussion Draft of Proposed Model Workmen's Compensation Law prepared by the U. S. Department of Labor which contains a "schedule" more generous than the present New York schedule, it is to be wondered if he has undergone a change of heart (*Highlights of Proposed Model Workmen's Compensation Law,* address given at 20th Annual Meeting of the Industrial Hygiene Foundation, Mellon Institute, Pittsburgh, Pa., November 17, 1955). Because this proposed model bill clearly invades an area reserved to the states it will not be commented upon here except to say that it does not begin to solve the many complex aspects of the problem.

(3) Sayer, H. D. *Occupational Loss of Hearing,* paper presented at Industrial Hygiene Foundation of America, Mellon Institute, Pittsburgh, Pa., November 19, 1952.

(4) *Slawinski vs. J. H. Williams & Co.,* 298 N. Y., 546, reargument denied 298 N. Y. 634.

(5) This requirement was adopted as a result of the recommendation of a Committee of Consultants on Occupational Loss of Hearing Appointed by the Workmans Compensation Board to evaluate the problem. It is based on the medical theory that part of the loss may be a reversible fatigue (i.e. a loss from which there may be recovery) as distinguished from permanent and irreversible nerve degeneration. It is applied upon the legal premise that until there has been the suggested separation from the exposure the net permanent causalty related hearing loss cannot be determined.

(6) Assembly Print 757, Int. 754, Senate Print 958, Intro. 928.

(7) The author presented the views of Associated Industries of New York State and argued that the bill was not only in the interest of the employers but, what was more important, that it was in the best interest of the employees because (a) it would protect the security of their jobs, (b) it would fully protect any employee sustaining a wage loss because of occupational hearing impairment as there would be no statutory ceiling in such a case except the weekly maximum and (c) it would protect the compensation system against a needless drain of its resources which should be preserved for the economic welfare of the state's working population.

(8) Other employer groups sponsoring this bill included the New York Self-Insurers Association, Empire State Forest Products Association and numerous local Chambers of Commerce and local Manufacturers Associations.

(9) A "stop-gap" bill (Senate Print 3680, Intro. 3084), also sponsored by Associated Industries of New York, which would have provided a year's moratorium on all awards "pending further study and consideration of the problem of industrial noise" was, in the closing days of the session, passed by the Senate but had to be recommitted in the Assembly.

(10) There were other matters on the agenda involving purely procedural questions some of which are peculiar to the New York statute.

(11) The Co-Chairmen of the Committee are Frank A. Patty, Chief Industrial Hygienist of General Motors Corporation and Dr. B. L. Josburgh, Manager Health Services, General Electric Company, who recently succeeded Dr. James H. Sterner, Medical Director of Eastman Kodak Company.

(12) Cooper, R. B.: *Occupational Loss of Hearing,* address before the Mutual Claims Conference, White Sulphur Springs, West Virginia, March 27, 1956.

(13) *Cf.* New York State Constitution, Article 1, Section 18. In a New York case, *Reinwald v. Builders Brick and Supply Co.,* 168 App. Div. 425 (1915), the Court pointed out that the New York Legislature, in enacting the Workmen's Compensation law, "put into effect the changed concept that the trade product should be charged with all consequences of inherent trade hazards."

(14) *The Relations of Hearing Loss to Noise Exposure,* report of Exploratory Subcommittee Z 24-X-2 of American Standards Association. This is a Subcommittee of Z 24 Committee on Acoustics, Vibration and Mechanical Shock; Acoustical Society of America, sponsor; report released in 1954.

AN EVALUATION OF ZH62XA MAGNESIUM SAND CASTING ALLOY

By

K. E. Nelson* and W. P. Saunders**

ABSTRACT

Magnesium sand casting alloy ZH62XA (Mg + 5.7 per cent Zn + 1.8 per cent Th + 0.7 per cent Zr) combines a reasonable degree of castability with excellent tensile properties at room temperature. Its creep resistance at moderately elevated temperatures appears to be improved over alloy ZK51A. Alloy ZH62XA can be welded by argon-arc or heli-arc methods and no special procedures are required to apply surface protecting treatments. This alloy is intended to supplement or replace alloys ZK51A and ZK61A which also have excellent tensile properties at room temperature, but because of certain aspects of castability have not gained wide acceptance in the magnesium foundry industry.

Introduction

Selection of an alloy for a specific application is influenced by two factors: (1) efficiency at which saleable castings can be produced and, (2) ability of the metal to withstand the conditions of operation of the part. Mg-Zn-Zr alloys such as ZK51A-T5 and ZK61A-T62 have been reported to have excellent tensile properties at room temperature[1,2,3] but are considered difficult to handle in the foundry.[2,4]

A new magnesium sand casting alloy termed TZ6, using the English alloy nomenclature, and ZH62XA-T5, using the A.S.T.M. system of alloy nomenclature, was introduced by Ball.[5] It was shown that this alloy exhibits high tensile strength at room temperature associated with intermediate creep resistance at elevated temperatures. In addition, it was reported that the alloy was completely free from microporosity and could be handled in the foundry in the same way as other magnesium-zirconium alloys. The nominal composition consists of Mg + 5.7 per cent Zn + 1.8 per cent Th + 0.7 per cent Zr.

The purpose of the work described in this paper is to discuss some of the castability features of this new alloy and to present the effects of variations in composition and different heat treatments on its properties. Creep data are presented and property comparisons among several magnesium sand-casting alloys are given. The term castability, as used in this paper, has been defined before[8] and refers to the ability of an alloy to fill a sand mold satisfactorily and to form a casting relatively free of deleterious defects.

Procedure

Methods used to produce the alloys evaluated have been thoroughly described in earlier publications and need no further description.[4,6-10] Melts were prepared using virgin materials in each case. The aluminum, thorium, zinc and zirconium contents were analyzed by wet chemical methods.[11] The zirconium contents are reported in two parts, "soluble" and "insoluble", referring, respectively, to the portions present in the alloy which are soluble and insoluble in dilute HCl acid.[11]

Standard ½-in. diameter separately-cast test bars, a special casting termed a slope casting and an airframe casting shown in Fig. 1 were used to determine tensile and creep properties of the various alloys. The slope casting is described in another paper (AFS T.P. 56-63) presented before this society.[12]

The slope castings were produced in the laboratory, while the airframe castings were produced in the production foundry. The same gating and risering was used for all alloys for the slope casting; on the airframe casting substantially the same gating was

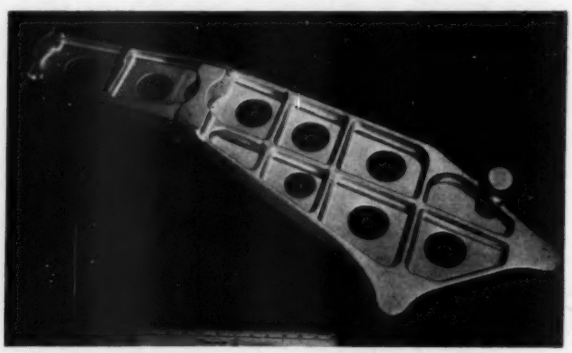


Fig. 1 — Photo of Casting A.

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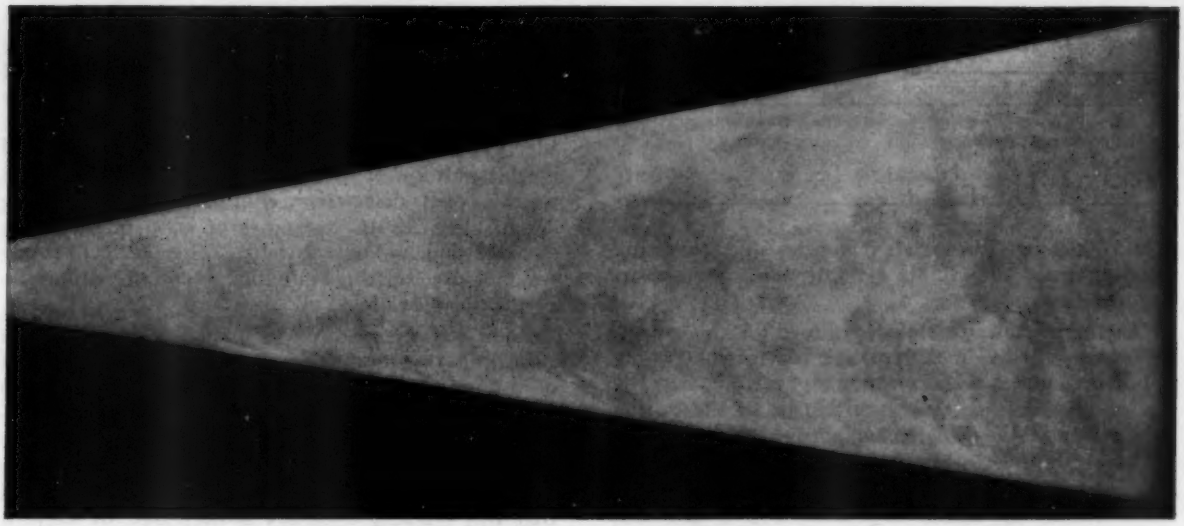


Fig. 2—"Swirl" porosity in ZK51A alloy.

used for all alloys except that additional precautions against dross-entrapment were necessary for those alloys containing zirconium.

The castings were heat treated along with the separately-cast bars in laboratory or production furnaces. The precipitation heat-treatments were performed in regular furnace atmospheres. Solution heat-treatments for AZ92A and ZK61A were carried out under a protective atmosphere containing about 1 per cent SO_2 gas. Sections from the castings were machined to conform to standard A.S.T.M. specified shapes for testing, or, when this was not possible, to dimensions listed as acceptable to the A.S.T.M.

Grain size measurements were made by the comparison method described by George.[13] The methods of tension and creep testing have been discussed in detail in previous publications.[8,14,15]

Discussion of Results

Castability—The effect of varying thorium and zinc contents on the castability of Mg + 0.7 per cent Zr alloys was investigated by pouring melts produced in the laboratory into special test castings. One casting which is termed a draw panel casting was designed to study the effect of various feeding systems on the porosity, draw and surface sink tendencies of the metal. A flat triangular shaped panel casting was used to provide further information on porosity tendency. This triangular panel has been described in an earlier paper.[16] Microporosity in these castings was detected by radiographic examination; draws and surface sinks were determined visually. Thorium additions ranged from 1 to 3 per cent while the zinc additions ranged from 4 to 9 per cent.

The effect of casting design on the incidence of microporosity in a sand casting is a known factor. Results from the above tests further indicate that some castings are more difficult to produce free from microporosity than are others. The draw panel casting shows that in alloys containing zinc and thorium, even 3 per cent thorium is not effective in achieving complete radiographic soundness with less than 7 per cent zinc. In the triangular panel, however, complete radiographic soundness is observed only in alloys containing a minimum of 5 per cent zinc along with 3 per cent thorium. Alloys containing 6 per cent zinc and 1 per cent, or greater, thorium also produce sound triangular panels. In both castings, at any level of zinc, increasing additions of thorium result in decreasing porosity tendency.

Some of the microporosity observed in these alloys is of a type peculiar to Mg-Zn-Zr alloys such as ZK51A. This porosity has been rather descriptively termed "swirl porosity" to differentiate it from the "lined-up" or "layer" type as found in the Mg-Al-Zn system of alloys. Figure 2 is a reproduction of a radiograph of the triangular casting showing "swirl" porosity in ZK51A alloy. Microporosity is influenced by the thermal gradients set up in the mold which are affected by metal flow and heat transfer after the mold is filled. As described by Elliott,[17] "microporosity has its inception late in the solidification process, after a considerable network of solid phase has formed. At this stage, there must be an intergranular flow of liquid through the channels remaining in the solid network, to compensate for the contraction of the remaining liquid phase. Microporosity results when this intergranular flow is inadequate." There are obvious differences in solidification characteristics between Mg-Al-Zn alloys and those of the Mg-Zn-Zr type which make apparent the differences in type of microporosity observed in these two systems of alloys.

The tendency of these alloys toward the formation of draws was determined from observations on the draw panel casting. Draws are surface shrinkage defects and are usually found at the junction between a massive section and a web. This condition results in a surface depression or separation in the casting wall, while it is in the semi-solid state. Increasing thorium from 1 to 3 per cent in Mg + 4-5 per cent Zn + 0.7 per cent Zr alloys shows little effect on the tendency of these alloys toward the formation of draws. At the

TABLE 1 — AN EVALUATION OF VARIOUS ALLOYS POURED IN CASTING A

Alloy	Visual Examination	Radiographic Examination
ZK51A	Draws, cope pits	Porosity in web areas
ZK61A	Small draw, cope pits	Large porosity
ZH62XA	Cope pits	Sound
AZ91C	Good	Sound

6 and 7 per cent zinc levels it appears that increasing thorium from 1 to 3 per cent decreases their draw tendencies.

The term "surface sinks," as used in this paper, refers to a shrink in the massive part of the casting solidifying under poor conditions of feeding. Usually this is on a surface of a massive part of the casting and occurs while the affected portion of the casting is largely liquid. Increasing the zinc content from 4 to 7 per cent in these alloys decreases the surface sinks. Variations in thorium from 1 to 3 per cent appear to have no effect.

While microporosity, draws, and surface sinks are problems to consider, it appears that a magnesium alloy containing 6 per cent zinc, 2 per cent thorium and 0.7 per cent zirconium (ZH62XA) would not present any irresolvable problems in that respect. An alloy containing 7 per cent zinc, 3 per cent thorium and 0.7 per cent zirconium is a further improvement over ZH62XA. However, as described later in this paper, its tensile properties are lower than those of ZH62XA.

ZK51A, ZK61A, AZ91C and ZH62XA (Mg + 5.7 per cent Zn + 1.8 per cent Th + 0.7 per cent Zr) alloys were poured in sand molds to produce castings similar to that shown in Fig. 1 (Casting A). These castings were gated and risered similarly, except that the zirconium-containing alloys were double screened in order to minimize the turbulence formed in the metal stream entering the mold. An evaluation of the quality of these castings as determined by the usual foundry inspection techniques is given in Table 1. While the cope pits[4,7,8,10] are not eliminated by the addition of thorium to ZK61A, the draw and microporosity defects are eliminated in this casting design. The AZ91C casting is of good quality both visually and radiographically. The AZ91C casting was not used in the evaluation of tensile properties of the various alloys.

Additional experience with ZH62XA alloy has been gained in the production foundry since completion of this test program. It has been observed that, in relatively complicated parts, this alloy has a tendency toward more serious cracking than would be expected in alloys of the Mg-Al-Zn system. Although not positively established, it has been conjectured that this cracking tendency is decreased by keeping the alloy saturated with zirconium (greater than 0.7 per cent).

A rough estimation of the linear shrinkage obtained in the mold on freezing was obtained by measuring length of the pattern and length of the casting.

TABLE 2 — COMPARISON OF SHRINKAGE OF VARIOUS MAGNESIUM ALLOYS CAST IN SAND MOLDS

Casting A

Composition	Average Shrinkage, in./ft	% Shrinkage over AZ91C Alloy
AZ91C	0.139	—
ZK51A	0.156	12
ZK61A	0.142	2
ZH62XA	0.150 and 0.156	8 and 12

TABLE 3 — EFFECT OF TREATMENT ON TENSILE PROPERTIES OF Mg-Th-Zn-Zr ALLOYS TESTED AT ROOM TEMPERATURE

Separately-Cast Test-Bars*

Composition, %			F			T5			T4			T6		
Th	Zn	Sol. Zr	% E	YS	TS	% E	YS	TS	% E	YS	TS	% E	YS	TS
0.93	4.00	0.80	15	18	36	8	21	36	18	13	35	12	16	35
1.72	3.97	0.78	9	16	30	5	19	31	18	13	34	17	15	35
2.77	4.04	0.83	7	16	27	5	18	28	13	13	33	13	14	34
1.37	5.01	0.81	10	23	38	7	27	39	18	14	36	15	18	37
1.56	4.98	0.85	14	20	37	9	24	39	14	14	34	12	16	37
3.03	5.18	0.86	5	17	28	5	21	30	17	14	35	16	16	36
1.42	5.90	0.87	10	21	37	9	23	38	18	12	36	14	17	37
1.74	5.89	0.87	8	20	35	6	24	37	11	12	34	14	16	37
3.25	5.91	0.88	7	16	31	6	20	32	8	12	33	14	15	36
1.05	6.88	0.88	6	23	36	3	22	36	12	13	37	9	21	37
2.03	6.74	0.96	8	22	38	5	24	36	16	12	36	12	18	38
3.03	6.65	0.96	8	19	34	4	23	34	14	11	35	12	15	36

* Results are averages of two tests.

% E = Elongation in 2 in.
YS = Yield strength, 1,000 psi
TS = Tensile strength, 1,000 psi

F = As cast
T5 = 16 hr at 400 F
T4 = 16 hr at 960 F
T6 = 16 hr at 960 F, air cool, 16 hr at 400 F

The difference between these measurements is recorded for each alloy in Table 2. This casting was made in a green sand mold and the measurements were made across a relatively non-restrained section. Casting design and the location of sand cores, of course, can change these values, but it is not expected that the relative differences would change. Significant differences are not encountered in the linear shrinkage among the Mg-Al-Zn alloys.

In a recent paper[4] commercial magnesium sand casting alloys were grouped into four classes with respect to castability. They were listed in decreasing order of castability as follows:

1. Mg-Al-Zn alloys except AZ63A alloy in applications requiring maximum freedom from microporosity.
2. Mg-RE-Zr alloys.*
3. Mg-Th-Zr alloys.*
4. Mg-Zn-Zr alloys.

*While the Mg-RE-Zr and Mg-Th-Zr alloys are not subject to microshrinkage problems, the addition of zinc to these alloys further reduces their tendency toward microshrinkage.

Many castability characteristics found in ZH62XA alloy are favorable as regards ease in obtaining a saleable casting. While good quality castings have been produced in this composition, the incidence of oxide skin formations and surface pitting along with draws and cracking problems require that it be classified in castability in group three, above.

Tensile Properties of Mg-Th-Zn-Zr Alloys—Effect of variations in composition and heat treatment on the tensile properties of Mg-Th-Zn-Zr alloys was determined from a large number of tests. Part of the results are reproduced in Tables 3, 4 and 5.

Combinations of thorium from 1 to 3 per cent and zinc from 4 to 7 per cent were added to Mg-Zr-base alloy. Their tensile properties in four different tempers were evaluated at room temperature. The results in Table 3 show that a wide range in properties can be obtained dependent upon the composition and heat treatment. It is obvious that no benefits in tensile strength can be derived from either the T4 or T6 treatments. From this work it appears that a magnesium-zirconium alloy with about 5 to 6 per cent zinc and 1 to 1.5 per cent thorium in the T5 temper has the best combination of ductility and strength.

Effect of time at various aging temperatures on the tensile properties of four Mg-Th-Zn-Zr alloys is shown in Table 4. The two-step treatment of 2 hr at 625 F followed by 16 hr at 350 F is that recommended by Ball.[5] The treatment at 625 F is recommended for stress-relieving of the cast-in stresses in accordance with the findings of Payne.[18] The superiority of the high zinc, low thorium alloy is readily apparent. It appears that a direct aging treatment of 24 hr at 300 F produces a good level of tensile properties. The higher thorium alloys respond to the two-step treatment slightly better than to the age at 300 F.

Thorium contents ranging from 1.1 per cent to 2.3 per cent were added to Mg + 6 per cent Zn + 0.7 per cent Zr alloys. Tensile properties at room temperature obtained on separately-cast test bars aged directly at 300 F are compared with those resulting from several two-step treatments in Table 5. Again there is a suggestion that the two-step treatment (625 F – 350 F) might be superior at the higher thorium levels. The possibility of developing still better properties by lowering the temperature of the second step from 350 F to 300 F and aging for a longer time also exists.

Comparison of Properties of ZH62XA with Other High Strength Magnesium Sand Casting Alloys—Two batches of ZH62XA alloy were poured into slope

TABLE 4 – EFFECT OF AGING TIME AND TEMPERATURE ON TENSILE PROPERTIES OF Mg-Th-Zn-Zr ALLOYS TESTED AT ROOM TEMPERATURE

Separately-Cast Test-Bars*

Composition, %			Time at	Properties at Aging Temperatures															
		Sol.	Aging	275 F			300 F			350 F			375 F			2 hr at 625 F + 350 F		
Th	Zn	Zr	Temp., hr	% E	YS	TS	% E	YS	TS	% E	YS	TS	% E	YS	TS	% E	YS	TS
2.0	4.9	0.96	8	7	20	34	13	22	39	7	20	34	13	21	39			
			16	8	21	36	12	21	39	7	22	36	7	22	36	9	24	40
			24	8	22	37	10	22	39	7	22	36	6	22	35			
			48	5	23	34	10	23	39	8	23	38	9	23	40			
3.0	4.9	0.97	8	6	17	30	4	18	28	5	18	30	5	18	29			
			16	4	18	27	4	18	28	4	18	27	5	19	30	2	21	31
			24	6	18	31	4	19	28	5	19	29	4	19	30			
			48	4	19	30	4	19	29	6	19	32	5	20	32			
1.1	6.0	0.95	8	9	29	42	8	27	42	9	26	41	7	25	42			
			16	4	29	39	9	28	43	8	27	42	8	26	42	6	28	42
			24	10	28	43	10	28	44	7	26	41	9	26	42			
			48	9	29	44	4	29	40	7	26	41	4	26	38			
1.9	5.8	0.79	8	8	22	38	11	22	40	4	21	34	8	22	38			
			16	8	24	39	7	23	38	4	22	34	6	22	37	4	25	38
			24	9	24	39	10	23	41	4	22	35	7	23	38			
			48	9	24	40	8	23	40	9	22	35	7	23	39			

* Results are of one test.

% E = Elongation in 2 in.
YS = Yield strength, 1,000 psi
TS = Tensile strength, 1,000 psi

TABLE 5 – EFFECT OF THORIUM CONTENT AND HEAT TREATMENT ON TENSILE PROPERTIES OF ZK61A ALLOY TESTED AT ROOM TEMPERATURE

Separately-Cast Test-Bars.

Composition, %			300^{24}			$625^{2}-350^{16}$			$625^{2}-392^{16}$			$625^{2}-300^{24}$			$625^{2}-350^{24}$			$625^{2}-400^{24}$		
Th	Zn	Sol. Zr	% E	YS	TS	% E	YS	TS	% E	YS	TS	% E	YS	TS	% E	YS	TS	% E	YS	TS
1.1	6.1	0.95	10	28	44	6	28	42												
1.4	5.7	0.93	5	26	39															
1.6	5.7	0.93							6	24	39									
1.6	6.2	0.89	5	28	40	5	28	40												
1.7	5.7	0.72	10	26	41	6	28	41												
1.7	5.9	0.84	7	27	40	8	28	42												
1.8	5.8	0.98	10	27	41															
1.8	5.9	0.94	5	28	39															
1.8	6.0	0.92	5	28	39				6	27	40	8	29	43	8	28	41	4	26	38
1.9	5.7	0.96							7	26	40									
1.9	5.8	0.96				4	25	38												
2.0	5.8	0.90	6	28	40															
2.3	5.7	0.99	6	24	38															

Note: Superscripts refer to hours at the given heat treating temperature.
*Results are from 1 to 16 tests.

casting molds[12] and separately-cast test-bar molds. These castings were given the two-step treatment. Bars were sectioned only from the bottom level of the slope casting. Their tensile properties at room temperature are compared with similar data on AZ92A-T6 and ZK51A-T5 alloys in Table 6. Except for the lower minimum and average elongations and tensile strengths on the AZ92A-T6, the properties of all three alloys are comparable. These low values are due to the presence of microporosity and to incomplete solution of the compound during heat treatment. The excellent elongation values for ZK51A-T5 and ZH62XA-T51 are obtained from the bar taken at the extremity of the slope casting in the thinnest section. This bar is completely sound and has the smallest grain size.

The airframe casting shown in Fig. 1 (Casting A) and separately-cast test bars were used to evaluate the tensile properties of three different high-strength alloys. In Table 7 the tensile properties of ZK51A-T5, ZK51A-T51, ZK61A-T62, ZH62XA-T5 and ZH62XA-T51 are compared. ZK61A-T62 was heat treated according to recommendations of Meier.[3] Use of the two-step heat treatment on ZK51A and ZH62XA alloys showed no advantage in tensile properties over the direct aging treatment. The properties of separately-cast test bars of ZK61A-T62 are significantly higher than those of the other alloys. However, due to the high incidence of microporosity in the airframe castings, properties of bars sectioned from the castings are no better than those from the ZK51A castings. Actually, the minimum values in both ZK51A and ZK61A alloy castings can be attributed to microporosity. While there is some evidence of microporosity

TABLE 6 – COMPARISON OF TENSILE PROPERTIES AT ROOM TEMPERATURE OF BARS SECTIONED FROM A SLOPE CASTING WITH SEPERATELY-CAST TEST-BARS

	Composition, %				Slope Casting					Separately Cast* Test Bars			
Alloy**	Al	Zn	Sol. Zr	Th	No. Castings	No. Bars		% E	YS	TS		% E	YS	TS
AZ92A-T6	8.75	2.0	—	—	3	18	Aver.	1.5	22.2	34.9	Aver.	1.7	24.0	38.5
							Range	0.5	19.9	24.4	Range	1.0	22.6	34.8
								-3.0	-23.4	-42.5		-3.5	-27.9	-41.0
ZK51A-T5	—	4.35	0.80	—	2	20	Aver.	9.8	22.2	38.2	Aver.	8.1	25.7	40.5
							Range	3.5	19.7	31.1	Range	6-11	25	38
								-20.0	-23.8	-41.4			-27	-42
ZK51A-T5	—	4.18	0.81	—	2	20	Aver.	10.6	21.5	38.0	Aver.	8.2	25.8	39.7
							Range	2.7	20.0	33.1	Range	3.5	25.0	34.5
								-20.0	-23.4	-40.5		-10.0	-26.5	-41.0
ZH62XA-T51	—	5.74	0.94	1.64	2	20	Aver.	6.2	22.8	37.4	Aver.	6.4	24.2	38.8
							Range	1.3	21.9	33.2	Range	2-10	23	36
								-12.0	-24.2	-41.1			-27	-41
ZH62XA-T51	—	5.67	0.96	1.86	2	20	Aver.	6.5	21.5	35.7	Aver.	7.0	26.1	40.1
							Range	2.0	20.0	29.1	Range	4.0	25.0	36.5
								-12.0	-23.5	-39.5		-10.0	-27.6	-42.7

*Results represent 3 - 4 castings or 12 - 16 test bars.
**Heat Treatment: AZ92A-T6 18 hr at 770 F Air Cool. 5 hr at 420 F
ZK51A-T5 24 hr at 390 F
ZH62XA-T51 2 hr at 625 F Air Cool. 16 hr at 350 F

TABLE 7 — COMPARISON OF TENSILE PROPERTIES AT ROOM TEMPERATURE OF BARS SECTIONED FROM CASTING A WITH SEPARATELY-CAST TEST-BARS

	Composition, %				Casting A					Separately-Cast Bars				
Alloy	Th	Zn	Sol. Zr	Heat Treatment	No. Bars		% E	YS	TS	No. Bars		% E	YS	TS
ZK51A-T5	—	5.2	0.85	16 hr at 350 F	42	Aver.	5.1	23.8	35.9	3	Aver.	6.0	25.7	39.1
						Range	2.0	21.0	28.9		Range	5.0	25.6	38.0
							-10.0	-25.4	-41.6			-8.0	-25.9	-41.1
ZK51A-T51	—	5.2	0.85	2 hr at 625 F Air cool. 16 hr at 350 F	42	Aver.	6.7	23.6	37.2	3	Aver.	5.3	26.7	40.3
						Range	1.0	19.1	29.1		Range	5.0	25.9	39.6
							-13.0	-24.7	-41.7			-5.5	-27.5	-41.0
ZK61A-T62	—	5.8	0.87	2 hr at 930 F. Air cool. 48 hr at 265 F.	83*	Aver.	6.0	23.9	36.3	12	Aver.	9	30.4	45.3
						Range	2	18.2	29.2		Range	5.5	28.5	43.7
							-13	-28.4	-45.0			-11	-32.0	-46.8
ZH62XA-T5	1.6	6.2	0.90	24 hr at 300 F.	42	Aver.	5.8	24.2	37.9	3	Aver.	5.2	28.2	39.9
						Range	2.5	22.6	33.8		Range	4.0	27.7	38.8
							-10.0	-25.9	-40.6			-7.0	-28.7	-41.3
ZH62XA-T51	1.6	6.2	0.90	2 hr at 625 F. Air cool. 16 hr at 350 F.	42	Aver.	6.8	24.6	38.1	3	Aver.	5.0	27.7	40.0
						Range	2.0	21.9	32.5		Range	3.0	27.5	38.4
							-11.0	-26.4	-41.3			-6.5	-28.1	-41.3
ZH62XA-T5	2.0	5.8	0.90	24 hr at 300 F.	42	Aver.	8.0	24.1	38.5	12	Aver.	5.7	27.6	39.8
						Range	3	22.6	32.8		Range	4.9	26.8	37.5
							-15	-25.5	-41.0			-9.5	-28.8	-41.9

% E = Elongation in 2 in.
YS = Yield strength, 1,000 psi
TS = Tensile strength, 1,000 psi
*These results are from two castings, others from one casting.

in bars from the ZH62XA castings, it is not enough to impair seriously their tensile properties. It is concluded that microporosity in ZK51A and ZK61A alloys will drastically lower their tensile properties. On the basis of these results it is evident that if sound ZK51A and ZK61A alloy castings cannot be produced, optimum properties are best obtained using ZH62XA alloy.

Grain size determinations on bars machined from various locations in casting A show the following average ratings:

ZK51A — 0.002 to 0.005 in.
ZK61A — 0.002 to 0.004 in.
ZH62XA — 0.002 to 0.007 in.

Presence of small oxide skins and microporosity made it difficult to determine any effects of average grain size on tensile properties.

Effect of Testing Temperature on the Tensile Properties of ZH62XA-T5—ZH62XA-T5 alloy at three levels of thorium content was tested in tension at room and elevated temperatures. The results in Table 8 show that at testing temperatures from 300 F to 500 F no differences in properties are detected between 1.4 and 2.3 per cent thorium.

Figure 3 compares the room and elevated temperature properties of the ZH62XA alloy containing 1.8 per cent thorium with data previously reported for AZ92A-T6 and ZK51A-T5.[1] While none of these alloys possess a decided advantage in tensile properties in any respect, ZH62XA-T5 appears slightly superior up to 500 F.

Creep Strength of ZH62XA-T5—Total extension is

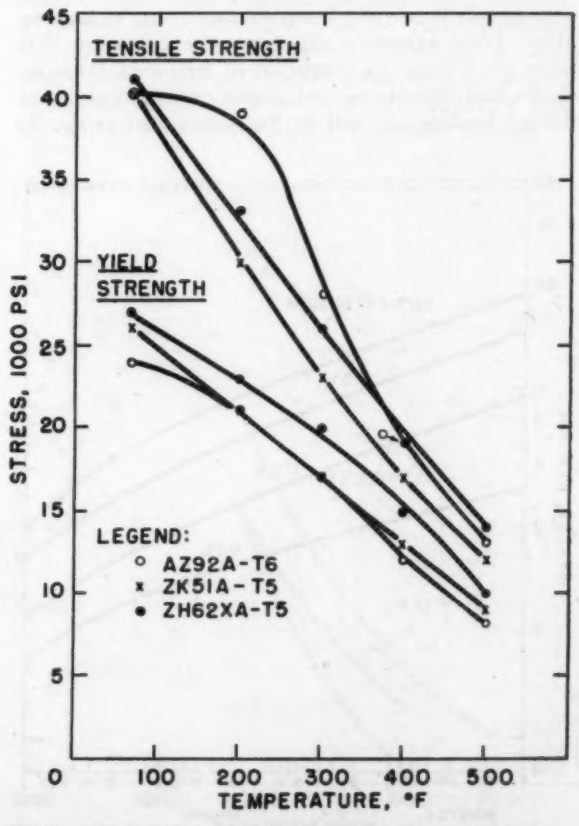


Fig. 3 — Comparison of tensile properties of several Mg-base sand casting alloys at room and at elevated temperatures. Separately-cast test bars.

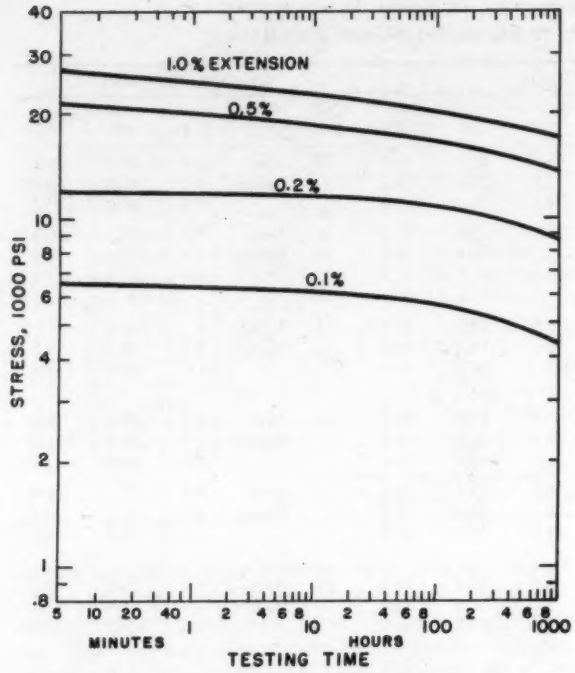


Fig. 4 — Stress for various total extensions at 200 F. Separately-cast test bars, ZH62XA-T5 alloy.

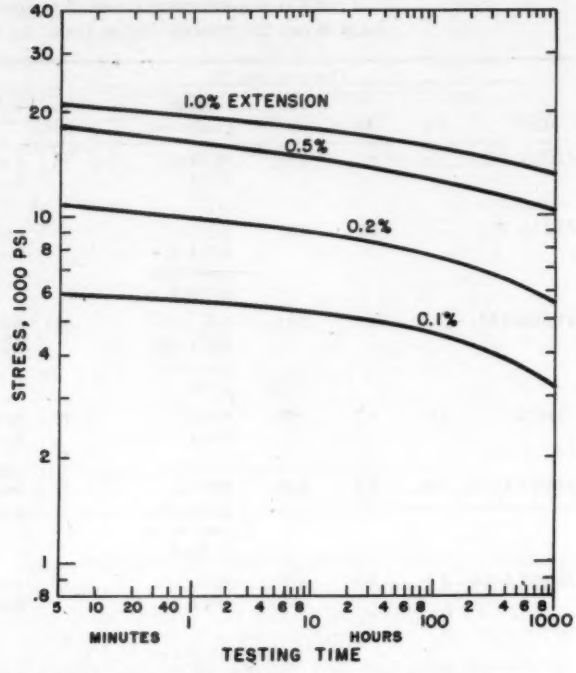


Fig. 5 — Stress for various total extensions at 300 F. Separately-cast test bars, ZH62XA-T5 alloy.

considered the most significant parameter for design purposes in presenting creep characteristics of a given alloy. Total extension represents the extension that takes place from the condition of zero load. This includes both the elastic and plastic extension obtained during loading, as well as the subsequent creep. In Figs. 4 through 7, the stresses necessary to produce 0.1, 0.2, 0.5 and 1.0 per cent total extension are plotted for testing times from 0 to 1,000 hr at four temperatures. The 0-hr points were derived from the autographic stress-strain curves obtained in the short-time tension tests.

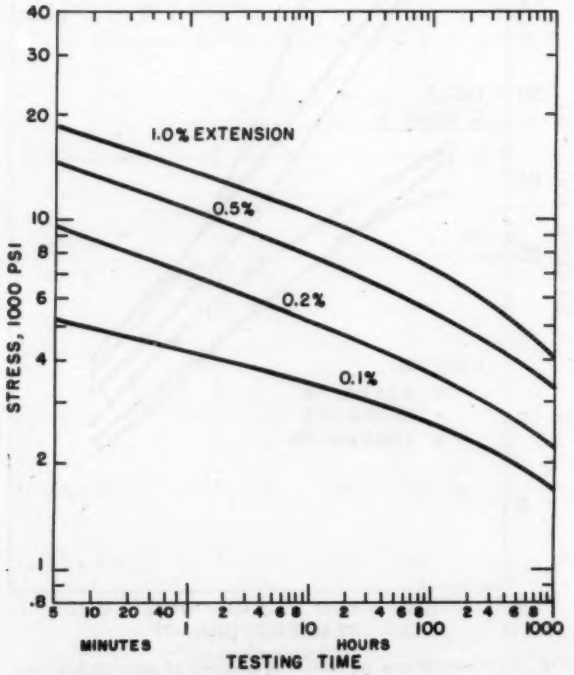


Fig. 6 — Stress for various total extensions at 400 F. Separately-cast test bars, ZH62XA-T5 alloy.

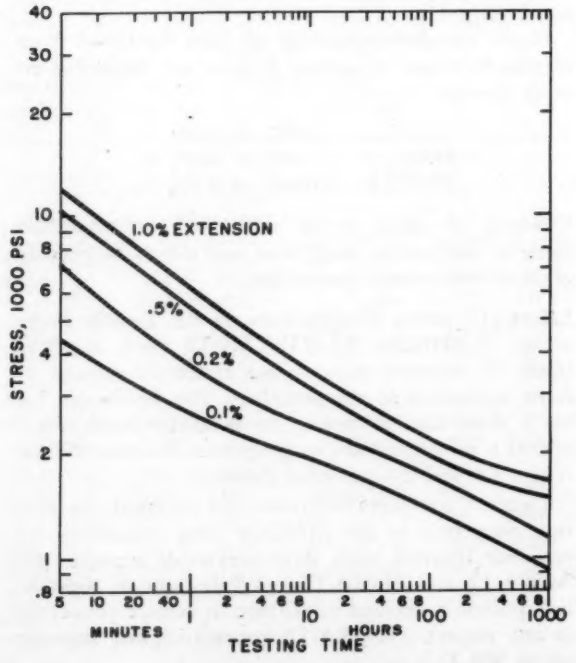


Fig. 7 — Stress for various total extensions at 500 F. Separately-cast test bars, ZH62XA-T5 alloy.

Table 8 — Effect of Thorium Content on Tensile Properties of ZH62XA-T5* Alloy at Elevated Temperature
Separately-Cast Test-Bars

Composition, %			Testing	No.	Average Tensile Properties			
		Sol.	Temp.,	of				Grain
Th	Zn	Zr	F	Bars	% E	YS	TS	Size, in.
1.4	5.7	0.93	70	3	5	26	39	0.002
			200	1	19	23	33	0.002
			300	1	23	20	26	0.002
			400	1	26	15	18	0.002
			500	1	29	10	13	0.002
1.8	5.8	0.98	70	3	10	27	41	0.002
			200	1	20	23	33	0.0015
			300	1	24	20	26	0.0015
			400	1	28	15	19	0.0015
			500	1	30	10	14	0.0015
2.3	5.7	0.99	70	3	6	24	38	0.002
			200	1	11	21	32	0.002
			300	1	22	19	26	0.002
			400	1	28	15	19	0.002
			500	1	32	10	13	0.002

* T5 treatment — 24 hr at 300 F

% E = Elongation in 2 in.
YS = Yield strength, 1,000 psi
TS = Tensile strength, 1,000 psi

The stresses required to obtain 0.2 per cent total extension for three different testing times are compared for alloys ZK51A-T5, AZ92A-T6 and ZH62XA-T5 in Figs. 8 through 10. The ZK51A and AZ92A data are from an earlier paper.[1] ZH62XA-T5 has better resistance to creep than AZ92A-T6 at all testing temperatures and is better than ZK51A-T5 up to about 400 F.

It is not intended that ZH62XA-T5 alloy be considered as a magnesium alloy for use at elevated temperatures. Rather, it should be considered as a high-strength alloy at room temperature with useful properties up to about 300 F. Above this temperature, if applications require resistance to creep at stress levels above 5,000 psi, consideration should be given to the Mg-RE-Zr or to the Mg-Th-Zr alloys. A comparison of these alloys with ZH62XA-T5 follows:

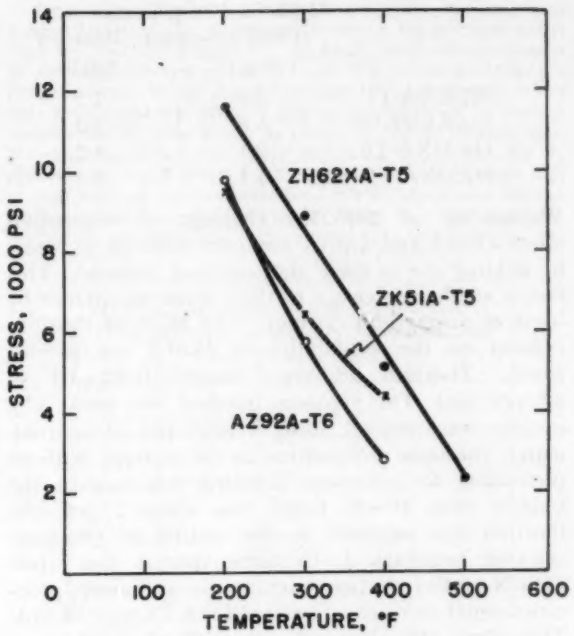


Fig. 8 — Comparison of 10-hr creep limits (0.2 per cent total extension) of several Mg-base sand casting alloys.

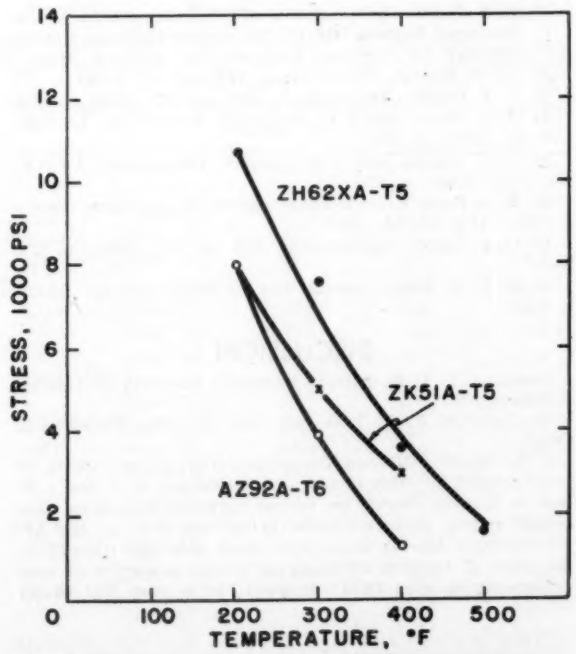


Fig. 9 — Comparison of 100-hr creep limits (0.2 per cent total extension) of several Mg-base sand casting alloys.

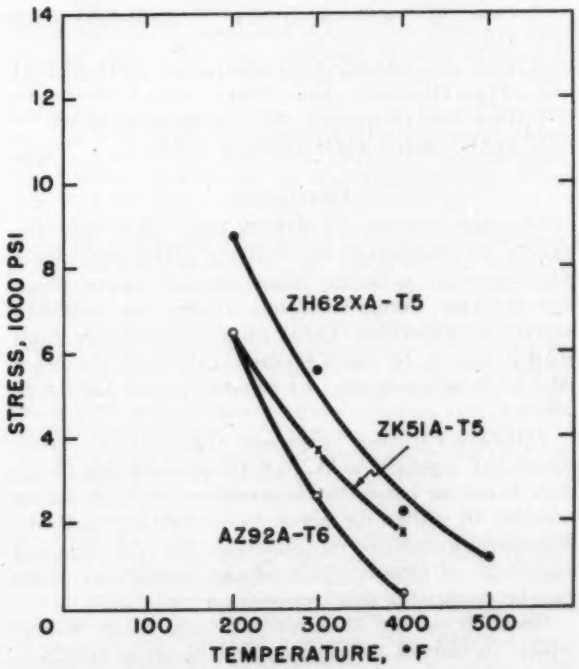


Fig. 10 — Comparison of 1000-hr creep limits (2.0 per cent total extension) of several Mg-base sand casting alloys.

	1,000-hr Creep Limits, 0.2% Total Extension, 1000 psi	
	400 F	500 F
ZH62XA-T5	2.3	1.2
Mg-RE-Zr	5.8	2.0
HK31XA-T6	9.1	4.2
HZ32XA-T5	7.4	5.0

Weldability of ZH62XA—Castings of magnesium alloys ZK51A and ZK61A are quite difficult to repair by welding due to their high cracking tendency. This fact is another deterrent to their wider acceptance by users of magnesium castings. The effect of thorium content on the weldability of ZK61A was investigated. Thorium additions ranged from 1.4 to 3.0 per cent. The argon-arc method was used. The castings were welded, using a filler rod of approximately the same composition as the castings, without preheating to accentuate cracking tendency in the welded areas. It was found that about 2 per cent thorium was required in the casting to eliminate cracking problems. It therefore appears that while ZH62XA alloy castings have tensile and creep properties equal to or superior to ZK51A-T5 and ZK61A-T62 alloys, this alloy not only is much superior in castability but also it can be repair welded more readily.

Surface Protection—The advantages of chemical treating and painting magnesium alloys are well known. The protective treatment applied, of course, will depend upon the application in which the casting will be used. The following treatments can be satisfactorily applied to ZH62XA-T5 alloy:

1. Dow No. 1 chrome pickle (MIL-M-3171 Type I).
2. Dow No. 7 dichromate treatment (MIL-M-3171 Type III).
3. Dow No. 9 galvanic anodize (MIL-M-3171 Type IV).
4. Dow No. 10 sealed chrome pickle (MIL-M-3171 Type II).
5. Dow No. 17 anodize (No specification issued).
6. HAE anodize (MIL-C-13335, ORD).

Conclusion

The introduction of ZH62XA-T5 alloy into the family of magnesium sand-casting alloys provides a high strength structural alloy for use up to about 300 F. This composition has a very low tendency toward microporosity. Other castability considerations dictate that it be rated in castability with the other Mg-Th-Zr alloys below the Mg-Al-Zn and Mg-RE-Zr alloys.

ZH62XA-T5 alloy possesses resistance to creep somewhat superior to AZ92A-T6 and ZK51A-T5 alloys. It can be joined by the usual methods, including welding by either the argon-arc or heli-arc methods. Standard procedures recommended for the chemical treatment of commercial sand-cast magnesium alloys can be used with this composition.

Selection of the thorium and zinc range within which to control the composition of alloy ZH62XA-T5 is based upon many compromises. The optimum combination in castability, tensile properties and weldability is obtained at a sacrifice of one or more of these properties. The range in analysis and specified tensile properties shown in Table 9 have been proposed for this alloy.

TABLE 9 — PROPOSED ANALYTICAL AND PROPERTY CONTROLS FOR ZH62XA-T5 ALLOY

Heat Treatment: 24 hr at 300 F.		
Composition Range, %:		
Thorium	1.4 – 2.2	
Zinc	5.2 – 6.2	
Zirconium	0.50 minimum	
Tensile Properties, Separately-cast test-bars:	Minimum	Typical
Elongation, % in 2 in.	4	8
Yield strength, 1,000 psi	22	26
Tensile strength, 1,000 psi	35	40
Tensile Properties, Average of 4 - 10 Bars from Castings:		
Elongation, % in 2 in.	2	
Yield strength, 1,000 psi	18	
Tensile strength, 1,000 psi	27	

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DISCUSSION

Chairman: E. V. BLACKMUN, Aluminum Company of America, Pittsburgh, Pa.

Co-Chairman: F. L. TURK, Joy Mfg. Co., New Philadelphia, Ohio.

J. W. MEIER[1]: *(Written Discussion):* The authors are to be congratulated for their interesting contribution of a new addition to research reports on various zirconium-containing magnesium casting alloys published in the last years in the AFS TRANSACTIONS. Having investigated, some time ago (1950-1952), the effect of thorium additions on various properties of magnesium casting alloy ZK61*, I would like to state that we ob-

1. Department of Mines and Technical Surveys, Ottawa, Ont., Canada.

*J. W. Meier, "The Effect of Thorium Additions on the Properties of Magnesium Casting Alloy ZK61," Research Report PM205, Mines Branch, Department of Mines and Technical Surveys, Ottawa, Canada.

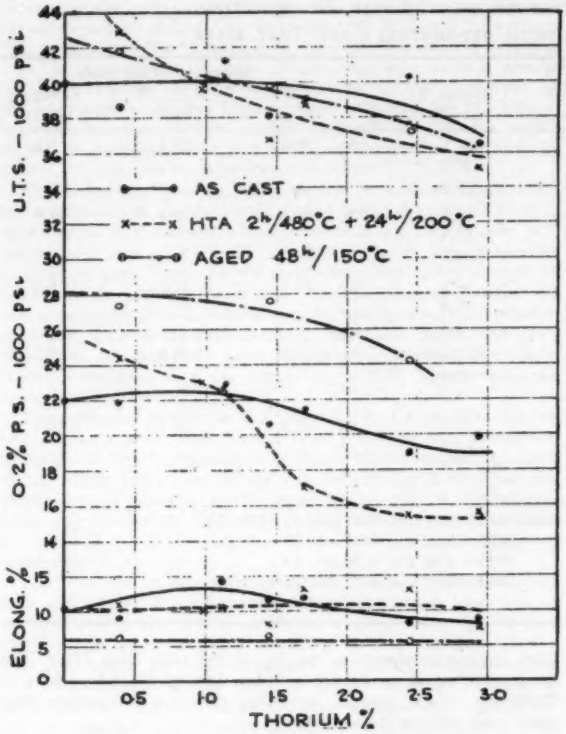


Fig. A — Effect of thorium addition on mechanical properties of Alloy ZK61.

tained in our study very similar results to those presented by the authors. Our data on tensile properties at room (Fig. A) and elevated temperatures (Table A), creep properties, etc., agree very well with the data presented in the paper.

TABLE A — EFFECTS OF THORIUM ADDITIONS ON ROOM AND ELEVATED TEMPERATURE TENSILE PROPERTIES OF MAGNESIUM CASTING ALLOY ZK61

ZK61 (Melt 3226) — 5.82% Zn, 0.77% sol. Zr, 0.08% insol. Zr
ZH61 (Melt 3326) — 5.78% Zn, 0.78% sol. Zr, 0.08% insol. Zr, 1.05% Th
ZH62 (Melt 3314) — 5.68% Zn, 0.70% sol. Zr, 0.11% insol. Zr, 2.20% Th

	Alloy Designation*	UTS kpsi	0.2% PS kpsi	Elong. % in 2 in.
Room Temperature	ZK61-T6	47.0	32.5	11.0
	ZH61-T6	41.5	24.4	10.0
	ZH62-T6	41.2	26.5	8.5
200 F	ZK61-T6	39.5	27.5	24.0
	ZH61-T6	34.3	24.9	22.0
	ZH62-T6	34.5	23.9	19.0
300 F	ZK61-T6	28.4	22.9	27.0
	ZH61-T6	27.0	20.8	28.5
	ZH62-T6	27.5	20.3	26.0
400 F	ZK61-T6	19.5	15.9	32.0
	ZH61-T6	19.8	17.0	31.5
	ZH62-T6	19.6	15.6	41.5
500 F	ZK61-T6	10.9	8.7	42.0
	ZH61-T6	14.2	10.9	45.0
	ZH62-T6	15.2	11.8	44.5
600 F	ZK61-T6	6.0	4.8	55.0
	ZH61-T6	10.0	7.4	38.0
	ZH62-T6	11.4	9.0	35.0

*ZK61-T6 — fully heat treated (2 hr at 500 C, 48 hr at 130 C). ZH61-T6 and ZH62-T6 — heat treated (2 hr at 330 C, 16 hr at 180 C).

The authors showed in Table 3 the effect of some heat treating cycles on the properties of various Mg-Zn-Th-Zr alloys (not Mg-Th-Zn-Zr, as stated in the title). Their disappointment in the ineffectiveness of the T4 and T6 treatments (using a temperature of 960 F) could have been avoided if they would have determined the effect of thorium additions on the solidus temperature of these alloys. As shown in Fig. B, thorium lowers this temperature for alloys containing 6 per cent Zn by approximately 58 F (from 986 F to 928 F) so that it seems that at least some of the bars were heat treated above their solidus temperature. The rather low results shown in Table 3 for the T5 temper are due to the high aging temperature used (400 F).

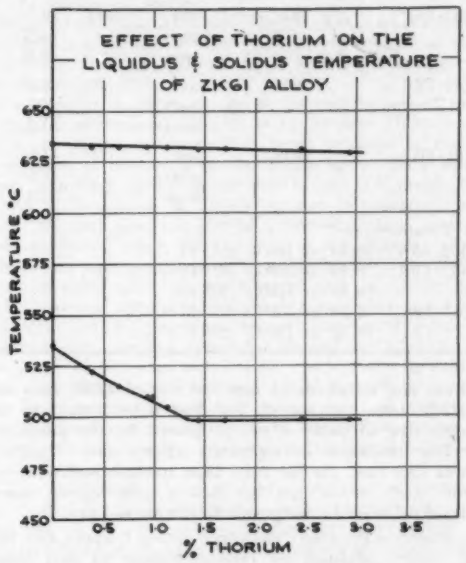


Fig. B — Effect of thorium on liquidus and solidus temperature of Alloy ZK61.

Unfortunately, I have also to disagree. My disagreement with the authors has nothing to do with their factual data on properties of alloy ZH62, and I agree in general with their evaluation of the foundry properties of this alloy. But I have to disagree with their misleading comparison of alloy ZH62 with magnesium casting alloy ZK61, which is based on the evaluation of obviously defective ZK61 castings. Although it is known[3] that the handling of this latter alloy in the foundry requires somewhat more care as compared with other magnesium casting alloys, it is also known[3] that sound castings in this alloy have been produced commercially for many years, and are being accepted to very strict aircraft specifications.

The authors state that the slope and airframe castings were all produced using the same gating and risering as for alloy AZ91C, and then just mention that "due to the high incidence of microporosity in the ZK61 alloy airframe castings, properties of bars sectioned from the castings are inferior". Any foundryman knows that to obtain sound castings one must use proper casting conditions, adapted to the characteristics of the alloy to be used. And, any metallurgist knows that a correct evaluation of the effect of alloy composition on the properties of castings can be made only on standard quality castings, that is, commercially saleable castings.

By using a standard gating and risering system for alloys of widely differing casting characteristics, the authors could find out only if this gating and risering was or was not suitable for the particular alloys. It would be difficult to find a foundryman who would use a gating and risering system designed for an aluminium-silicon alloy for the production of a casting in the aluminum—10 per cent magnesium alloy (G10, 220) or, similarly, an 85-5-5-5 bronze gating would never do for manganese or aluminum bronzes. Alloy ZK61 is similarly a difficult alloy just as aluminum bronze or aluminum alloy G10 and has, therefore, to be handled with special care.

The authors agree that the ZK61 alloy castings were defective, but still used them in their investigation. We had tested

TABLE B — COMPARISON OF TENSILE PROPERTIES AT ROOM TEMPERATURE OF TEST BARS CUT FROM CASTING "X" WITH SEPARATELY-CAST TEST BARS

Alloy	Melt No.			Casting "X" UTS kpsi	Casting "X" 0.2% PS kpsi	Casting "X" Elong. % in 4D		Separately-Cast Bars UTS kpsi	Separately-Cast Bars 0.2% PS kpsi	Separately-Cast Bars Elong. % in 2 in.
ZH62-T5A	3815	ave.	(20)	39.0	23.5	11.1	(2)	39.6	25.1	9.2
		max.		41.2	24.7	20.0				
		min.		34.8	22.0	2.5				
ZH62-T5B	3815	ave.	(20)	39.4	25.9	11.5	(2)	42.1	28.1	11.0
		max.		41.9	28.9	17.5				
		min.		32.8	23.2	5.5				
ZK61-T6	3817	ave.	(20)	39.3	29.1	6.5	(2)	45.6	31.5	6.2
		max.		46.5	34.6	16.0				
		min.		30.6	22.8	2.5				
ZK61-T6	2699	ave.	(70)	39.0	26.3	6.0	(2)	47.8	33.0	9.5
		max.		46.2	31.1	13.0				
		min.		31.3	22.9	3.0				
ZK61-T6	2049	ave.	(17)	41.6	27.9	7.0	(6)	47.5	32.4	10.2
		max.		46.0	30.9	13.5				
		min.		34.5	26.3	3.0				

Heat Treatment:
ZH62-T5A 16 hr at 180 C (355 F)
ZH62-T5B 2 hr at 330 C (625 F) + 16 hr at 180 C (355 F)
ZK61-T6 2 hr at 500 C (530 F) + 48 hr at 130 C (265 F)

Composition:
Melt 3815 — 5.61 Zn, 0.85 sol. Zr, 1.90 Th
3817 — 5.92 Zn, 0.73 sol. Zr
2699 — 5.91 Zn, 0.75 sol. Zr
2049 — 6.02 Zn, 0.76 sol. Zr

hundreds and hundreds of bars cut out of ZK61 alloy castings of various sizes and shapes, but had never found so low an average value of proof stress, as quoted by the authors. This is because moderate microporosity affects only slightly proof stress of 18.2 kpsi (in the fully heat treated condition) can be obtained only if the test bar had a gross defect, something which would never be acceptable in commercial practice.

To illustrate the effect of proper gating I would like to show a few results obtained on castings similar to that shown in Fig. C. The results of tensile tests on bars cut out from these castings are presented in Table B and compared with separately-cast test bars which were used to check the melt quality. The first two castings, Melts 3815 and 3817, one in alloy ZH62, the other in ZK61, were gated similarly, the other two castings, both in ZK61, were gated and risered differently in each case.

Although all of the castings show very good properties it is quite obvious that the last casting (Melt 2049) was produced under the most favorable conditions. The comparison of mechanical properties of this casting with those of the ZH62 alloy casting, as shown at the top of the table (the ZH62 alloy results are equal or better to those quoted by the authors in Table 7), would lead to somewhat different conclusions than those presented by the authors.

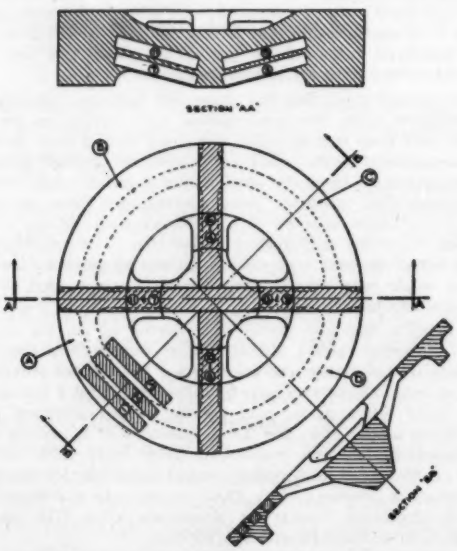


Fig. C — Location of test coupons in ZK61 alloy casting.

But, "castability" is not just the incidence of microporosity, which, I agree, is more pronounced in alloys ZK51 or ZK61 than in ZH62. The latter alloy, on the other hand, is much more "drossy" and therefore less fluid, has a much higher shrinkage, tends to "surface sinks", and is at least as susceptible to hot cracking as the two other Mg-Zn-Zr casting alloys.** To speak, therefore, of a "much superior" castability is somewhat exaggerated.

The advantages of alloy ZH62 are obviously due to the fact that it is weldable and less susceptible to microporosity. Otherwise, ZH62 should be listed in the category of rather "difficult" casting alloys of definitely lower tensile properties than the Mg-Zn-Zr alloys.

MESSRS. NELSON AND SAUNDERS *(Reply To Mr. Meier):* The authors are grateful for the discussion submitted by Mr. Meier and are happy to have their work on ZH62XA so nearly in agreement with his results on the same alloy. We are in complete agreement as to the basic findings, namely that the advantages of ZH62XA alloy over ZK51A and ZK61A alloys lie in its weldability and soundness.

The point brought out regarding the solidus of Mg-Zn-Th-Zr alloys reported in Table 3 of the paper is well taken. An error was made in the selection of a solution heat treating temperature. However, a recent metallographic examination of some of the test bars did not indicate severe melting. We have also recently rechecked the non-equilibrium solidus of an alloy containing 5.95% Zn, 1.59% Th and 0.91% Soluble Zr. This was also done by metallographic examination. The non-equilibrium solidus of this composition appears to be between 960 and 970 F. This confirms Mr. Meier's suggestion that at least some of the solution heat treatments were carried out at too high a temperature. This solidus temperature does not agree with that of 928 F as reported by Mr. Meier in Fig. B.

We must disagree with Mr. Meier's philosophy relative to the proper selection of an evaluation technique. In our evaluation of properties we have shown that, if made sound, ZK61A alloy has very good tensile properties. However, in both our research and production foundries we have had considerable difficulty in producing sound ZK61A alloy castings. Variations

**R. A. Dodd, "Hot Tearing and Hot Cracking of Some Industrial Magnesium Alloys," Research Report PM197, Mines Branch, Department of Mines & Technical Surveys, Ottawa, Canada (February 10, 1956).

in casting size and complexity do, of course, influence the ease at which sound castings can be produced. It is our experience that most of the ZK61A castings we have poured contain porosity.

D. J. WHITEHEAD[2] *(Written Discussion):* The authors are to be congratulated on their timely survey of the properties of the high zinc containing high strength alloys of the Mg-Th-Zn-Zr series. Interest in TZ6 (ZH62XA) alloy for airborne applications is increasing and will no doubt be further fostered by Messrs. Nelson and Saunders' contribution.

In Europe the position with regard to high strength magnesium-zirconium casting alloys differs from what might be surmised from the authors' abstract in that we have made very considerable use of Z5Z (ZK51) alloy for a variety of aircraft castings ranging from control levers weighing a few ounces to undercarriage castings of several hundred pounds weight. In our experience, Z5Z is less subject to microporosity in practice than Mg-Al-Zn alloys such as AZ92, AZ63 and AZ81, and high properties are realisable from the largest castings. We would therefore tend to rate the castability of Z5Z higher than the authors have seen fit to do.

The principal defects of Z5Z alloy lie in its susceptibility to hot cracking and poor weldability, although the former is probably of lesser importance in England due to the tendency to use softer cores than is the practice in the U.S.A. For the production of castings where these factors are of critical importance, rare earth metals have been added to Z5Z to form the alloy RZ5 (ZE41 A-T5). This alloy is also in quantity production in Europe.

The addition of rare earth metals to Mg-Zn-Zr alloys is however not without its disadvantages in that there is some loss of strength and ductility. The addition of thorium instead of rare earth metals largely avoids this and, as the authors have shown, TZ6 alloy offers the best compromise for an optimum combination of castability, tensile strength and weldability. Thorium is an expensive addition, however, and consequently TZ6 alloy has not replaced Z5Z alloy in Europe but has so far been used only for those applications where full advantage can be derived from its properties.

We are largely in agreement with the authors' findings on the castability characteristics of TZ6 alloy. The incidence of microporosity is of a very low order at a full zirconium content, but in common with the other thorium and rare earth metal containing alloys, greater attention to gating technique and pouring conditions is necessary to avoid oxide inclusions.

The occurrence of "swirl porosity" in some Mg-Zn-Zr alloys which the authors illustrate is influenced by grain size. With coarse grained material occurring as a result of low soluble zirconium content, swirl porosity is replaced by heavy layer type porosity of the form characteristic to Mg-Al-Zn alloys. It is also possible to produce swirl porosity in certain non-zirconium alloys if a sufficiently fine grain size is achieved.

With regard to the phenomenon of cope pitting, we are rather at a loss to comment on it in view of the fact that it appears to be unknown in England. We are inclined to think that it may be a turbulence effect and have its origin in fundamental differences in the running and gating techniques applied to magnesium sand castings in our respective countries.

Our conclusions on the heat treatment of TZ6 differ somewhat from those of the authors. For single stage heat treatments we find that maximum tensile properties are developed after aging for 24 hours at 480 F but that rather better properties are, on average, developed by a double heat treatment of 2 hours at 625 F followed by 16 hours at 355 F. We have observed a secondary peak at 300 F in the aging temperature—yield strength curve for 24-hour heat treatments, but find that equivalent yield strength and better ultimate tensile strength are obtained after heat treatment at 480 F. We accordingly advocate use of the double heat treatment. We agree that there is a possibility that lowering the temperature of the second step to 300 F and increasing the time may give slightly better properties. A temperature of 355 F is normal however for heat treating Z5Z alloy, and the gain in properties would have to be appreciable to justify the change in Europe.

MESSRS. NELSON AND SAUNDERS *(Reply To Mr. Whitehead):* The authors are pleased to hear from a representative of the company which has had so much to do with the development of the Mg-Zr type of alloy.

The only point in which we are at variance in respect to this alloy and its operational controls is in its heat treatment. As stated in the paper, we observed a better response to the two-step treatment as compared to the simple age at 300 F only in test bars analyzing at the upper end of the Th range. However, even at these levels of Th the increase in tensile properties is small. We feel that the use of a two-stage heat treatment involves additional effort on the part of the foundry without enough of an increase in properties to justify this effort. If problems of stress-relief are encountered and it can be shown that the two-step treatment produces appreciably more relief than does the simple aging treatment, or should other advantages be made known, such a treatment should be recommended.

The listing of alloys as to castability, as we have said before, becomes the prerogative of each individual foundry. In our own case we find complicated shapes in ZK51A (Z5Z) alloy very difficult to produce free of microporosity as detected by radiography. The sensitivity of tensile properties to this porosity has been confirmed in our paper.

We are, of course, aware that the addition of rare earth metals to ZK51A (Z5Z) produces a more castable alloy. This composition, RZ5, (recently accepted by the ASTM and designated as ZE41A, using the ASTM system of nomenclature) has not been produced in any quantity in our foundry. As Mr. Whitehead points out, there is some loss of strength and ductility in ZE41A compared to ZK51A alloy. It is our feeling that, not only is ZH62A comparable in castability and weldability to ZE41A alloy, but, in addition, exhibits higher tensile properties than ZE41A and properties equivalent to those of ZK51A. The increased cost of the Th in ZH62A is somewhat offset by the difficulty in maintaining close analytical controls on ZE41A alloy.

J. G. MEZOFF[3]: Would the authors make a comparison of the castability of ZK61 and A1-10Mg alloys?

MESSRS. NELSON AND SAUNDERS *(Authors' Closure):* The composition discussed in this paper is the basis for a patent recently issued to Magnesium Elektron Limited (U.S. Patent No. 2,750,288 dated June 12, 1956). It has also been accepted by the ASTM and is listed in ASTM specification B80-56T. Alloy ZH62A is a good compromise for a composition possessing an optimum combination of castability, tensile strength and weldability.

2. Magnesium Elektron Ltd., Manchester, England.

3. Dow Chemical Co., Midland, Mich.

EFFECTS OF SECTION SIZE VARIATIONS IN A TEST CASTING ON PROPERTIES OF SOME Mg-Al-Zn ALLOYS

By

W. E. Pearson*

ABSTRACT

A test casting termed the "slope casting" which possesses a range of section sizes has been developed to provide tensile properties equivalent to those obtained from typ.cal production sand castings. The value of this laboratory casting is illustrated by using it in an investigation of a group of Mg-Al-Zn alloys. The alloys studied in the T4 condition include a series of 16 containing 5 to 9 per cent Al, 0.0 to 1.0 per cent Zn and 0.2 per cent Mn. Six other compositions tested in the as-cast temper include 5 and 6 per cent Al + 0.4 per cent Zn + 0.2 per cent Mn and the commercial alloys AZ81XA, AZ63A, AZ92A and AM100A. The results indicate that an alloy designated AZ61 (Mg + 6.2 Al + 0.7 Zn + 0.2 Mn) has a favorable combination of tensile properties in castings in both the T4 and as-cast conditions. The slope casting is shown to be useful in alloy development, heat treatability, grain size and porosity tendency studies.

Introduction

Mechanical properties of magnesium-base casting alloys are ordinarily determined on standard ½-in. diameter cast-to-shape tensile test bars. In the later stages of alloy development, however, the need for more extensive tensile data is often recognized. To fill this need, production-type castings are sometimes sectioned and tested. This introduces the section size variable which is lacking when separately-cast bars only are tested.

In our laboratory we have recognized the need for a special type of test casting which would incorporate the section size variable and yield a range of tensile properties equivalent to those found in typical production castings. The "slope casting" was designed to meet these requirements. We have found it useful in alloy development studies for evaluating mechanical properties, response to heat treatment, grain refinement and to some extent porosity tendencies and castability. The object of this paper is to describe this laboratory test casting and illustrate its use during some investigations of alloys in the Mg-Al-Zn system.

Properties of commercial Mg-Al-Zn alloys, including data from actual castings, have been reported.[1] Fox[2] in England and Busk and Marande[3] in this country have made intensive investigations on the Mg-rich corner of the Mg-Al-Zn system with mechanical properties being determined on separately-cast test bars. A detailed comparison of U.S. and European alloys has also been made.[4]

Many investigations into the relationships between casting section size and mechanical properties have been conducted on cast irons including spheroidal graphite iron.[5-14] These workers employed test bars of several diameters plus Y-blocks and keel blocks of different sizes.

Ruddle[15] has demonstrated the correlation between temperature gradients and tensile strength of an Al-Cu alloy by casting plates of several thicknesses. A "step bar casting" was used by Reichenecker[16] to simulate section sizes of ⅜ in. to 2 in. He measured the mechanical properties of six Cu-base alloys and one aluminum composition with bars machined from this casting.

In the field of magnesium-base alloys, however, most of the reported mechanical properties have been determined on separately-cast test bars or bars machined from actual production castings.

Description of the Slope Casting

A test casting designed to provide a range of tensile properties must necessarily include a range of section



Fig. 1 — Drag and cope views of slope castings.

*Metallurgical Laboratory, Magnesium Department, The Dow Chemical Co., Midland, Mich.

sizes, or, more specifically, variations in solidification rate. A spread in tensile properties is then created by the effects of cooling rates on such factors as grain size and compound distribution.

The slope casting is an adaptation of the well-known step casting design, the main difference being a smooth and gradual increase in section thickness rather than a few abrupt changes. Its general shape and size can be seen in Fig. 1 which shows drag and cope views with rigging attached. Significant dimensions are included in Fig. 2. The casting is of such a size that green-sand molds can be conveniently produced on a bench molding machine. The enlarged riser is cut into the cope mold by hand or could be formed by a simple loose piece on the pattern.

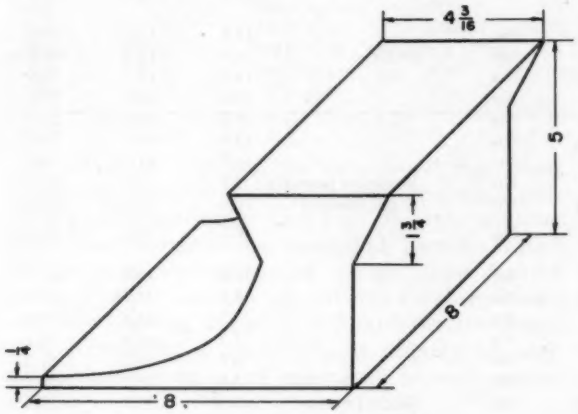


Fig. 2 — Sketch of slope casting.

The minimum thickness at the thin edge of the casting is ¼ in. This increases at an increasing rate until the slope curve opens into the riser. This design insures strong directional solidification and tends to produce the maximum mechanical properties in the thinnest section. The volume/area ratio of the thinnest section is about the same as a ¼-in. thick plate which should result in a cooling rate about equivalent to that of a ½-in. diameter separately-cast test bar. The heaviest part of the casting may be considered to cool similarly to a 3 in. diameter cylinder under a riser or to a 1½-in. thick plate or a 4½-in. cube.

Figure 3 is a sketch showing the locations from which test bar blanks are cut. The slope casting can be sectioned into several horizontal levels, each about ¾ in. thick. Ordinarily, however, only the bottom slice, level 1, is used in obtaining tensile properties of Mg-Al-Zn alloys because considerable porosity is present in the upper levels. When alloys having a very low porosity tendency, such as EZ33A, which contains about 3 per cent rare earths, 2.7 per cent zinc and 0.7 per cent zirconium, are cast, the upper levels may also be sectioned to provide additional test bars. There are ten bars in level 1, numbered 1A through 1J, beginning with the thick section. The 1H, 1I and 1J sections are machined into 6⅞-in. long flat bars, ⅛ in. to ¼ in. thick, having a 2-in. gage length, and the other seven into ⅜-in. diameter rounds, 6½ in. long with a gage length of 1½ in. The dimensions of all the bars conform to sizes acceptable to A.S.T.M.

Entire castings may be heat treated, but in most cases level 1 alone is given the desired heat treatment and then cut into blanks for the individual bars. Little difference is observed in the rate of compound solution, in level 1 at least, whether the whole casting or a small section is heat treated at one time.

The relative soundness of slope castings of various alloys can be determined by radiography, metallography or fracturing. Two sections are usually radiographed: level 1 and ¾-in. thick vertical slice through the center of the upper levels. Test bars are also often radiographed after machining.

In most cases metallographic data are obtained on samples cut from the reduced sections of machined bars after tensile testing. However, data on grain sizes and compound ratings may also be gathered on samples from a ½-in. thick strip cut from the side of level 1 and containing sections 1A to 1J.

Procedure

The slope castings and separately-cast test bars prepared for this investigation were poured into green-sand molds from Mg-Al-Zn alloy melts of about 45 lb. These melts were made with virgin materials and the melting was done in gas-fired settings. Chlorine was used for degassing and the metals were super-heated for grain refinement. The pouring temperature was about 1450 F for all melts.

Designations for the two tempers studied in this investigation are as follows:

F - as-cast

T4 - solution heat treated (details in tables).

Solution heat treatments were performed in circulating air electric resistance furnaces with a protective atmosphere of 1 per cent SO_2.

Sixteen different compositions were tested in the T4 temper. These included a series of 15 alloys whose nominal compositions were 5, 6, 7, 8 and 9 per cent aluminum respectively, with zero, 0.4 and 1.0 per cent zinc at each level of aluminum. The other alloy was 6.5 per cent aluminum plus 1.0 per cent zinc. All these alloys also contained about 0.2 per cent manganese. Some of the alloys fall within the composition ranges of the commercial casting alloys AZ81XA and AZ91C.

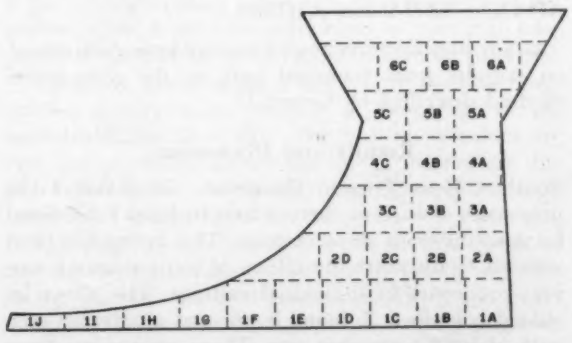


Fig. 3 — Location of bars cut from slope casting.

TABLE 1 — PROPERTIES OF INDIVIDUAL BARS MACHINED FROM SOLUTION HEAT TREATED SLOPE CASTINGS

Composition % Al	Composition % Zn	Heat Treatment	Metallography Bar No.	Aver. Grain Diam., in.	Compound Rating	Voids	Visual Defects at Fracture Type	Tensile Properties % E in 2 in.	Y.S. x 1000, psi	T.S. x 1000, psi
6.02	0.40	790 F, 16 hr	−1J	.005	0	No		17.0	11.8	36.3
			−1I	.005	0	No		16.5	11.8	36.0
			−1H	.005	0	No		17.5	11.2	35.9
			−1G	.006	1	No		16.6	11.2	35.9
			−1F	.008	1	No		14.7	10.8	34.7
			−1E	.008	1½	No		14.0	10.6	35.0
			−1D	.008	1½	No		14.0	11.3	35.1
			−1C	.007	1	No		16.0	11.1	35.4
			−1B	.007	1	No		9.3	10.9	29.9
			−1A	.006	1	No	Slight porosity	10.7	11.4	34.2
7.16	0.36	790 F, 16 hr	−1J	.004	0	No		13.5	12.4	37.8
			−1I	.004	0	No		13.5	12.1	37.6
			−1H	.005	0	No	Oxide	14.5	12.2	38.3
			−1G	.007	½	No		14.7	11.1	36.4
			−1F	.007	1	No		15.3	10.9	36.4
			−1E	.009	2	No		13.3	11.0	35.7
			−1D	.009	2½	No		12.0	11.8	34.8
			−1C	.009	2	No		14.0	11.2	35.3
			−1B	.008	2	No	Moderate porosity	9.3	11.2	32.4
			−1A	.008	1½	No	Slight porosity	10.0	11.7	34.2
8.03	0.36	790 F, 16 hr	−1J	.006	½	No		16.5	12.1	38.2
			−1I	.005	½	No		16.0	12.0	37.6
			−1H	.004	1	No		15.5	11.7	37.2
			−1G	.006	2	No		14.7	11.8	37.8
			−1F	.007	3	No		11.3	12.1	35.1
			−1E	.008	4	No		8.7	12.1	31.9
			−1D	.007	4	No		8.0	12.2	30.8
			−1C	.007	4½	Yes	Slight porosity	7.3	12.3	29.2
			−1B	.007	4	No	Moderate porosity	6.0	12.4	26.9
			−1A	.006	3½	No		8.7	12.1	32.5
9.05	0.41	790 F, 16 hr	−1J	.004	1½	No		15.0	14.9	41.0
			−1I	.004	1½	No		15.0	15.3	40.2
			−1H	.005	2	No		9.5	15.3	36.3
			−1G	.009	3	No		8.7	12.4	33.1
			−1F	.010	4	No		5.3	13.4	27.7
			−1E	.009	4½	No		3.3	12.2	23.2
			−1D	.010	5	No	Moderate porosity	1.3	13.3	19.6
			−1C	.010	5	Yes	Moderate porosity	1.3	12.4	19.7
			−1B	.010	4½	Yes	Heavy porosity	1.3	11.9	19.1
			−1A	.009	5	No		4.0	11.5	25.5

Testing in the as-cast condition was limited to six alloys. The nominal compositions of two of these were 5 and 6 per cent aluminum, each with 0.4 per cent zinc. The others were commercial alloys having the following A.S.T.M. designations: AZ63A, AZ81XA, AZ92A and AM100A.

Grain sizes and compound ratings were determined on samples from fractured bars by the comparison method described by George.[17]

Results and Discussion

Solution Heat Treated Condition — In Table 1 the properties of each of the ten bars in level 1 are listed for four different slope castings. This group has been selected to illustrate the effects of section size on various properties in individual castings. The alloys included contain 6, 7, 8 and 9 per cent aluminum each with about 0.4 per cent zinc. These castings are from the group of 16 tested in the solution heat treated condition.

As shown in Fig. 3, the 1J bar is machined from the thinnest part of the slope casting and then in reverse alphabetical order bars are taken from increasingly thicker sections. Directional solidification proceeds from the thin edge towards the riser. Therefore, the 1J bar should have the fastest solidification rate while 1B and 1C freeze most slowly. The effects of increasingly longer freezing times are to promote grain growth and a coarse network of $Mg_{17}Al_{12}$ compound in the grain boundaries, and in the case of 1B and 1C there is a greater tendency toward microporosity. In this investigation, microporosity, when it has been observed in fractures and radiographs, has been rated as slight, moderate or heavy. The rating used depends, of course, on the size and appearance of the porosity-affected area.

TABLE 2 — EFFECTS OF COMPOSITION ON PROPERTIES OF HEAVY SECTION BAR 1B MACHINED FROM SOLUTION HEAT TREATED SLOPE CASTINGS

Composition		Heat Treatment	Average Grain Diameter, in.	Compound Rating	Visual Defects at Fracture	% E in 2 in.	Tensile Properties	
% Al	% Zn						Y.S. x 1,000, psi	T.S. x 1,000, psi
5.04	<.02	800 F, 16 hr	.009	0	Small oxide inclusion.	15.3	9.5	32.6
5.05	.37	790 F, 16 hr	.007	1	—	18.0	9.8	34.2
4.97	1.03	790 F, 16 hr	.0075	1	Slight porosity.	12.7	10.6	32.6
5.98	<.02	800 F, 16 hr	.007	1	—	16.7	10.7	35.3
6.02	.40	790 F, 16 hr	.007	1	—	9.3	10.9	29.9
6.18	1.02	790 F, 16 hr	.006	1	—	11.3	12.0	34.4
6.57	.96	790 F, 16 hr	.0075	1	—	9.3	11.7	32.7
7.21	<.02	790 F, 16 hr	.009	3½	Slight porosity.	8.7	11.1	29.3
7.16	.36	790 F, 16 hr	.008	2	Slight porosity.	9.3	11.2	32.4
7.29	.96	790 F, 16 hr	.007	4	Slight porosity.	10.0	12.0	34.5
8.11	<.02	800 F, 16 hr	.009	2	—	12.0	11.5	35.2
8.03	.36	790 F, 16 hr	.007	4	Slight porosity.	6.0	12.4	26.9
7.86	.90	790 F, 16 hr	.008	5	Heavy porosity	3.3	12.9	24.6
9.00	<.02	790 F, 16 hr	.012	4	Moderate porosity.	4.0	11.0	21.4
9.05	.41	790 F, 16 hr	.0095	4½	Moderate porosity.	1.3	11.9	19.1
9.06	.98	790 F, 16 hr	.013	4	Heavy porosity.	4.0	11.0	17.5

As might be expected, we have found that heavy continuous compound networks resulting from slow solidification are much harder to put into solution than thin discontinuous compound particles which are found in sections cooled quickly. The data in Table 1 show that as section thickness increases, compound ratings and grain sizes both become larger. Heat treatability is also adversely affected by high alloy content since more compound is then present as cast.

Considering these several factors, it is reasonable to expect that bars machined from slope castings will have lower elongation and tensile strength values as section thickness increases. This trend is found in all four of the castings under discussion. Maximum tensile properties are obtained usually in the 1J bars and minimums are most often found in 1B. The 1A bar is better than the 1B due to its position at the back of the casting where it has an additional cooling surface. Yield strength ordinarily does not vary greatly with section size as it is not affected much by small variations in grain size, compound or porosity.

Higher alloy content causes a greater spread in the tensile properties of slope castings in the T4 condition. Higher values in thin sections result from the strengthening effects of more aluminum and zinc in solution. In the heaviest sections more massive compound remains after solution heat treatment, lowering properties in these areas. There is also an increased tendency for detrimental amounts of porosity to be present in the heavy sections in compositions of higher alloy content. Increasing zinc content is worse than corresponding increases in aluminum in this respect. This trend has been shown in past work.[3]

Average properties obtained on all ten bars represent both thin and thick sections and are therefore comparable to properties obtained when production castings are sectioned for certification of properties.

It should be emphasized that the higher compound ratings listed in both Tables 1 and 2 for alloys containing 7 to 9 per cent aluminum do not necessarily indicate that commercial alloys of these compositions cannot be more completely heat treated. In this investigation all solution heat treatments were terminated at 16 hr. It is common production practice to give larger castings more prolonged treatments when necessary, resulting in lower compound ratings and improved tensile properties.

The data reported in this paper are not intended to represent the properties of completely solution heat treated castings. Instead, an effort has been made to evaluate the effects of such factors as ease of heat treatment on tensile properties in castings.

The effects of section size, and in turn solidification rate, on properties in the T4 condition in the bars cut from slope castings are shown in Fig. 4 to 9 for all of the 16 alloys tested in this temper. Compositions have been grouped according to aluminum content with tensile strength and elongation for each group plotted side by side on a single figure. An examination of the resulting curves brings out some general trends.

At the 5 and 6 per cent aluminum levels there is only a slight variation in tensile strength among all 10 bars. In each casting there is a slight decrease in elongation in the thicker sections.

At 7 per cent aluminum there is a very noticeable decrease in properties in the heavier sections. In the 8 per cent aluminum series the drop is even greater, although the 8 per cent aluminum + zero zinc composition appears to be quite insensitive to thickness.

The three alloys in the 9 per cent aluminum range exhibit a very steep decrease in properties in the sections thicker than 1G. The minimum values are very low. As stated above, the low properties are due to a combination of relatively coarse grain sizes, high compound ratings and, in some cases, heavy porosity.

The effect of adding zinc at the various levels of aluminum is usually an increase in tensile strength in the thin sections and lower strength in the heavy sections. Higher zinc also tends to lower elongation in all sections.

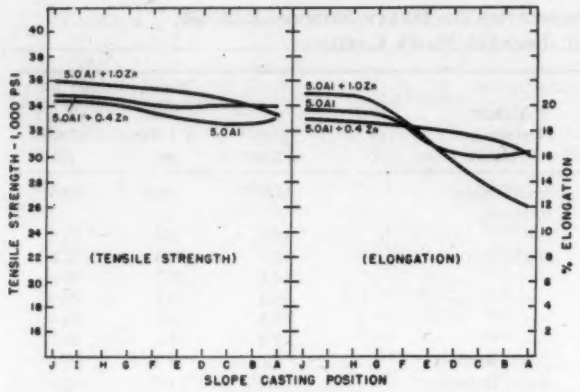


Fig. 4 — Effect of section size in the slope casting on tensile strength and elongation in the solution heat treated condition.

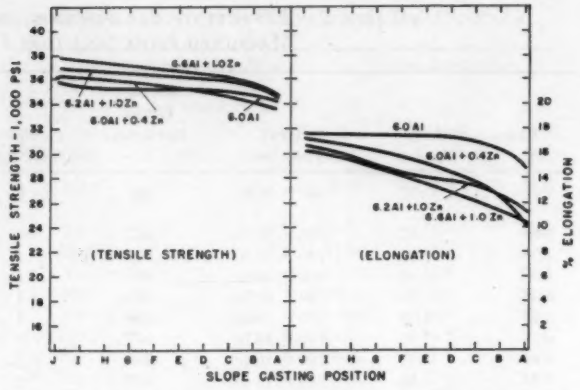


Fig. 5 — Effect of section size in the slope casting on tensile strength and elongation in the solution heat treated condition.

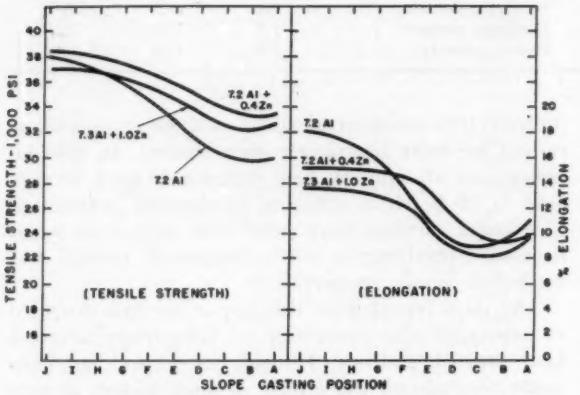


Fig. 6 — Effect of section size in the slope casting on tensile strength and elongation in the solution heat treated condition.

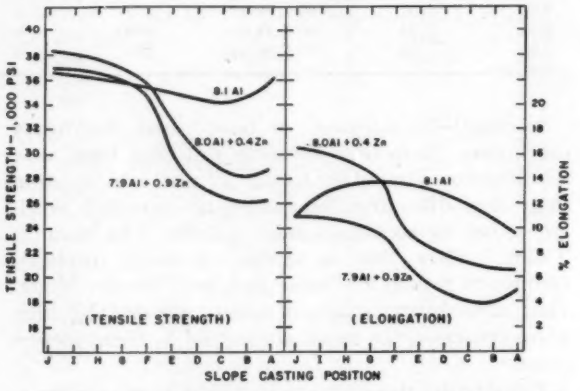


Fig. 7 — Effect of section size in the slope casting on tensile strength and elongation in the solution heat treated condition.

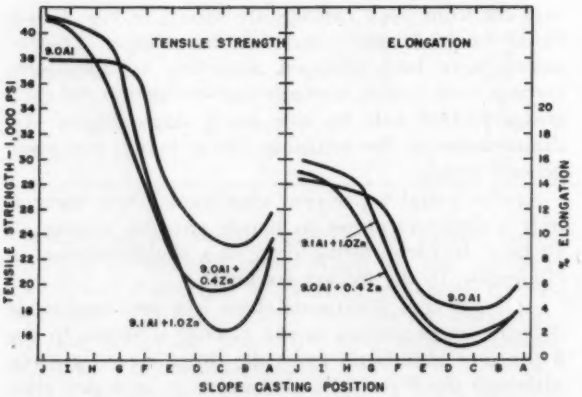


Fig. 8 — Effect of section size in the slope casting on tensile strength and elongation in the solution heat treated condition.

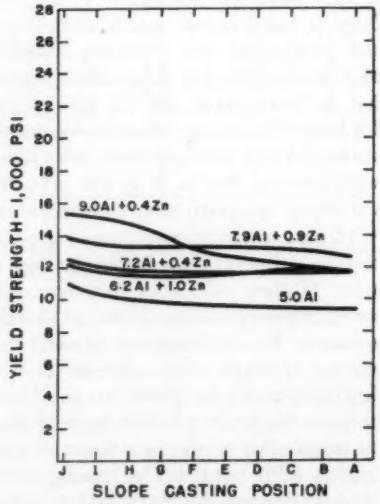


Fig. 9 — Effect of section size in the slope casting on yield strength in the solution heat treated condition.

Yield strength for the most part does not vary much with section thickness. Plots for some of the alloys are shown in Fig. 9. Only in the 9 per cent aluminum + 0.4 per cent zinc alloy is any significant drop apparent. These curves also indicate the expected increased yield strength with alloys higher in aluminum and zinc.

Tensile properties and metallographic data on all the 1B bars have been compiled in Table 2. In many slope castings this bar has the poorest properties of all ten bars of level 1. The compound ratings in this section are definitely higher at and above 8 per cent aluminum plus zinc. There appears to be some trend toward larger grain sizes at the high composition side of the series, but this may not be real.

TABLE 3 — MECHANICAL PROPERTIES OF BARS MACHINED FROM LEVEL 1 OF SOLUTION HEAT TREATED SLOPE CASTINGS AND SEPARATELY-CAST TEST BARS

Composition		Properties of 10 Bars									Properties of Separately-Cast Test Bars			
		Average			Minimum			Maximum						
% Al	% Zn	% E in 2 in.	Y.S. x 1000, psi	T.S. x 1000, psi	% E in 2 in.	Y.S. x 1000, psi	T.S. x 1000, psi	% E in 2 in.	Y.S. x 1000, psi	T.S. x 1000, psi	No. of Bars	% E in 2 in.	Y.S. x 1000, psi	T.S. x 1000, psi
5.04	<.02	17.6	9.7	33.4	15.3	9.1	32.3	20.0	11.1	34.5	2	17.8	11.8	34.6
5.05	.37	16.7	9.8	33.8	10.0	9.3	31.2	20.0	10.4	35.0	2	16.0	11.2	34.7
4.97	1.03	17.0	10.7	34.8	12.0	10.0	32.6	21.0	11.5	36.1	2	14.8	12.2	35.9
5.98	<.02	16.6	11.0	35.0	12.0	10.4	33.3	18.5	11.8	36.0	2	16.2	11.0	35.6
6.02	.40	14.6	11.2	35.0	9.3	10.6	29.9	17.5	11.8	36.3	3	16.5	12.3	37.1
6.18	1.02	14.2	11.7	35.9	11.3	11.3	34.4	16.5	12.3	37.0	4	16.1	12.5	37.6
6.57	.96	13.4	11.6	36.3	9.3	11.2	32.7	16.0	11.9	37.7	2	15.5	13.1	38.9
7.21	<.02	13.1	11.2	33.5	7.3	10.5	28.2	18.0	12.2	37.9	1	16.0	12.8	38.1
7.16	.36	13.0	11.6	35.9	9.3	10.9	32.4	14.5	12.4	38.3	2	17.5	12.4	39.2
7.29	.96	10.8	11.5	34.8	6.7	10.7	31.4	13.5	12.0	37.0	3	12.7	12.5	37.3
8.11	<.02	11.7	11.7	35.4	9.0	11.0	33.9	14.7	12.5	36.8	4	12.8	12.0	36.3
8.03	.36	11.3	12.1	35.7	6.0	11.7	26.9	16.5	12.4	38.2	2	14.5	12.6	39.0
7.86	.90	7.2	13.1	30.8	3.3	12.3	24.6	11.3	13.9	36.6	2	11.8	13.8	37.4
9.00	<.02	8.3	12.2	30.7	4.0	11.0	21.4	13.5	12.6	38.0	1	11.5	13.6	36.7
9.05	.41	6.5	13.3	28.5	1.3	11.5	19.1	15.0	15.3	41.0	2	15.5	15.0	41.6
9.06	.98	8.1	12.8	28.6	2.0	10.9	15.9	16.0	14.5	41.3	2	12.8	15.4	41.6

The characteristic discoloration which identifies porosity was seen in the fractures of several of these bars. These fracture defects, incidentally, were also checked by radiography with similar results. Almost no indications of porosity were seen up to a concentration of 7 per cent aluminum plus zinc. Above 8 per cent alloy content more and heavier porosity was present. At this same level the tensile properties dropped to very low values.

The properties of these 1B bars indicate the minimum properties which may be expected in solution heat treated castings. This statement applies to the majority of magnesium castings, not to very thin nor to very heavy section castings. Actually, none of the alloys in this series have particularly high porosity tendencies and, in many cases, properly risered production castings will be free from detrimental amounts of porosity. However, it appears obvious that lower minimum properties will be more common with the high composition alloys.

Average, minimum and maximum tensile properties in the T4 temper obtained in the bottom levels of slope castings in all 16 compositions are listed in Table 3. Separately-cast test-bar data are also included.

Maximum properties in slope castings are about equal to those of test bars, while minimum elongation and tensile strength values tend to decrease above about 7 per cent aluminum plus zinc.

The effect of composition on tensile strength is shown graphically in Fig. 10. Alloy content in this case is the sum of aluminum and zinc for each composition with zinc being always 1.0 per cent or less. In this way fewer curves are needed to present the data. A striking difference is apparent here in the strengths of separately-cast test bars compared to the averages obtained for slope castings.

The test-bar curve shows a steady increase with increasing alloy content. The 10-bar average for slope castings, on the other hand, reaches a maximum strength of about 36,000 psi near 7.2 per cent alumi-

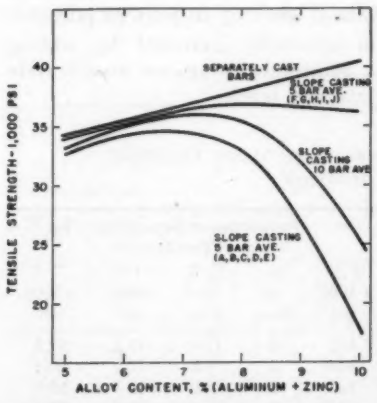


Fig. 10 — Effect of alloy content on tensile strength in the solution heat treated condition.

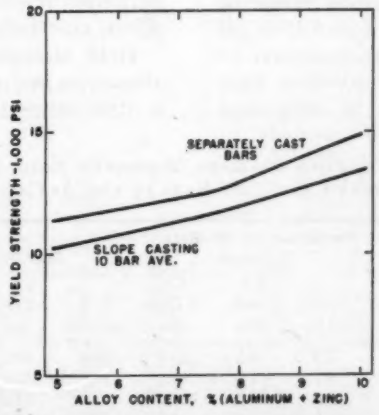


Fig. 11 — Effect of alloy content on yield strength in the solution heat treated condition.

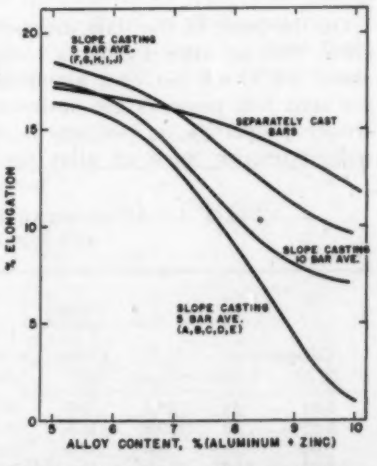


Fig. 12 — Effect of alloy content on elongation in the solution heat treated condition.

num plus zinc and then decreases rapidly above 8 per cent alloy content. The scatter of values about the average curve for castings is quite small. Minimum strengths in slope casting are, of course, several thousand psi below the averages and the scatter is also much greater. As we have pointed out, there is a definite trend toward very low minimum strengths at the high composition side of the range.

The strong influence of section size on tensile strength is illustrated by plotting separate curves for bars from the thinner and the thick sections of level 1 of slope castings. The former group of five bars (F, G, H, I, J) has an average curve which reaches its maximum of 37,000 psi at 8 per cent aluminum plus zinc and only decreases to about 36,000 at 10 per cent. The average values for the heavy sections (A, B, C, D, E) are naturally lower and the drop in strength is very rapid when alloy content exceeds 8 per cent. These curves emphasize again that the tensile strength of this family of alloys is largely dependent on section thickness as well as total alloy content.

The relationship between yield strength and composition is fairly simple. This is shown in Fig. 11 for the T4 condition. The average curve for slope castings shows the same increase in yield strength with rising alloy content as the test bar curve, but lies about a 1,000 psi below it. Since this property is not affected greatly by section size, no separate curves are shown for thin and thick sections. The minimum yield strength for each casting is not much lower than the average value for the same composition.

Plots of elongation vs. composition again for the T4 temper, are in Fig. 12. Ductility is generally lowered by increased alloy content. Since elongation is quite sensitive to many variables such as grain size, compound ratings and porosity, it is not surprising to find the average values for the ten slope-casting bars are considerably below the separately-cast test bar curve. However, the average elongation for the F, G, H, I, J bars is only slightly below the test bar curve at more than 7 per cent alloy content and is as good or better in lower compositions. The five thick section bars have the poorest ductility as would be expected.

On the basis of the data discussed above, we conclude that an alloy having a composition range of about 5.6 to 6.8 per cent aluminum and 0.4 to 1.0 per cent zinc possesses the optimum combination of tensile properties in castings in the solution heat treated temper. Such an alloy would be designated AZ61 in the A.S.T.M. system. As will be shown in the next section, this composition also has properties as-cast which are superior to those of present commercial alloys.

Actually an alloy similar to AZ61, dowmetal E having the composition magnesium + 6.5 per cent aluminum, was used to a limited extent for sand castings many years ago. Use of this composition was discontinued in favor of the later developed ternary Mg-Al-Zn alloys.

The information obtained from the slope casting in the present study shows that due to ease of heat treatment and low porosity tendency, AZ61 does possess a superior combination of tensile properties in actual castings.

This example emphasizes the value of using the slope casting in addition to separately-cast test bars in alloy development work. More extensive use of this laboratory casting may reveal some other neglected alloys whose properties in castings are relatively unknown, but might be very attractive for particular applications.

Most of the favorable characteristics of AZ61 are also shared by AZ81XA; however, due to its lower alloy content, AZ61 may have some advantages over AZ81XA.

Additional development work on AZ61 on a larger scale and under actual production conditions must be completed before standards can be proposed for this alloy. Castings of varying size and complexity should be studied to evaluate more completely the castability features and mechanical properties of this tentative composition range.

Properties As-Cast – Table 4 contains tensile properties of as-cast slope castings and test bar for several compositions, including some commercial alloys.

The average and minimum properties in slope castings in the as-cast condition are presented graphically in Fig. 13 to show the relationship between them and composition. Average tensile strength is highest in alloys of low composition. These also have high minimum tensile strengths. AZ81XA-F has a little higher average tensile strength than the other commercial alloys tested and is far superior at the minimum level. The very low minimums in the other alloys, especially AZ63A, are due in part to porosity.

Yield strength is generally increased by adding aluminum and zinc. Minimum values are usually only a little below the average levels.

TABLE 4 – MECHANICAL PROPERTIES OF BARS MACHINED FROM LEVEL 1 OF SLOPE CASTINGS AND SEPARATELY-CAST TEST BARS IN THE AS-CAST CONDITION

Composition		Properties of 10 Bars									Properties of Separately-Cast Test Bars			
		Average			Minimum			Maximum						
% Al	% Zn	% E in 2 in.	Y.S. x 1000, psi	T.S. x 1000, psi	% E in 2 in.	Y.S. x 1000, psi	T.S. x 1000, psi	% E in 2 in.	Y.S. x 1000, psi	T.S. x 1000, psi	No. of Bars	% E in 2 in.	Y.S. x 1000, psi	T.S. x 1000, psi
5.04	.34	13.1	9.7	30.3	7.3	9.1	25.8	20.0	10.7	34.1	4	13.9	11.8	32.2
6.02	.40	9.6	11.5	29.5	6.7	11.0	25.0	15.5	12.4	34.8	2	9.8	12.2	30.4
7.60	.59	4.0	12.6	24.4	2.0	11.6	20.0	7.0	13.6	31.5	3	6.0	15.1	28.0
6.44	3.00	4.0	11.6	22.8	0.0	8.6	9.6	8.5	14.8	33.4	3	5.2	15.0	28.4
9.42	1.90	1.3	14.7	20.6	0.7	12.1	12.1	3.0	16.9	27.4	3	1.8	17.9	24.2
9.94	<.02	1.5	14.5	20.1	0.7	13.4	14.6	3.0	16.2	26.8	3	2.5	17.6	25.3

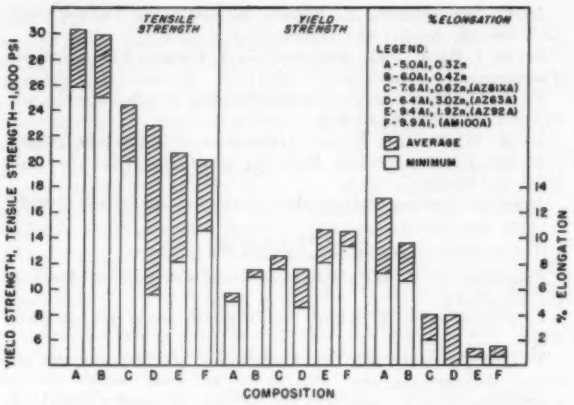


Fig. 13 — Average properties of slope castings in the as-cast condition.

Ductility, like tensile strength, is highest in alloys such as the 5 per cent and 6 per cent Al + 0.4 per cent Zn and lowest in AZ92A and AM100A.

The six alloys tested tend to fall into three groups. The two with lowest alloy content have the best levels of tensile strength and elongation and lower yield strength. Their minimum properties generally lie only a little below the average values.

AZ81XA and AZ63A are in an intermediate position as regards tensile properties in the as-cast state. However, AZ81XA is far superior to AZ63A in that its minimum properties are much higher. This is due to the improved soundness of AZ81XA over AZ63A.

The two well-known commercial alloys, AZ92A and AM100A, have almost identical average tensile properties in slope castings in the as-cast condition. AM100A exhibits higher mimimum values because it is a composition having a very low porosity tendency.

The reason for better levels of tensile strength and elongation in castings with low alloy contents is quite straightforward. Much less massive compound is present in the as-cast state causing the alloy to be less brittle than those of higher composition. Conversely, when complete compound solution is attained these alloys show less improvement in strength and ductility than those with more aluminum and zinc, assuming soundness and equivalent grain size in all cases.

Properties of Upper Levels — As indicated in Fig. 3, bars may be machined from the upper levels of the slope casting as well as level 1. This is feasible for compositions such as EZ33A which have very low porosity tendencies. However, in most Mg-Al-Zn alloy castings considerable microporosity is encountered above level 1 making the bars from these levels useless unless the effects of gross porosity are being studied.

Tensile properties of two alloys in level 1 through 4 are in Table 5. The first, 5 per cent Al + 0.34 per cent Zn, was tested without heat treatment. A large drop in average and maximum properties is shown between levels 1 and 2. With this low composition, level 2 was sound and the drop in values is due to the fact that level 2 contains no thin section bars. At levels 3 and 4 increasing amounts of porosity are encountered causing an additional decrease in strength and ductility. Even so, it is interesting to see that the average and minimum properties of all 4 levels of this casting are superior to those in Table 4 for level 1 of AZ81XA-F, AZ63A-F, AZ92A-F and AM100A-F.

The other casting included in this table was given a solution heat treatment. This alloy contained 6 per cent Al + 0.4 per cent Zn and is the same one listed in the other four tables. Again a drop in properties is experienced in level 2 as expected. One bar of the four in this level showed porosity in its fracture. In level 3, two bars were unsound and the average properties again decreased slightly. Both bars in level 4 were porous, but properties drop only slightly from level 3. Comparing these data to that in Table 3, it is apparent that this alloy has average and minimum properties in levels 1 to 4 which are superior to those obtained in level 1 in alloys in the AZ91C range and the high side of the AZ81XA range. It appears that the 6 per cent Al, 0.4 per cent Zn composition is not as sensitive to porosity as those with higher alloy contents.

Since we have found that the properties of level 1 are comparable to those found in the usual run of production castings, bars are seldom sectioned from the upper levels from Mg-Al-Zn alloy slope castings.

TABLE 5 — MECHANICAL PROPERTIES OF LEVELS 1 TO 4 OF TWO SLOPE CASTINGS

Composition				Mechanical Properties								
				Average			Minimum			Maximum		
% Al	% Zn	Level	No. of Bars	% E in 2 in.	Y.S. x 1000, psi	T.S. x 1000, psi	% E in 2 in.	Y.S. x 1000, psi	T.S. x 1000, psi	% E in 2 in.	Y.S. x 1000, psi	T.S. x 1000, psi
5.04	.34	1	10	13.1	9.7	30.3	7.3	9.1	25.8	20.0	10.7	34.1
As-cast		2	4	8.7	9.4	26.0	5.3	9.4	24.2	10.7	9.9	27.8
		3	3	7.1	9.2	24.5	4.7	9.0	21.5	8.7	9.7	27.0
		4	2	5.0	9.6	20.8	4.7	9.0	20.8	5.3	9.4	20.8
		Total	19	10.4	9.6	27.4	4.7	9.0	20.8	20.0	10.7	34.1
6.02	.40	1	10	14.6	11.2	35.0	9.3	10.6	29.9	17.5	11.8	36.3
-T4		2	4	11.0	10.3	30.6	8.0	10.2	26.6	14.0	10.4	33.8
790 F,	—	3	3	10.2	10.2	29.4	8.0	10.0	26.8	14.7	10.4	34.0
16 hr		4	2	9.3	10.2	29.2	9.3	10.0	29.1	9.3	10.2	29.3
		Total	19	12.6	10.7	32.6	8.0	10.0	26.6	17.5	11.8	36.3

Summary

A special laboratory type casting has been developed which provides a range of tensile properties equivalent to those found in production castings of magnesium alloys. This "slope casting" is useful in alloy development work to provide tensile property data for thick and thin sections, in heat treatment studies, grain size, and porosity tendency work.

A series of Mg-Al-Zn alloys have been investigated using the slope casting. These data show that in the as-cast and solution heat treated tempers alloys such as AZ61 and AZ81XA of low-composition have superior average and minimum tensile properties in castings. They are more sound and easier to heat treat then alloys in the AZ91C range for instance.

It has been demonstrated once again that separately-cast test bars do not necessarily indicate the range of properties in actual castings. Section size should be considered in design calculations due to its effects on heat treatability, grain size and mechanical properties of castings.

While we have said that the range of tensile properties obtained by testing level 1 of slope castings corresponds to those found in production castings, it is obvious that this comparison is not valid for *all* castings.

Thin-wall castings less than ¼ in. thick or other fast-cooled sections cannot be expected to exhibit properties similar to even the thin sections of the slope casting. Very heavy section castings, 3 in. thick or more and others where large masses of slowly-cooled metal are present tend to have coarse grain and to be difficult to heat treat—more so than the heaviest sections of the slope casting.

Also, we should emphasize that this test casting gives results reflecting several variables acting together. It is not a pure section size test but rather a solidification rate test producing a casting showing all the effects of solidification rate on grain size, compound distribution, porosity, etc. To reflect the effect of section thickness alone, sample castings should be sound and completely heat treated. However, we feel the present casting does involve the variables present in production castings and, therefore, the data obtained are significant in alloy evaluation studies.

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DISCUSSION

Chairman: E. V. BLACKMUN, Aluminum Company of America, Pittsburgh, Pa.

Co-Chairman: F. L. TURK, Joy Mfg. Co., New Philadelphia, Ohio.

W. D. WALTHER[1] *(Written Discussion):* Mr. Pearson has obtained information that should be of great value to the magnesium casting industry. He should be congratulated for the work he has reported. He has rendered a service by recommending adoption for many uses of AZ61 and AZ81A over such alloys AZ63 and AZ91C.

Some of the comments I have to offer are based on work completed by Dix Chandley of M.I.T. under a contract granted by Frankford Arsenal.

It is curious that magnesium foundries have so long used alloys AZ63A, AZ91C, even AZ92A, in spite of erratic and poor mechanical properties obtained in castings made therefrom. In a report*, the sensitivity of these alloys to solidification rate was clearly demonstrated. Test bars of alloy AZ92A, for example, taken from actual castings and from apparently sound regions, may have tensile and yield strengths and elongation 25 to 33 per cent lower than standard test bar values, and then be considered pretty good.

Standard test bar values for such alloys as AZ91C and AZ92A give an erroneous impression because the ½-in. test bar section induces a rapid solidification rate, and consequently disperses the ternary compound so that it is not continuous at dendritic grain boundaries. As pointed out by Mr. Pearson, and many others, a ½-in. round section gives approximately the same solidification rate as a ¼-in. plate section (and we all would consider such a section rather thin). Besides providing a better dispersion of secondary phases, a rapid solidification rate actually reduces the amount of ternary compound formed.

The ternary compound of which we speak, $Mg_3Al_2Zn_3$, melts at 685 F (636 C). It forms as a consequence of non-equilibrium solidification of the melt. One may feel sure of having a significant amount of the ternary compound in ⅛-in. to ¾-in. plate sections of castings, and even more under risers of such section thicknesses. Castings of AZ63A and AZ92A are often manufactured for use in the solution-treated or solution-treated and aged condition. The solution treatment in many foundries consists essentially of a short preheat at 680 F for two hours, and then raising the temperature to 770 F for some definite time.

Such a heat treatment inevitably will cause melting of any $Mg_3Al_2Zn_3$ phase present at grain boundaries. The short time at 680 F is not at all sufficient to permit diffusion of enough zinc and aluminum atoms away from the grain boundaries to decompose $Mg_3Al_2Zn_3$. Even 48 hours at 680 F is insufficient for accomplishing this decomposition. Thus, on raising the temperature above 685 F, a molten phase is produced at grain boundaries. With sufficient time at 770 F, depending upon the extent of segregation and alloy composition, zinc and aluminum atoms diffuse away from the grain boundary toward the center of the dendrites, in accordance with the dictates of achieving an equilibrium of composition.

Under such circumstances the liquid phase is in effect "solidifving by diffusion" and a contraction will occur because of solidification shrinkage. Since there is no additional molten metal present to feed the contraction, porosity will form at grain boundaries, with a subsequent deleterious effect on strength. Is it little wonder then why we obtain erratic mechanical properties from AZ92A etc., depending upon amount of ternary eutectic formed, and the heat treatment procedure used. (The phase $Mg_{17}Al_{12}$ under some conditions of heat treatment, also may produce fusion porosity.)

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Die castings and permanent mold castings often will obviate the difficulties encountered with AZ63A etc., since they often have solidification rates equally as fast as a ½-in. round, separately cast, sand-cast test bar. On the basis of principles of solidification, sand-cast AZ92A and AZ63A will inevitably be susceptible to grain boundary and interdendritic fusion during commercial heat treatment.

Mr. Pearson has pointed out that there is a difference in grain size of 0.005 to 0.01 in. in his slope castings. To this difference he has attributed some decrease in the mechanical properties with increasing section thickness. In wrought alloy AZ61 there is virtually no change in tensile strength, yield strength, or elongation over the grain size range 0.003 -0.010 in.* The difference in mechanical properties obtained in the slope casting is mainly a function of microporosity induced by shrinkage, at casting and during heat treating, and the distribution of intermediate phases $Mg_{17}Al_{12}$ and $Mg_3Al_2Zn_3$ at dendrite boundaries.

AZ61 is of such composition to produce a limited amount of ternary eutectic phase during solidification. This is why this alloy exhibits the combination of good tensile, yield strength and good elongation in the solution treated condition.

MR. PEARSON *(Reply to Mr. Walther's Discussion)*: The ternary phase $Mg_3Al_2Zn_3$ does not exist in magnesium base alloys with an Al:Zn ratio of greater than about 3:1.[1] Thus the phase is not found in AZ92A, but may be present in significant amounts in AZ63A.

The non-equilibrium solidus of AZ92A is not below 750-755 F. Fusion voids are uncommon in AZ92A castings preheated at least two hours at a temperature below 750 F and then held at 770 F for 16-20 hours. We cannot agree that AZ92A is inevitably susceptible to fusion during the usual commercial heat treatment cycle. In fact, AZ92A is one of the most satisfactory magnesium alloys in use today. With attention to good foundry practice and proper heat treatment, a high level of mechanical properties in castings can be consistently attained.

In the case of AZ63A, the accepted non-equilibrium solidus is indeed about 685 F,[2] and fusion is often encountered in this alloy making the heat treatment of heavy sections a real problem. AZ63A is, of course, ordinarily solution heat treated at a maximum temperature of about 730 F.

In summary, while the fusion problem definitely causes difficulties in the heat treatment of AZ63A due to the presence of $Mg_3Al_2Zn_3$, this phase seldom, if ever, occurs in AZ92A, AZ91C, AZ81A or AZ61. The favorable combination of mechanical properties obtained in AZ61 results from ease of heat treatment and a relatively low microporosity tendency. For these reasons AZ81A and AZ61 can be recommended as replacements for AZ63A and AZ91C in the as-cast condition and the T4 and T5 tempers. AZ81A and AZ61 are not suitable substitutes for AZ92A-T6 or AZ91C-T6 since the lower alloy content compositions do not age harden sufficiently to produce the desired levels of yield strength.

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J. W. Meier[3] *(Written Discussion)*: I would like to congratulate the author for his interesting paper and the presentation of data which should prove very valuable in the never-ending discussions about the actual properties of castings.

I am especially interested in this paper because we are investigating various aspects of the same problem and presented some of our results at the Annual Meeting of the CIMM at Quebec.** Our investigation was undertaken to study various factors which affect the properties of cast magnesium alloy test bars, whether they are separately cast, attached to or cut out from castings. This includes shape and size of the test bar, the effect of machining, of casting cross-section, location of the test bar in the casting, pouring temperature, etc. In this study we used separately-cast test bars, flat plates of different thicknesses, round bars of various diameters, step castings, a wheel-shaped test panel, and actual production castings.

First, the problem of casting shape. The choice of the slope casting for a preliminary evaluation of various alloys, especially of experimental alloys, is probably very useful, but only if it is used as one of a series of other evaluation tests and if the quality of the test coupons is truly representative of the casting quality requested in commercial castings. The use of obviously defective test coupons, as for example shown in some bars in Table 4 (where the UTS for AZ63 is as low as 9600 psi, against an average for this alloy of some 28-30,000), is wrong – because the evaluation of properties of castings which would be commercially unacceptable has no practical value.

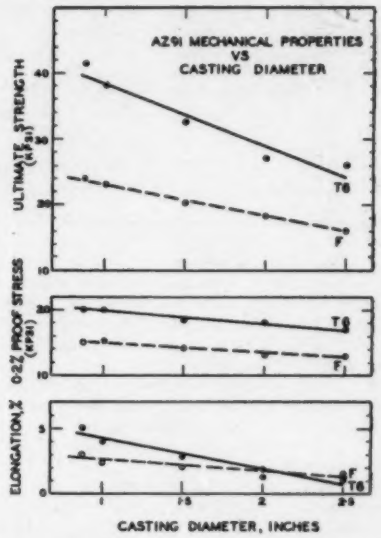


Fig. A — AZ91 Mechanical Properties vs Casting Diameter

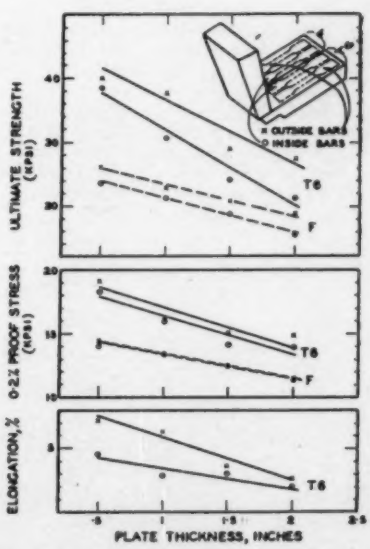


Fig. B — Effect of Plate Thickness and Location of Test Bars on Mechanical Properties

*"Study of Techniques for Refinement of Cast Structures in Magnesium," Pitman-Dunn Laboratories, Contract No. DA-19-020-ORD-2849, by Foundry Section, Department of Metallurgy, M.I.T.

2. Department of Mines & Technical Surveys, Ottawa, Canada.

**A. Couture and J. W. Meier, "The Effect of Various Factors on the Mechanical Properties of Cast Magnesium Alloy Test Bars," presented at the Annual General Meeting, Canadian Institute of Mining and Metallurgy, Quebec City, Quebec, (April 11, 1956).

TABLE A – EFFECT OF TEST BAR SHAPE ON TENSILE PROPERTIES OF MAGNESIUM ALLOY SAND CASTINGS

AZ91-F (Melt No. 3635 - 8.87 Al, 0.63 Zn, 0.35 Mn)
AZ91-T6 (Melt No. 3806 - 8.99 Al, 0.64 Zn, 0.28 Mn)
Heat Treatment: 24 hr at 410 C (770 F), air cool
16 hr at 200 C (392 F)

	No. bars	UTS kpsi	0.2% PS kpsi	Elong. % in 4D
As Cast (F)				
Round bars*	14	32.1	20.8	4.0
Flat Bars**	16	29.5	17.8	4.8
Heat Treated (T6)				
Round bars*	12	41.9	26.9	4.2
Flat bars**	13	39.7	22.5	5.2
Separately-Cast Test Bars				
As Cast (F)	2	26.5	15.0	4.8
Heat Treated (T6)	2	40.3	20.7	5.0

*Gauge diam (D) of round bars – 1/8 in., gauge length – 1/2 in.
**Thickness of flat bars – 5/32 in., wide – 1/2 in., gauge length – 2 in.

Another question arises whether this slope casting is really representative of all or at least most of the solidification patterns found in commercial castings. We found that there are quite a few factors which have a statistically significant effect on the properties of a test bar, and I would like to illustrate this by some examples of our findings.

Figure A shows the effect of cross-section on the properties of round bars and, as would be expected, the cross-section affects the UTS and the elongation values appreciably, and the YS much less. It is also logical that the T6 temper is more affected than the as-cast material.

Figure B shows a similar pattern for flat plates of different thickness cast in alloy AZ80. But, in this case it was found that statistically significant variations of properties are found between the inside and the outside bars. This could, of course, be changed by a variation of the gating but it is shown because it underlines the importance of the location of the test bar in the casting. Another experiment showed that in some cases it is important how the test bar is located in the cross-section itself, that is, if it is machined out from the middle of the section or from a location near the surface.

Table A compares the properties of bars cut out of the same castings but machined to different shapes, that is, round and flat test bars. Both of the test bar shapes were substandard because the plates were cast in 3/16-in. and 1/4-in. thicknesses. It is quite obvious that a significant difference in UTS and YS was obtained both in the as-cast and in the T6 conditions. (The rather high results are due to the use of substandard test bars.)

There are other factors which are important such as, the size of the test bars, the effect of machining, the gage diameter to gage length ratio, etc.

I mention all these factors, because I believe that the slope casting, or any other test casting, can be useful only if all these factors are taken into consideration and the comparison alloy is made under standardized conditions.

Another important point is the heat treatment. The author mentions, and it is quite obvious from the values given in the tables, that the heat treatment was not sufficient, in some cases at least, to obtain properties typical for the alloys investigated. I consider this a great mistake, because a comparison of properties of alloys has to be done under commercially usable conditions. As long as this is an internal laboratory test and is understood by metallurgists, this would not be harmful. But, the publication of such data may be harmful because some metallurgically innocent designer or user of castings may see these figures and read in the paper that such properties are typical in larger castings, and may be scared to use any magnesium alloy. If I were a sales manager I would be very unhappy about such data, especially because we know that in castings produced commercially to rigid specifications such low properties would not be tolerated. Here we have them only because the heat treatment was inadequate and because some of the test coupons were unsound.

One further comment, the author says "that it has been demonstrated once again that separately-cast test bars do not necessarily indicate the range of properties in actual castings." This is true, but this is not the function of a separately-cast test bar and no foundryman or metallurgist claims that it is so. We want to use the separately-cast test bar only as measure of melt quality and heat treatment, and let's not confuse these definitions.

MR. PEARSON *(Reply to Mr. Meier's Discussion)*: Mr. Meier has presented some interesting comments and additional information in his discussion. Obviously, as he points out, any comparison of alloys should be made under a certain set of conditions. As described in our paper, such factors as pouring temperature, location of bars and heat treatment were all held as constant as possible. It is true that flat bars were necessarily machined from the thinner sections of the slope casting. However, in the past, we have not seen any constant and reproducible effect on mechanical properties of flat vs. round test bars.

The fact that all sections of the castings of the various compositions were not fully solution heat treated was one of the variables being evaluated. Ease of heat treatment is one of the significant advantages of low alloy content compositions.

As emphasized in the summary section of the paper, the slope casting is intended for laboratory use. Like most test castings, its design is a compromise and cannot, of course, represent all the situations encountered in production castings. We agree that the results obtained from such tests should be interpreted by qualified personnel, but certainly this is true in the case of most research data.

CRACKING AND LIFE OF INGOT MOLDS

By

J. Duflot*

ABSTRACT

Examination of data submitted by French basic Bessemer (Thomas) steel plants and foundries made it possible to show the desirability of having a low M/I ratio for 3- to 5-ton square ingot-molds:

M/I = Mold weight/Ingot weight
approximately equal to 1.

However, in spite of favorable results of ingot-mold iron consumption per ton of steel ingots, these light ingot-molds have a greater tendency to fail by cracking than by crazing. This paper describes possible causes of mold cracking.

The study of a 3.5-ton mold for 3.6-ton ingots has shown that some foundry practices may have an influence on the occurrence of cracking.

The part played by chemical composition of the mold, especially of elements present in trace amounts, is studied. The general character of the occurrence of cracking on the hottest walls of the mold is also shown.

I. Introduction

Ingot-mold cracking is an intricate problem. It involves stress-distribution in the mold and the physical properties of the iron in the hot state. Both are little known and difficult to investigate. This is a very important problem, since cracking often causes comparatively new ingot molds to be scrapped.

Thus, the problem of mold cracking prevention often is the same as the problem of eliminating a premature cause of ingot molds scrapping, the cost of which constitutes a significant part of the production cost of steel, at least in Europe.

This problem presents itself differently for the square 4-ton ingot molds which are used in the French basic Bessemer steel plants, for the slab molds used in strip mills, and for the hot-topped molds in which the ingot remains for a fairly long time. In this paper the author shall limit himself to the study of the cracking of the square molds used for 3- to 5-ton ingots. To begin with, we are going to state the problem in the case of square molds; we are going to show the interest of light molds and how cracking affects them. Then, a number of causes of cracking will be indicated, as they were described to the author in the French steel plants and foundries. Subsequently, the author will present information he derived from the examination of a number of 3.6-ton ingot molds.

II. Cracking Problem of 3- to 5-Ton Square Molds

1. Causes of Ingot-Mold Scrapping

Briefly, the main causes for scrapping ingot molds are as follows:

A. Burning – Elongated hollow grooved into the mold wall by the molten steel jet.

B. Crazing – Network of scores on the inner surface of the molds which impedes stripping, or gives the ingot a defective skin.

C. Tearing – Removal of an iron splinter, adhering or not to the ingot.

D. Cracking – Crack starting on the inner surface and developing gradually through the whole thickness of the mold. It may reach the corner or get through the mold wall. It is this defect which will be more particularly examined in this paper.

E. Clink – The clink, which occurs with a distinctive sound, is a cracking which extends to the whole, or nearly the whole height of the mold. It is an extreme case of cracking.

From an investigation[1] conducted by the French Joint Committee on ingot molds, whose findings have been used as a base for the present paper, it appears that in 18 French basic Bessemer steel plants, in 1951 there were on the average 61.6 per cent crazing, 34.3 per cent cracking, and about 5 per cent of accidental causes (ear breakage, burnings, etc.).

If one draws, for different steel plants, the life histograms of the cracked molds (F), and of the crazed ones (C) (Fig. 1 a), the following cases may be obtained:

A. The cracked molds (F) histogram may be located on the left of the crazed molds (C) histogram,

B. The cracked molds (F) histogram may be superimposed upon the crazed molds (C) histogram,

C. The cracked molds (F) histogram may be located on the right of the crazed molds (C) histogram.

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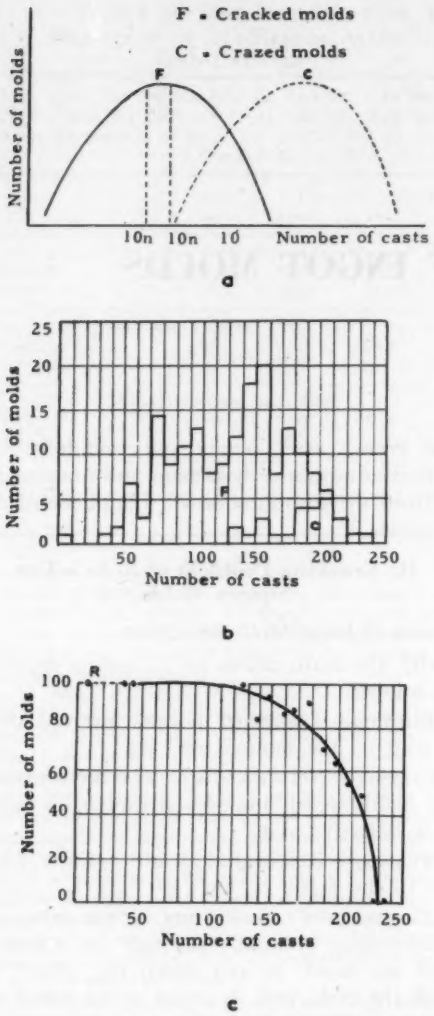


Fig. 1 — Histograms for cracked and crazed molds.

It is easy to understand that in the first case there are among the molds which have been scrapped after a comparatively low number of melts, for instance from 80 to 90 melts, more cracked molds than crazed ones. Contrariwise, in the same case, among molds having been used for many melts, for instance from 150 to 160 melts, there are more crazed molds than cracked ones. In the third case (C) where the cracked molds histogram is located on the right of the crazed molds histogram, for molds scrapped after 80 or 90 melts, there will be more crazed molds than cracked ones; contrariwise, for molds scrapped between 150 and 160 melts, there will be more cracked molds than crazed ones. This can be expressed by stating that the percentage R of cracked molds in the total number of molds which have been scrapped between $10\,n$ and $10\,n + 9$ melts will vary according to cases A, B or C.

—the percentage will decrease with n in case A.
—it will be about constant in case B.
—it will increase with n in case C.

Instances of these three cases can be found in the steelmaking practice, as well as in the technical literature.[1] The author shows here only the histograms (Fig. 1B) relating to the mold studied in Section IV, which correspond to the first case, as well as curve R (percentage of cracked molds in the total number of scrapped molds) as a function of the number of melts poured into the molds (Fig. 1C).

Steelmen know that, generally speaking, crazing is related to the conditions of mold use: more crazing occurs when the ingot remains longer in the mold, or when the molds are too hot when used again. Crazing is fairly well controlled since the mold is scrapped when the surface condition of the ingot, related to the surface condition of the inner face of the mold, is no longer satisfactory.

Contrariwise, cracking is related to the ability of a mold to withstand the effects of expansion and shrinkage resulting from every teeming operation. It is a scrapping cause which cannot be controlled, except by controlling the heating of the mold walls.

Under these conditions, the relative position of histograms F and C (Fig. 1) can be considered as representing the iron capacity to assure the service required:

1. if F is on the left of C (A), the iron quality or the mold wall thickness are not sufficient for the conditions of use,

2. if F is superposed on C (B) the iron, or the mold, are well adapted to the conditions of use,

3. if F is on the right of C (C), resistance of the iron is too high, or the mold wall is too thick for the conditions of use.

It seems, therefore, that once histograms F and C have been drawn, the problem is to increase or decrease the mold wall thickness until the histograms are superimposed. In fact, we shall see subsequently that light molds are more economical in spite of their marked tendency to cracking. This involves, in the case where histogram F is on the left of histogram C, that the superimposition of the two histograms is to be obtained through improvement of the iron quality, or of molding conditions (cf Section III) to retain the benefit of light molds, while getting a proportion

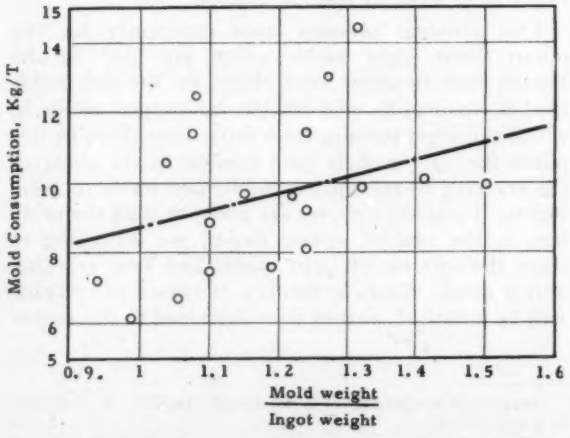


Fig. 2 — Relationship of mold consumption vs M/I (Mold Weight/Ingot Weight). 1 kg = 2.2 lb.

of cracked molds about equal to that of crazed molds. It is the whole problem of the light molds cracking. The author does not pretend to have solved the problem entirely.

2. Light-Weight Molds

Figure 2 shows, for the French basic Bessemer steel plants, the consumption in kilograms of iron per ton of steel as a function of the ratio mold weight/ingot weight.

It can be seen that the iron consumption increases markedly when this ratio M/I increases.

On the other hand, for the different steel plants whose consumptions are plotted in Fig. 2, if we draw the diagram giving the percentage of cracking in the number of scrapped molds as a function of ratio M/I = mold weight/ingot weight (Fig. 3), or the diagram giving the percentage of cracking as a function of the mean wall thickness (Fig. 4), or the diagram giving the percentage of cracking as a function of the mold consumption expressed in kilograms per ton of steel (Fig. 5), we find that the percentage of cracking is higher for the lighter molds.

Thus, the light molds are definitely of interest from an economical standpoint; however, their use results in an increase of the molds scrapped due to cracking. This enhances the importance of learning more about cracks.

3. Case of the Slab Molds Used in Openhearth Steelmaking, and of the Square Molds Under 3 Tons

The problem of the optimum value of the ratio mold weight/ingot weight is not restricted to square molds of about 3 to 5 tons. However, if for the latter the optimum value of ratio M/I is approximately 1, the case is different for the other molds. Figure 6 shows the evolution of ratio M/I as a function of the weight of the ingot body, expressed in metric tons, for a number of square electric steel molds, of openhearth slab molds, as well as for the square basic Bessemer molds already considered. Disregarding the

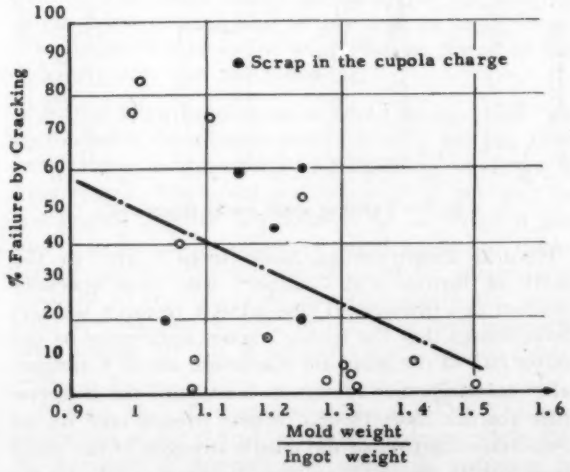


Fig. 3 — Relationship of failure by cracking vs M/I (Mold Weight/Ingot Weight).

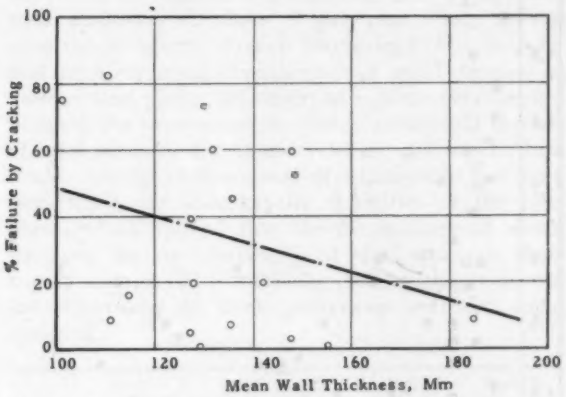


Fig. 4 — Relationship of failure by cracking vs mean wall thickness.

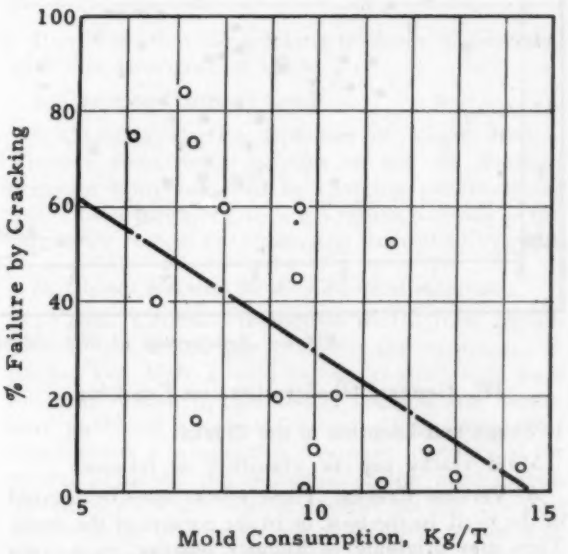


Fig. 5 — Relationship of failure by cracking vs mold consumption. 1 kg = 2.2 lb.

obsolete molds, there is a marked tendency toward a decreasing evolution of ratio M/I when the ingot body weight increases. The dotted line curve gives the optimum value of this ratio according to N. H. Bacon.[2]

The appearance of Fig. 6 and Bacon curve show that the M/I ratio has to be chosen taking into account the ingot weight. Other problems may be involved, design problems for instance.

In the case of openhearth slab molds used in modern strip mills, there is a problem of design for the horizontal section of the mold; especially, account being taken of the ratio between the long and the short sides of the mold, it is advisable to choose the appropriate ratio of the thicknesses in the corners, as well as in the large face and the small face of the mold. This problem will not be discussed in this paper which will be limited to discussion of the square molds.

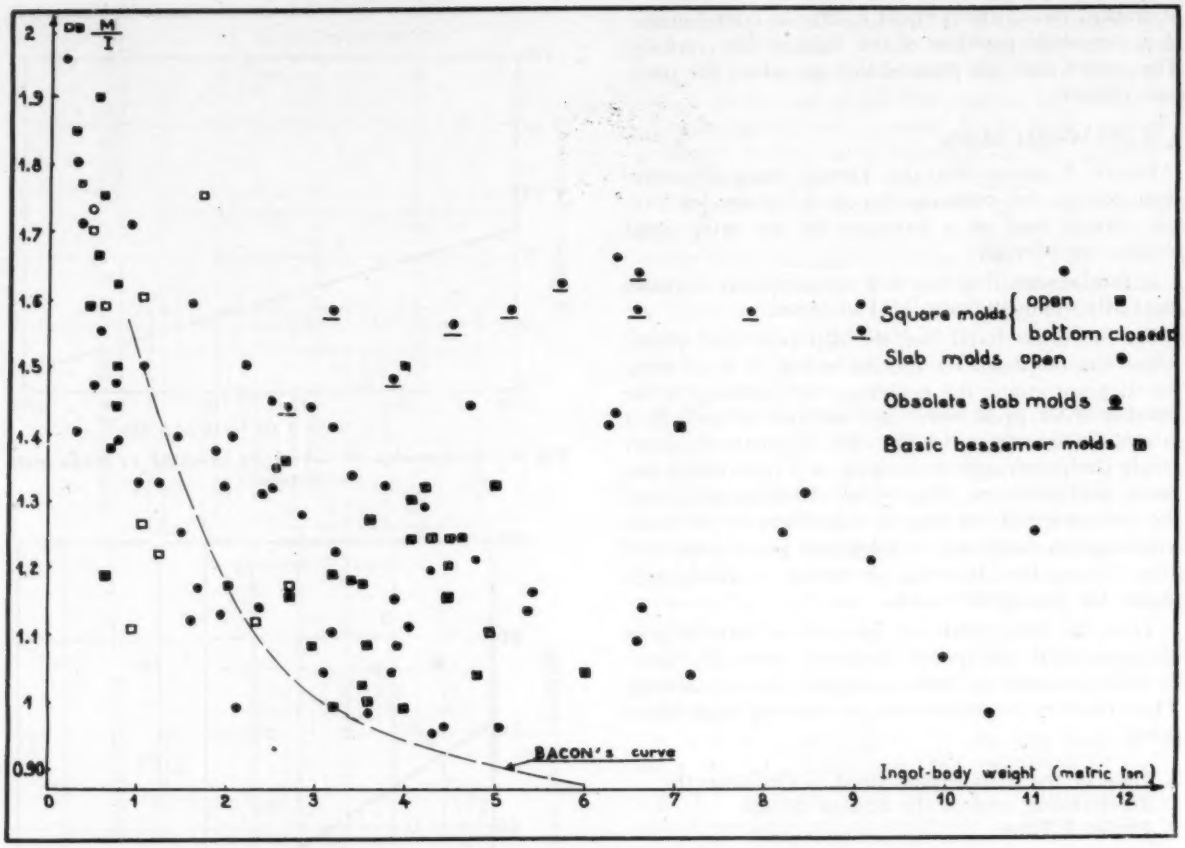


Fig. 6 — Relationship of M/I (Mold Weight/Ingot Weight) vs Ingot Weight.

III. General Observations on Cracking

1. Shape and Location of the Cracks

Mold cracks can be classified as follows:

A—**Vertical Cracks.** These cracks may be located in the head, in the foot, or in the corners of the mold. They may originate in foundry practice, in a trunnion, a stripping hollow, a relief boss acting as a preferential cracking zone.

B—**Vertical Cracks Reaching the Mold Corner.** These cracks are often a cause of premature scrapping of the mold, and are similar to some kinds of clinks.

C—**Horizontal Cracks.** These cracks do not occur frequently in square molds open at both ends. However, they may be found in some hot-topped (bottom closed) molds, cast up hill with an iron that is too cold. They may also be found in large slab ingots.

D—**Network Cracks.** These cracks appear only in molds having been used for a great number of melts. We have observed marked instances of these cracks in molds which had been cast with a particular iron.

Figure 7 shows diagrams of these various types of cracks.

2. Cracking Causes

A. **Stresses in the Molds** – Mold cracking obviously originates in stresses. For the study of mold stresses during teeming, we refer to the work of J. W. Grant:

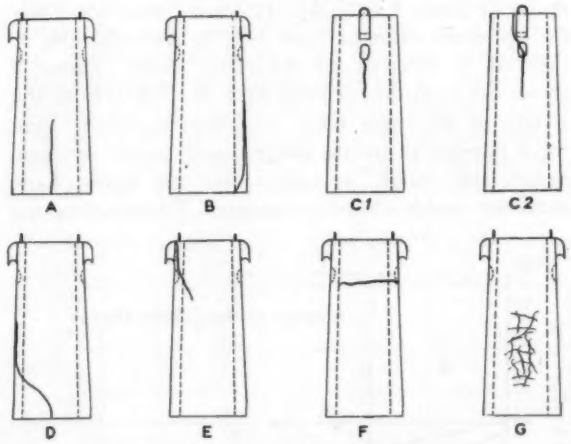


Fig. 7 — Types of mold cracks illustrated.

"Physical Properties of Mold Irons"[3] and to the work of Buttler and Glaisher[4] who have specially studied this problem. These BISRA research workers have shown that the tensile stresses undergone by the outer face of the mold are maximum about 4 minutes after teeming; they remain at least until the temperature reaches 750F (400C). These stresses take up an important fraction of the tensile strength of the mold iron (up to about 70 per cent in some cases). Under these conditions it is readily understandable that

comparatively low residual stresses may be enough to cause sudden cracking—or clink—the first time the mold is used. These stresses do not disappear when the mold cools off, and we have observed, during a qualitative study effected by means of strain-gages, and after having relieved the stresses by cutting the mold with a saw, that the inner face of a 1200 Kg bottom-closed mold with superimposed head, was in tension.

B. **Influence of Service Conditions** – During investigations on ingot-molds, we have observed (Table 1) that, at least in two plants, the hottest faces, i.e. the faces in front of one another in the mold line, showed a more marked tendency to cracking. In both cases, at least 66 per cent of cracked molds showed cracks on the hottest faces. This point is important, as it should lead to use casting pits or casting cars so designed as to make it possible to leave a good distance between the molds. Likewise, in the case of hot-topped molds, the number of molds cracked in the head has decreased when the stationary hot-tops have been replaced with floating hot-tops, as the upper edge of the mold was less hot, hence had more resistance.

TABLE 1

Steel Plant	A		B
Number of ingot-molds	97	48	51
Number of cracked ingot-molds	95*	41	50
Number of ingot-molds cracked on the hottest face	64	31	35
Percentage of ingot-molds cracked on the hottest faces	66%	75%	70%

*Disregarding the G type cracks

3. Conditions Which Promote Cracking

Following investigations in many foundries and basic Bessemer steel plants,[1] we present a non-exhaustive list of examples or particular conditions having caused cracking in square molds used in basic Bessemer steel plants.

A. *Causes Related to the Cupola Charge*

In French practice steelmaking molds are manufactured from iron and mold scrap remelted in the acid cupola. The usual charge comprizes 50 per cent of virgin hematite iron and 50 per cent of mold scrap. The main cracking causes which can be traced to the cupola charge are the following:

1. **Too High Percentage of Mold Scrap.** This fact corroborates the observation that with 100 per cent mold scrap in the charge the cracking tendency is exaggerated. The usual percentage of 50 per cent in the French practice can however be increased if the object is the manufacture of small molds, or thick-walled molds, or molds which must be particularly resistant to crazing, or in the case where the mold scrap is composed of molds which have been scrapped after a short life due to defective inner surface condition.

2. **Too High Percentage of Scrap.** Over 10 per cent scrap in the charge, cracking increases markedly. In this case, it appears that the mold structure is altered, likely through dilution of carbon or silicon or through the effect of oxides brought by the scrap. However, the resistance to crazing can be improved with additions of about 5 per cent scrap as was reported in several French foundries.[1] However by making a statistical investigation of mold clinking in various steel plants, we observed a direct relationship between the percentage of clinked molds and the use of steel scrap in the foundry where the molds were made. Among 19 foundries, of which only five used steel scrap, the four captive foundries of the steel plants which ranked first for the number of molds scrapped due to clinking used steel scrap in their cupola charge. The following relationship was observed between the clinks percentage and steel scrap charging:

	Plant A	Plant B	Plant C	Plant D
Percentage of clinks in the scrapped molds	5%	5%	3%	2.3%
Percentage of steel scrap in the charge	10%	10%	10%	5 %

It is likely that the cracking tendency is correlated with this occurrence of clinks.

3. **Charging Oxidized Scrap.**

4. **Charging Certain Qualities of Virgin Iron** — There is some doubt whether or not the synthetic hematite irons, obtained by charging pyrites-residue in the blast furnace, give worst results. On this point, experience records are conflicting from plant to plant.

B. *Causes Related to the Chemical Analysis*

1. **Total Carbon** – It appears that a total carbon content that is too low promotes the occurrence of cracks. Too high a total carbon content may have the same results. Hengstenberg, Knehans and Berndt have published a surface "carbon - silicon - mold life" which shows a maximum.[8] We have investigated for several foundries the influence of the total carbon content and of the silicon content on the cracking tendency; the results are not conclusive, it can only be stated generally that a low total carbon content and a low silicon content, on the one hand, and a high total carbon content and a high silicon content, on the other hand, give less satisfactory results.

2. **Phosphorus** – In France, the mean phosphorus content of the mold irons used in the foundries we have visited is 0.115 per cent. It seems that under 0.2 per cent, upper limit given by the British Ingot Mould Subcommittee, to avoid a phosphorus eutectic[5] no deleterious effect of phosphorus is observed. Over 0.2 per cent, the records of several foundries show conflicting results. Some molds have shortened lives and even clink, while others show satisfactory service lives.

3. **Sulphur** – The presence of sulphur is universally considered as promoting cracking; this is also true of chromium and copper. We shall give hereafter chromium and copper contents of satisfactory molds and of defective molds which have been made under the same conditions, showing that the effect of these elements considered deleterious is not so well defined as claimed by some authors.

C. *Causes Related to Molding, Teaming and Unmolding*

1. **Defective Positioning of the Mold Core** — In the tests that will be described in Section IV, we shall give, for a mold with a M/I ratio equal to unity, a limit for which it does not appear that core shift may be involved.

2. **Ill-Fed Molds** — In foundries where bottom pouring is used, when the iron is not hot enough a crust forms on the surface of the molten iron rising in the mold. If this crust is stopped by steel parts sunk in the iron or by the back draft of the pattern, inclusions are obtained which are similar to the teeming laps of steel. These inclusions may cause horizontal cracking of the mold. As a rule, it is possible to remedy this condition by increasing the number of the gates. Teeming arrests cause horizontal cracks which have a similar origin.

3. **Iron Temperature at Teeming Time** — The tests we have conducted seem to show that under a defined temperature (2200 F in the case described in Section IV), the proportion of prematurely cracked molds tends to increase.

4. **Locked-In Stresses Due to a Core Lacking Collapsibility, or to a Core Barrel, Which Has Been Removed Too Late** — This cause has been mentioned by several foundries. It seems also to be involved in some cases of molding using a cement core. In the foundries where no braided straw rope is used between the core barrel and the mold core, they have to remove the core barrel rather early, both for production reasons and to prevent the stresses caused by a tardy removal of the core barrel. The average time after which the core barrels not fitted with straw have to be removed is about 1 hr 45 min to 2 hr in the French practice.

5. **Too Early Removal of the Sand** — It seems that, except the cases where the ingot-molds are removed from the sand when they are still red hot, the cooling in the foundry sand is slow enough to prevent build-up of detrimental stresses. In particular, it does not seem that after such cooling locked-in stresses remain, that could be relieved by aging. Aging has not been observed to have an influence on the service life of the ingot molds, when the latter were drawn from the sand at about 300 F (150C).

D. *Causes Related to Ingot Mold Design*

1. **Ratio of Thickness at Corner Level/Mid-face Thickness.** In the case of square molds, it is mainly this ratio which has to be considered from a design standpoint. In the French practice, the average value of this ratio is 0.9. This value is similar to the 0.95 ratio recommended by BISRA. Ratios lower than 0.9 have often given rise to untimely cracks in the corner, unless they apply to very thick walled ingot molds. As far as light molds are concerned (M/I = 1) the trend would rather be to adopt ratios higher than 0.95, or nearly 1, to prevent corner cracks. This corroborates the BISRA's opinion since the BISRA's design leads to light ingot molds.

2. **Corner Radius** — Considering the ingot molds only as a casting, its corner radii ought to be as large as possible. However, in this case, the ingots would have a tendency to show hot cracks in the corners, although the extra-mild (very low carbon) basic Bessemer steel is not particularly sensitive to this defect. Nevertheless a compromise must be worked out between the requirements of the foundry and those of the steel plant; the medians in France for the basic Bessemer ingot-molds are the following:

upper corner radius/ mean inner side	extreme values	0.058 — 0.144
	medians	0.081
lower corner radius/ mean inner side	extreme values	0.066 — 0.161
	medians	0.091

These values are somewhat higher than the 0.06 ratio given by the 3rd Report of the Ingot-Mould Subcommittee.[6] French practice consists in adapting the ratio corner radius/mean inner side to the mold weight, giving it a value close to 0.06 only for ingot molds weighing more than 4 tons.

IV. Cracking of a 3.5-Ton Square Mold

1. Mold Features

Weight: about 3.5 tons

Average weight of the basic Bessemer rimming steel ingot: 3.6 tons

Ratio: Mold weight/Ingot weight = M/I = 0.975

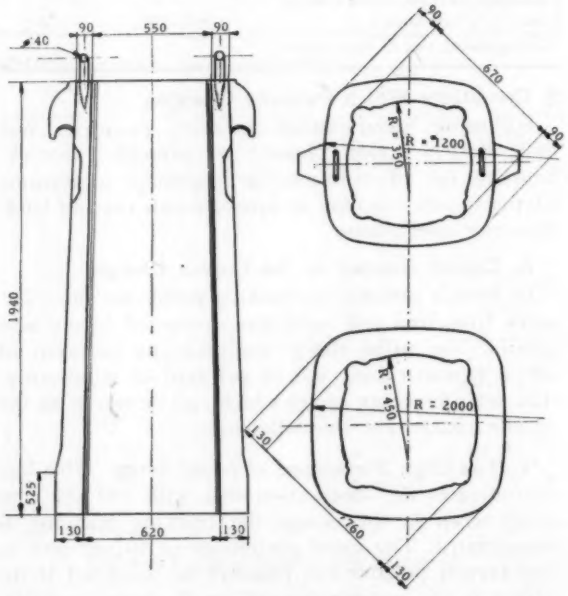


Fig. 8 — Design of M5 mold.

The design of this ingot mold is illustrated in Fig. 8. This particular mold will be referred to hereafter as mold M5. This mold has been made the subject of a particular study because during several years, thanks to favorable conditions of use, it has afforded a regularly low consumption of ingot-mold iron per ton of steel, about 7 kg (15.4 lb) per ton, which is an excellent result.

2. Manufacturing Conditions of Mold M5

As all French ingot molds, mold M5 was manufactured with cast iron melted in the acid cupola. The usual charge comprizes 50 per cent hematite iron and 50 per cent mold scrap. The coke used to melt these alloying elements is metallurgical coke (oven-coke) in the proportion of 8.4 per cent of the charge, disregarding the coke bed. Other production items: about 2.1 per cent limestone, 10.6 cubic meters (375 cu ft) of air per kg (2.2 lb) of coke, and 800 cubic meters (28,250 cu ft) of air per ton of cast iron.

The median analysis of the virgin iron used is the following: T. C., 3.8%; Mn, 0.6%; P, 0.07%; S, 0.024%; Si, 2.7%.

Median analysis of the mold scrap: T. C., 3.35%; Mn, 0.5%; P, 0.085%; S, 0.08%; Si, 1.90%.

Median analysis of the iron as-tapped from the cupola: T. C., 3.5-3.6%; Si, 1.9%; Mn, 0.45%; S, 0.07-0.08%; P, 0.085%.

The average casting temperature, measured with an immersion thermocouple (thermocouple Le Chatelier Pt-PtRh, with 13% Rh) was about 2250 F (1230 C); the temperature loss during the casting of the ingot mold was about 70 F (20 C). The gating and vent are located as shown by Fig. 9. The median casting time of the ingot mold is about 2 min, 30 sec.

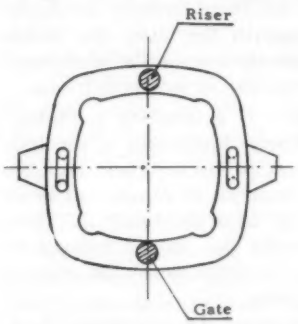


Fig. 9 — Position of riser and gate.

The ingot mold is molded using a metal pattern, ramming simultaneously the outer sand and the core. No straw rope is used around the core, which makes it necessary to remove the core barrel fairly early, about 2 to 2½ hr after the casting operation. The shakeout is done about 56 hr after casting, on the average.

3. Conditions of Use of the Ingot Molds

The ingot molds are used in casting pits to manufacture very low carbon basic Bessemer steel only. In the steel plant which uses them, there are two casting pits, each ingot mold being used all its life in the same pit. The ingot molds are spray-tarred on a tipping device after cleaning. They are used an average of two times a day, as the plant comprizes two 8-hr shifts followed by a non-production shift devoted to maintenance. On the average, the ingot molds are used every 8 hr, but for about 13 per cent of the heats the "mold-interval" is shorter.

The ingot remains in the mold for 37 min (median of 7300 measurement). The extreme values of these measurements vary between 16 and 274 minutes; 95 per cent are between 25 and 65 min.

4. Causes of Discard of Mold M5

The median life of mold M5 was 140 heats, which corresponds to a consumption of about 7 kg (15.4 lb) per ton of ingot. The main cause of discard is cracking (see Fig. 1 and Table 1). However, a number of ingot molds with a very extended service life are scrapped due to crazing. The cracks observed on mold M5 (cf Fig. 7) are the following:

Type A, i.e. a vertical crack starting from the foot and reaching the top of the mold. As a rule, this crack is initiated on the inner surface of the mold, in connection with crazings. It spreads gradually through the whole thickness of the mold, then extends to the top.

Type C. Either type C1: a crack is initiated under the stripping lugs, at the bottom of the small hollow provided for the teeth of the stripper tongs, and reaches to the foot of the mold, or type C2: the crack reaches to the top of the mold along the stripping lug and often baring the handling ring, it causes the mold to be discarded somewhat early.

Type G cracks are observed on ingot molds which have been used for a great number of heats. Type D cracks seldom occur.

5. Effects of the Molding and Use Conditions

A. **Molding Variables** — Figure 9 shows the locations of the riser and of the gate. The ingot molds are top poured and, when the mold is full, a delay is observed to fill the pouring basin which will feed the

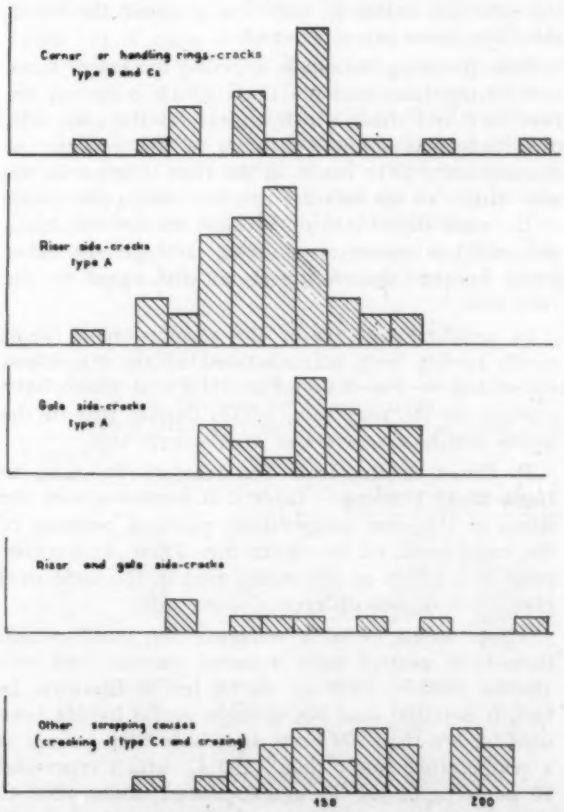


Fig. 10 — Bar graphs showing frequency and causes of scrapping of M5 mold.

TABLE 2 – INFLUENCE OF IRON TEMPERATURE BEFORE POURING ON MOLD LIFE

Cast iron temperature	Number of ingot molds	Average life (no. of heats)	Ingot molds having withstood more than 150 heats			Ingot molds having withstood less than 100 heats		
			Number of ingot	Ingot molds percentage	Average life	Number of ingot	Ingot molds percentage	Average life
			Casting pit I					
<1200 C (2200 F)	11	130.5	3	27.3	165	3	27.3	89
1200-1250 C (2200-2280 F)	50	148	22	44	180.8	2	4.5	75
>1250 C (2280 F)	6	135.6	1	17	174	—	—	—
			Casting pit I average : 144.3					
			Casting pit II					
<1200 C (2200 F)	7	134	1	14	169	1	14	93
1200-1250 C (2200-2280 F)	28	134	7	25	156.8	2	7.1	57
>1250 C (2280 F)	5	141.2	1	20	159	—	—	—
			Casting pit II average : 134.9					

casting during the solidification of the iron. No feed is provided on the riser side. At the steel plant, the ingot molds are aligned in a row in the casting pit, the stripping lugs being located in front of the pit walls. The sides which have no stripping lugs are the hottest in operation, they have either the riser or the gate.

Figure 10 shows that about 61 per cent of the molds are discarded due to type A cracks (Fig. 7) in opposite sides. This corresponds to about 70 per cent of all the ingot molds discarded for cracks. This corroborates the statement according to which the hotter sides are more prone to crack.

Now, if among the cracks affecting the hotter faces, one distinguishes between those which occur on the riser face and those which appear on the gate side, one finds that about two-thirds of the majority of discards are due to cracks in the riser side, i.e. in the side which, in the foundry practice used, corresponds to the most ill-fed portion because, on the one hand, the metal is colder on pouring, and on the other hand, because the casting is not fed again on the riser side.

In another steel plant, the proportion of ingot molds having been manufactured in the conditions illustrated by Fig. 9 and Par. III.2 and which have cracked on the riser side, is also double that of the molds which have cracked on the gate side.

B. Effect of the Iron Temperature Previous to Ingot Mold Pouring – Table 2 is a summary of the effect of the iron temperature prior to pouring of the ingot mold, on its service life. These data correspond to a group of 108 molds used in the same steel plant, but in two different casting pits.

There seems to be a tendency for molds which have been poured with a metal on the cold side (under 2200 F, 1200 C) to be less satisfactory. In fact, it is found that out of eight molds having been used for less than 100 heats, four have been poured at a temperature lower than 1200 C, which represents 20 per cent of the 18 molds poured under 1200 C. The four others represent only 5 per cent of the 90 molds poured over 1200 C. Among the 18 ingot molds which have been poured at a temperature lower than 1200 C, six molds have been discarded due to cracks of the type C2 (Fig. 7), or 33 per cent of the cold-poured molds, while only 14 out of the 108 molds, or 13 per cent of the whole, and eight out of the 90 molds poured over 1200 C, or 9 per cent show the same defect along the handling rings.

The results of Table 2 are more definite for basin I than for basin II, because in the latter the molds apt to live longer have had their duration shortened due to somewhat less favorable service conditions.

C. Effect of Core Shift – If a diagram is plotted with the abscissa as the lower thicknesses at the foot of the molds, the zone most prone to crack, and the ordinate as the mold life in terms of number of heats poured, it is found that up to a mold-foot thickness of 5 in. (128 mm), lower by $\frac{9}{32}$ in. (7 mm) than the average which is $5\frac{9}{32}$ in. (135 mm), the average duration is close to 140 heats.

If it is distinguished between the minimum thicknesses according to their location: riser side, gate side, or stripping lugs sides, no definite differences are found. It might be concluded that the light ingot molds under study are not oversensitive to core shift. Under 5 in. (128 mm), i.e. under a $\frac{9}{32}$-in. (7-mm) shift, the available data are too few to warrant a definite conclusion.

D. Effect of the Pouring Rate – We have not observed any peculiar effect of the pouring rate excepting perhaps for the colder cast irons (temperature lower than 2200 F, 1200 C), and this even though the series of 108 molds has been made by two operators, one of whom had a pouring time of between 1 min, 20 sec and 2 min, 50 sec (median 2 min, 10 sec) and poured 92 molds, and the other a median pouring time of 3 min, 10 sec.

6. Effect of Metal Analysis

A. Carbon, Silicon, Manganese – To investigate whether the foundry practice had an effect on the life of type M5 molds, a comparison has been made, in identical service conditions, between a series of 24 molds manufactured by Foundry A, and a series of 24 molds manufactured by Foundry B.

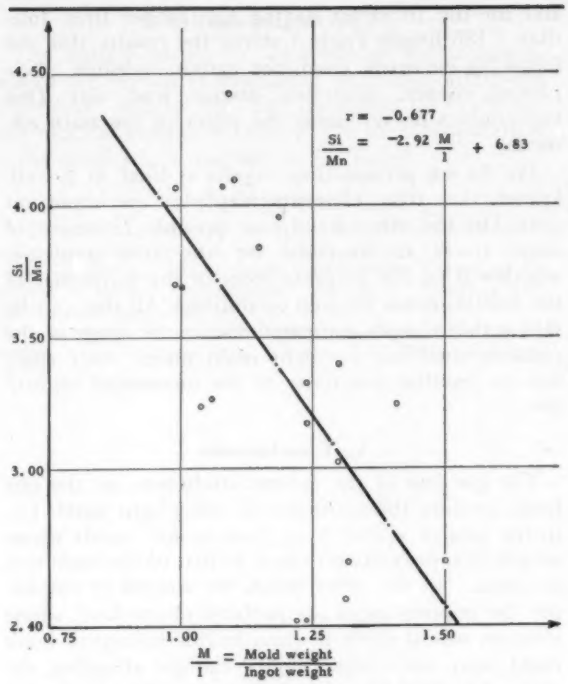


Fig. 11 — Relationship of Si/Mn vs M/I (Mold Weight/ Ingot Weight).

TABLE 3

Foundry	A	B
C total (median)	3.56	3.69
Si (median)	1.89	1.55
Mn (median)	0.44	0.63
Si/Mn (median)	4.25	2.46
Mean life	133.5	171

Table 3 shows life differences, statistically significant, between Foundries A and B. The best results have been obtained with the higher total carbon contents and manganese contents, on the one hand, and the lower silicon contents on the other. Obviously, it is not possible to conclude that this variation of analysis is the cause of the life improvement, since we have two simultaneous variables: the foundry and the analysis, but it can be seen that low silicon and higher manganese have afforded good results. This is an important point for the French practice. In fact, as shown by Fig. 11 the French foundries had a tendency to match Si/Mn ratio with ratio M/I = mold weight/ingot weight, considering that the thinner the molds the greater their tendency to crack, hence the higher their silicon content and the lower their manganese content ought to be.

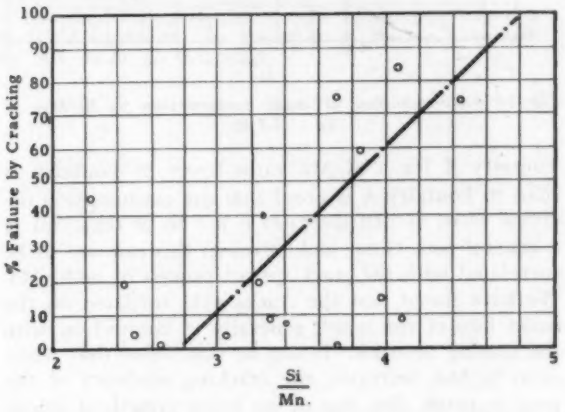


Fig. 12 — Relationship of failure by cracking vs Si/Mn.

Figure 12 gives the crack percentages as a function of Si/Mn ratio for a number of French basic Bessemer steel plants; lastly, Fig. 13 shows that the consumption of ingot mold iron decreases when Si/Mn ratio increases. A similar diagram has been published by Ristov (Fig. 14, Ref. 7). We are of the opinion that the consumption decreases when Si/Mn ratio increases is only a secondary effect of the fact that, on the one hand, light molds afford lower consumption, as already showed above, and that on the other hand light molds have the highest Si/Mn ratios.

The fact that we have obtained excellent consumption values in the cross-test between Foundry A and

TABLE 4 — DETERMINATION OF THE ELEMENTS HELD TO HAVE A DETRIMENTAL EFFECT, FOR THE BEST AND FOR THE LEAST SATISFACTORY MOLDS.

	Average life : 94 heats for 15 ingot molds Poor molds — 110 heats or less					186 heats for 16 ingot molds Most satisfactory molds — 159 heats or more				
Elements	Number of determinations	Limits L****	Limits U****	Range 87%**	Median	Number of determinations	Limits L****	Limits U****	Range 88%***	Median
Ct for memory	15	3.27	3.48	3.32 3.46	3.41	16	3.30	3.50	3.30 3.50	3.42
Cc*	15	0.76	1.12	0.85 1.10	0.95	16	0.75	1.12	0.81 1.10	0.92-0.93
S	15	0.064	0.104	0.072 0.103	0.085	16	0.068	0.106	0.070 0.095	0.078
P	15	0.066	0.088	0.068 0.086	0.075	16	0.068	0.138	0.070 0.092	0.084
Cu	15	0.13	0.20	0.13 0.17	0.16	16	0.14	0.20	0.145 0.19	0.16
Cr	15	0.042	0.059	0.044 0.056	0.050	16	0.040	0.058	0.042 0.054	0.048
As	15	0.0425	0.0475	0.0432 0.0463	0.0443	16	0.0375	0.0463	0.0390 0.0457	0.0431
Pb	15	0.013	0.028	0.013 0.020	0.016	16	0.012	0.032	0.012 0.030	0.013
Sn	15	0.0225	0.035	0.025 0.032	0.027	16	0.021	0.032	0.022 0.031	0.029

*We give here, for information purposes, the determination of the Cc for both extreme mold groups, knowing that the sampling conditions (35 mm round) and the determination make these values hazardous

**87% Range : i.e. 13 ingot molds
***88% Range : i.e. 14 ingot molds
****L = Lower U = Upper

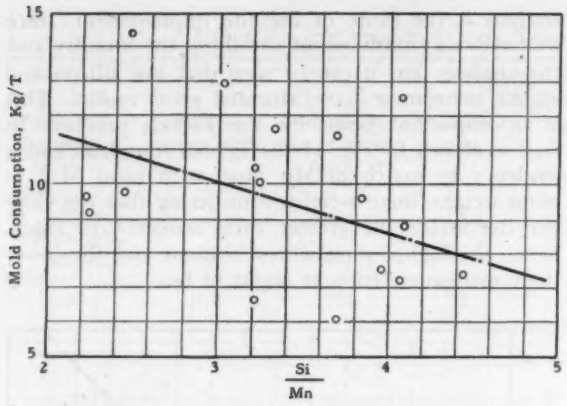


Fig. 13 — Relationship of mold consumption vs Si/Mn. 1 kg = 2.2 lb.

Foundry B for a Si/Mn ratio lower in Foundry B than in Foundry A is proof that the consumption decrease when Si/Mn increases is not to be regarded as a general fact. Good mold lives in this case are to be correlated with the crack spread process of mold M5. We have stated that the cracks were initiated on the inner face of the mold, generally in connection with the crazing network. It can be concluded that when ratio Si/Mn decreases, the cracking tendency of the iron decreases also, due to the lower growth of silicon and to the more pearlitic nature of the iron, as has been well shown by Rocquet and Olette.[9] There is, therefore, a later starting of the cracks, and their spread is not so easy, the physical strength of the cast irons being higher. Of course, this could not be indefinitely extrapolated, and if the manganese is too high or the silicon too low, the clinking tendency of the molds will increase. This also implies that there is a possible interaction between the crazing and the cracking processes.

B. Effect of the Elements Contained in Small Quantities in Ingot Mold Iron – For the series of 108 molds already mentioned, we have had the elements supposed to promote fragility determined, both for a group of molds having a poor life median (94 heats) and for the 16 molds having had longer lives (median : 186 heats) Table 4 shows the results (for the following elements: combined carbon, sulphur, phosphorus, copper, chromium, arsenic, lead, tin). Our experiments do not show the effect of the trace elements.

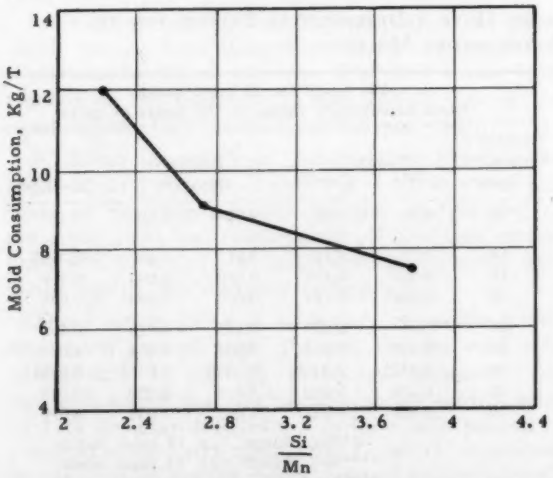


Fig. 14 — Relationship of mold consumption vs Si/Mn according to Ristow.[7] 1 kg = 2.2 lb.

We do not present these results as final. It is well-known that trace elements may play an important part. On the other hand, the possible influence of slight traces of elements on the iron structure, whether it be the graphite from or the properties of the matrix, is not yet well established. All that can be said is that roughly estimated and in the range of the contents used, for the light mold under study there was no peculiar sensitivity to the mentioned impurities.

V. Conclusions

The purpose of the present study was, on the one hand, to show the economics of using light molds i.e., in the case of square 3- to 5-ton molds, molds whose weight is approximately equal to that of the ingot cast in them. On the other hand, we wanted to emphasize the importance of the problem of cracking, whose solution would result in considerable savings of ingot mold iron, since light molds, though affording the most favorable metal consumptions are prone to cracking.

The problem of ingot mold cracking is an intricate one, involving many factors, as shown by the partial list of cracking causes given in Section II. Variations of many factors, sometimes ill-controlled, may cause an increase of the proportion of cracked molds. In the crack study, there is an important problem of details. Nevertheless, it does seem that in all steel plants it is the hottest faces, or the hottest parts of the mold which show the highest tendency to cracking. It is therefore desirable to provide over-designed casting pits and casting cars, to minimize as much as possible the heating of the mold faces. This precaution may also have the best results as to the other great cause of mold discard, crazing, whose occurrence and development is markedly favored by excessive heating of the mold.

The study of mold M5 cracking has shown that, under given service conditions, the pouring practice and the iron temperature play a part. It is important to locate the risers on the mold faces which are least stressed in service, and not to pour the ingot molds at a temperature lower than 2200 F (1200 C), to minimize the number of premature discards and that of stripping lug cracks.

As far as analysis is concerned, it is possible to make with a conventional charge containing 50 per cent mold scrap, excellent molds having lower Si and higher Mn than was required, in agreement with German results,[7] by the usual French foundry practice. On the contrary, our knowledge is still inadequate, or the ingot mold manufacture is not yet sufficiently under control, to define clearly the effect of certain trace elements in cast iron, as well as the influence of quality variations of virgin iron. It is possible that later progress brings this latter opinion into question again.

Acknowledgments

The author wishes to acknowledge his indebtedness to the French "Commission technique mixte des lingotières" and to the "Etablissements de WENDEL" and "Société LORRAINE ESCAUT" for their cooperation in the work which has led to his paper.

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MOLDING SANDS, MOLDING METHODS AND CASTING DIMENSIONS

By

R. W. Heine*

New molding processes have received a great deal of attention in the past few years. Shell molding, D-process, pressure molding, CO_2-process, and diaphragm molding, to name a few, have been widely heralded as providing a new era of progress in molding. Amid the claims and accomplishments of the new methods, it is worthwhile to take a new and critical look at an old process, green sand molding.

What are the true capabilities of the process? How well is it now being used relative to its ultimate possibilities? What can be done to get the best results from it? Complete answers to these questions cannot be given for lack of certain fundamental information. However, the general direction of improvements can be discussed.

To get the best green-sand casting, the best mold possible must be produced from a combination of molding sand and molding equipment. This mold should have uniform sand density, high hardness, high strength and thermal stability. Inevitably certain characteristics of the sand mixture will affect the ability of the process to obtain such a mold. Therefore, consider first the molding sand.

Density of Sand in the Mold

Green sands are fundamentally a mixture of sand grains, clay and water. When clays are added to sand, and the mixture is tempered with water, there is a large reduction in density of the mixture. For example, a raw silica sand may weigh 100-110 lb per cu ft while a freshly aerated molding sand may weigh as little as 40 to 50 lb per cu ft. The magnitude of the reduction in density depends on the amount of clay, type of clay, water content, and additives in the mixture. It also depends on sand processing such as, mulling, aeration, and riddling. Furthermore, actual sand density on a conveyor, elevator, in a sand hopper, or in a flask depends on how much work is done toward compacting the sand from its initial low density.

The magnitude of these changes is partially illustrated in Fig. 1 in terms of the density of standard 2.0-in. x 2.0-in. diameter sand specimens as related to number of rams with the standard AFS rammer. Density of the raw sand, no clay or water added, is shown on the upper-most curve. Density of the tempered mixtures is reduced as shown by the middle two curves. But, only about 90 per cent of the mixture is sand grains, the balance being clay and water. The sand grain fraction in the rammed specimens therefore has very much lower density than that of the raw silica sand. This is an important observation.

Clay and water hold the sand grains apart. Water causes the clay to swell, become sticky and thus reduce sand density as shown in Fig. 2. Now the density shown in Fig. 2 occurs after the compaction caused by one AFS ram. Density of the sand as mixed and aerated is likely to be only 50 per cent of that shown in Fig. 2. The importance of moisture control of molding sand is emphasized by Fig. 2. Just as important, though not shown in Fig. 2, is the need for clay content control. Low bentonite content sands

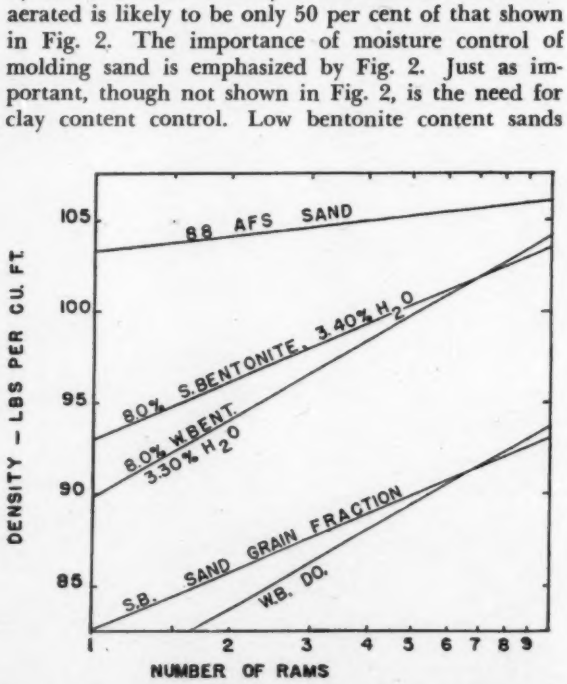


Fig. 1 — Comparison of densities of 2-in. x 2-in. diam specimen of raw sand: sand, clay, water mixtures: and sand grain fraction in mixture as a function of ramming.

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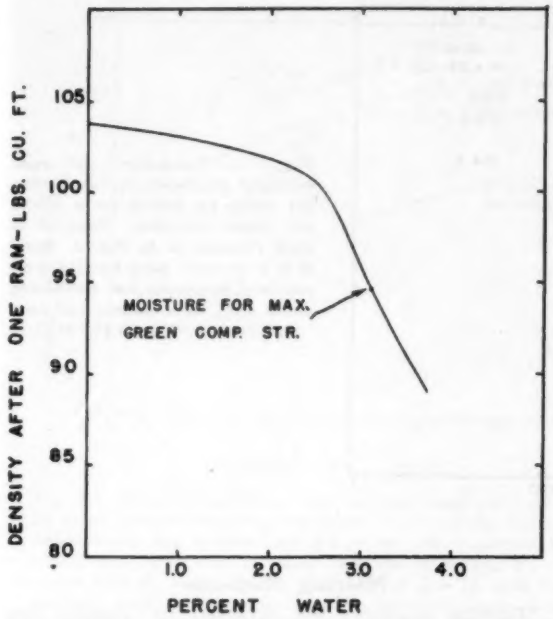


Fig. 2 — Effect of moisture control on density after one ram with standard AFS rammer: 8% southern bentonite, 89 AFS GFN, 4-screen sand. A 14.2% decrease in density is caused by addition of 3.70% H_2O.

will show less decrease in density at a given per cent water than will high bentonite content sands.

Clay and water hold the sand grains apart. But the object of molding is to push sand grains together, and thus develop higher density and strength. Molding accomplishes compaction of the sand. The sand density in a mold occurs at some level between the minimum of freshly mixed and aerated sand and the maximum possible. Maximum density is achieved if there is sand grain to sand grain contact throughout the mass. This maximum density is never reached in normal molding.

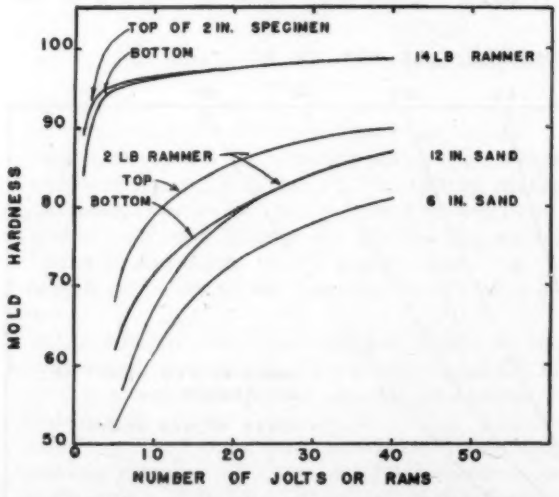


Fig. 4 — Effect of number of jolts with 3.0-in. jolt stroke on hardness at parting and effect of number of rams with 14- and 2-lb weights in AFS rammer on hardness of top and bottom of 2-in. x 2-in. diam specimen. Same sand as in Fig. 3.

Let us consider the densities actually produced by ramming and jolting in an 89 AFS, 4-screen sand, bonded with 8.0 per cent southern bentonite, and 3.40 per cent H_2O. Figure 3 shows the sand density reached by ramming 2.0-in. x 2.0-in. diameter specimens with 14-, 4-, and 2-lb weights dropped 2.0 in. in the AFS rammer. Figure 3 also shows the density reached by jolting 12.0-in. and 6.0-in. sand heights on a standard 3.0-in. jolt stroke molding machine. Note in Fig. 3 how actual density depends on the molding device and its power.

Ramming appears much more powerful than jolting in ability to compact the sand: especially the 14-lb rammer weight. The effect of the same molding actions on mold hardness of the sand is shown in Fig. 4. The same effects are evident. Mold hardness is a convenient means of measuring density indirectly. Any given sand will show a definite relationship of mold hardness and density — high hardness meaning high density. However, different sands should not be compared at equal mold hardness especially in comparisons of bentonite and fireclay bonded sands.

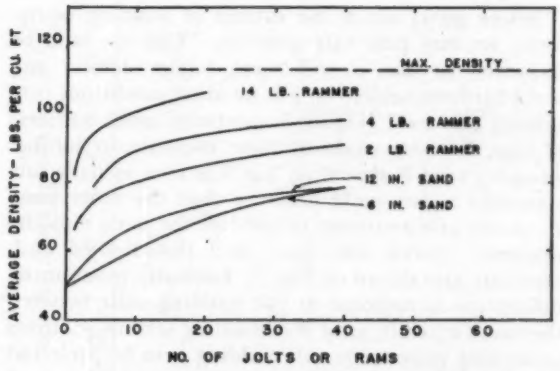


Fig. 3 — Relation of average sand density to number of rams with 14-, 4-, and 2-lb weights in AFS ramming device and to number of jolts with 12-in. and 6-in. sand columns on a 3.0-in. jolt stroke machine. Sand is an 89 AFS GFN, 4-screen sand, bonded with 8.0% southern bentonite, 3.40% H_2O.

Figures 3 and 4 show that the jolting action of a common 3.0-in. jolt stroke machine compacts this sand to low density and only fair mold hardness relative to maximum. It should be no surprise to have mold cavity enlargement occur when the casting is poured. Over-size and over-weight castings result from low mold hardness associated with poor compaction of the sand. Fundamentally, the molding process always puts sand in the mold at some density between the minimum and the maximum. The only question is where will the mold density be between the low and the high?

The success of any green sand molding process may be gauged by the degree to which it compacts the sand toward the optimum conditions of high mold hardness and density. It is inherent in the sand to develop low density and hardness. It is a requirement of the molding operation to develop high hardness and density for the sand involved. Molding machines differ greatly in their ability to compact the sand

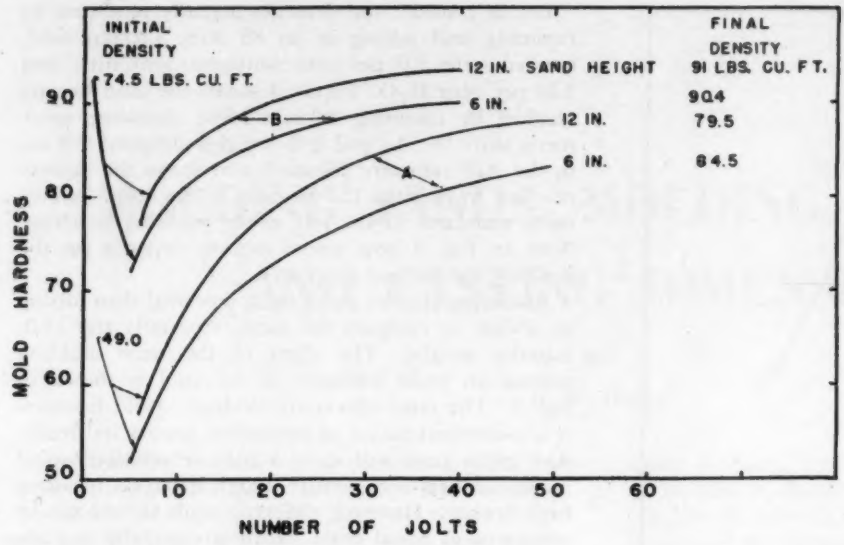


Fig. 5 — Comparison of mold hardness produced in two different sands by jolting on a 3.0-in. jolt stroke machine. Sand A is sand referred to in Fig. 3. Sand B is a foundry sand bonded with southern bentonite but containing about 2.0% of a commercial sand additive; both at 3.4% H_2O.

sufficiently to get a hard mold and accurate dimension castings.

Before going on to the subject of molding equipment, we may pose this question. "Can the sand be improved so that it will reach higher density and mold hardness under the jolt molding conditions considered this far?" Figure 5 compares mold hardness of two different sands in their response to jolting. Foundry sand B shown on Fig. 5 is seen to have substantially higher mold hardness than the other sand, A, at any given number of jolts on the same molding machine. Initial and final sand densities of both sands are also shown on Fig. 5. Evidently pronounced differences in response to jolt molding exist between the two. Certainly, sand B containing certain additives promoting response to jolt molding is to be preferred over sand A on any jolt machine. The molding sand then is a contributive factor in reaching high mold hardness and density on any particular molding machine.

Molding Machines

Molding machines are expected to compact low density sand mixtures sufficiently to obtain the mold hardness and density necessary for making good castings. Work and power are required to compact the sand into a strong hard mold. For jolt molding, hardness and density developed depends on the effectiveness of the jolting action. Length of the jolt stroke and the weight of the sand determines the power applied from momentum of the sand at the instant of impact. The jolt stroke is a mechanical characteristic of the machine. Sand weight per unit area of projected parting surface depends on density and height of sand in the flask at the start of jolting. Thus initial density affects the power offered by the machine.

The relationship of jolting power to jolt stroke and sand weight is shown in Fig. 6. Because of the differences in power shown in Fig. 6, mold hardness and density of a given sand would be expected to vary

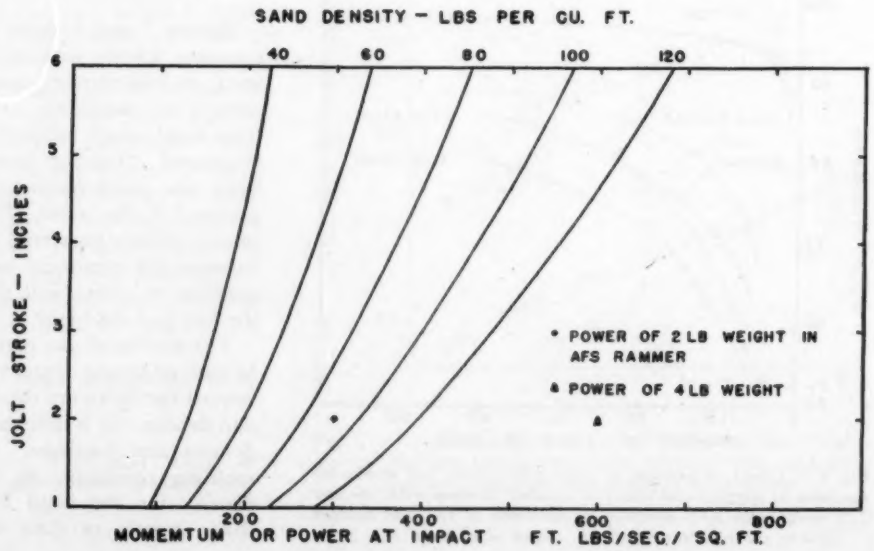


Fig. 6 — Power applied to base of a 12-in. sand column as a function of jolt stroke. Multiply by x ÷ 12 for column x inches high.

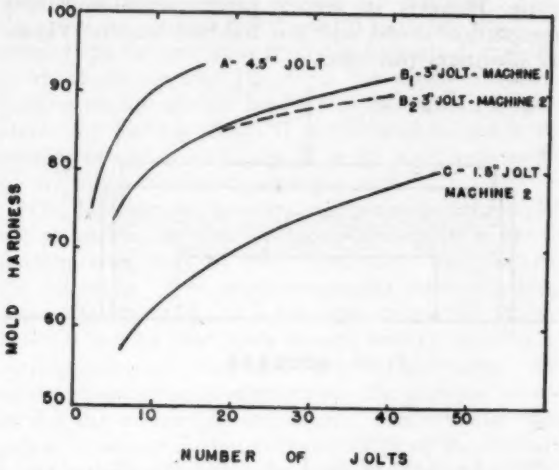


Fig. 7 — Effect of number of jolts on mold hardness produced by three different molding machines; A, B, and B_2. Effect of reduction of jolt strokes from 3.0 in. to 1.50 in. caused by a 220-lb jolt table load on mold hardness is shown; B_2 compared with C. Foundry sand C used here is a 72 AFS GFN, 4-screen sand, bonded with western bentonite, carbonaceous additives, tempered at 4.5% H_2O.

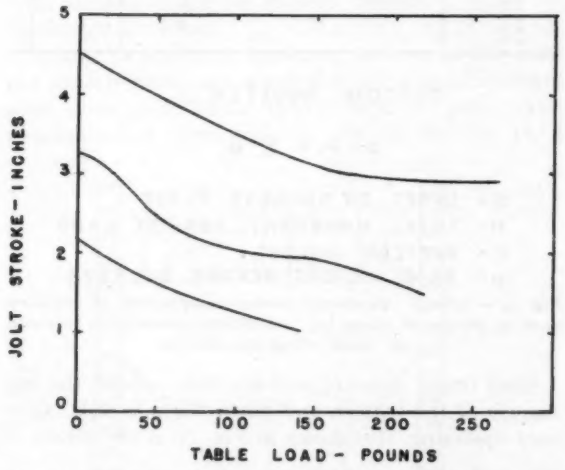


Fig. 8 — Jolt stroke vs table load for three different molding machines.

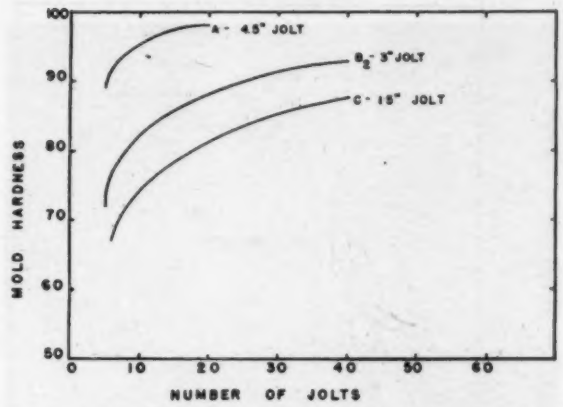


Fig. 9 — Effect of number of jolts on mold hardness produced by two different molding machines, A and B_2. Same machines as in previous graphs. Effect of reduction of jolt stroke from 3.0 in. to 1.5 in. by a 220-lb jolt table load is shown; B_2 compared with C. Foundry sand C used here is a 4-screen sand bonded with southern bentonite, carbonaceous additive, and tempered with 4.0% H_2O.

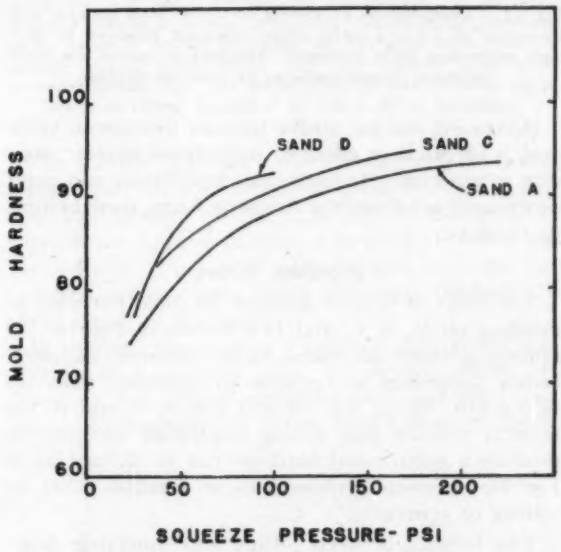


Fig. 10 — Effect of squeeze pressure applied to 2.0-in. x 2-in. diam specimen on mold hardness.

between machines having different jolt strokes. Included on Fig. 6 is the ramming power of a 2.0-lb ram weight dropped 2.0 in. on a 2.0-in. x 2.0-in. diameter specimen. Power supplied by the 2-lb weight is of similar magnitude as that of a jolt molding machine. Figure 6 also shows that a shallow flask, low sand height, will result in less ramming power than a deep one.

The effect of three different jolt strokes on mold hardness is shown in Fig. 7 for foundry sand C. This sand is a western bentonite bonded sand containing additives to increase response to jolting. Curves A and B, in Fig. 7 show data obtained on two different molding machines of 4.5-in. and 3.0-in. stroke respectively with a 20-lb load on the jolt table. Curves B and B_2 were obtained from two different makes of machines, both with 3.0-in. jolt. Curves B_2 and C were obtained from the same machine at jolt strokes of 3.0 and 1.50 in. respectively. In this case, the jolt stroke was reduced from 3.0 to 1.5 in. by putting 200 lb additional weight on the table, to simulate a flask, pattern, and sand load. The effect of table load on jolt stroke of this machine is shown in Fig. 8.

It is quite evident from Figs. 7 and 8 that it is inherent in the machine to produce low mold hardness when the table is over-loaded. The sand should of course not be blamed for a condition peculiar to a molding machine. However, sands of superior response to jolting produce higher mold hardness even though the table is overloaded. In Fig. 9 is shown another foundry sand, D, bonded with southern bentonite, special additives, and a lower per cent of H_2O. In Fig. 9, curve C for the 3.0-in. machine, B_2, shows much higher mold hardness at a given number of jolts when compared with curve C in Fig. 7, even when the jolt table is loaded with 220 lb.

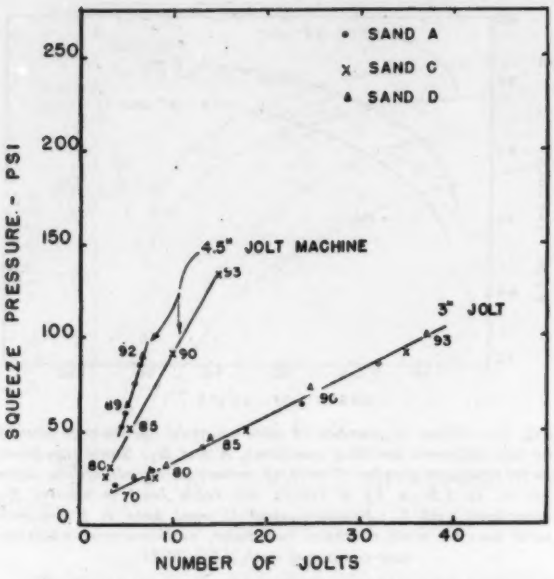


Fig. 11 — Combination of jolting of 12-in. sand column and squeezing of 2.0-in. x 2.0-in. diam specimen required to produce equivalent mold hardness. Numbers on curves are mold hardness. Same sands as in previous graphs.

Decreasing the jolt stroke with an increase in table load is identical in effect to inadequate squeeze pressure in squeezing. In both cases low density and mold hardness result from the machine being used beyond its capacity.

Squeezing Action

The effect of squeeze pressure on mold hardness of molding sands, A, C, and D is shown in Fig. 10. As squeeze pressure increases, mold hardness increases. Minor differences in response to squeezing show up in Fig. 10. Using Fig. 10 and Fig. 5, 7, and 9, the squeeze pressure and jolting conditions necessary to produce a given mold hardness can be defined as in Fig. 11. A given hardness can be reached either by jolting or squeezing.

The balance between jolting and squeezing determines which will establish the final mold hardness. For example, suppose a mold hardness of 90 is reached with 6 jolts on the 4.50-in. machine. Then, a squeeze pressure over 75 psi is required to produce any further increase in density (and mold hardness). In this case, a squeeze pressure of 50 psi would do little more than flatten and harden the squeezed surface. No increase in hardness could be expected at the parting surface. Similarly on the 3.0-in. jolt machine in Fig. 11, a squeeze pressure of 65-70 equals 25 jolts. It is a common experience in the foundry that squeezing sometimes increases mold hardness after jolting and sometimes does not. The reason is indicated.

Compaction by squeezing alone involves inherent limitations due to the manner of movement of a sand aggregate under pressure. The geometry of pattern and flask determine the extent of these limitations. When a simple block type pattern is squeezed with a flat squeeze plate, high density is achieved easily between the top of the pattern and the squeeze plate. However, on vertical pattern walls insufficient compaction occurs and low hardness results. Figure 12 illustrates this effect.

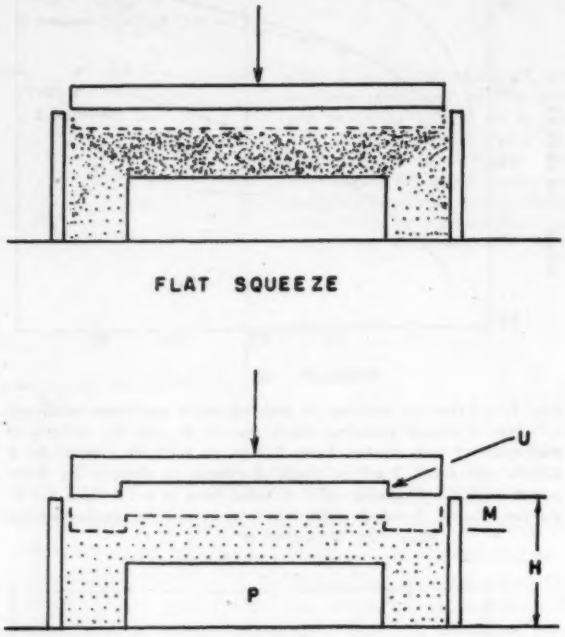


Fig. 12 — Sketch illustrates heavy compaction of molding sand in horizontal plane but insufficient compaction in vertical plane along the pattern.

Sand under pressure will not flow around the top corners of the pattern and down the side wall. Contour squeezing, also shown in Fig. 12, is one means of

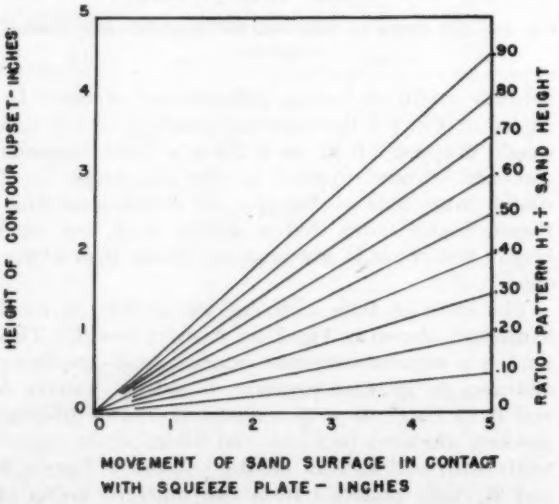


Fig. 13 — Relation of height of contour upset on squeeze plate to total movement of squeeze plate against sand.

producing more uniform density and hardness. The design of the contour board needed to obtain uniform density can be predicted from the formula U = P/H x M shown on Fig. 12. U is the upset or raised portion on the squeeze board, P is the pattern height above the parting plane, H is the sand height from parting to top of sand, and M is the total movement of the squeeze board against the sand.

The formula can be set up graphically for simple use as on Fig. 13. For an example, consider a block pattern 4 in. high in sand 8 in. deep, like that in Fig. 12, and a 2.0-in. movement of the contour board. Then, P/H = 0.50, M = 2.0, and from the graph U = 1.0 in. An intricately shaped pattern requires a correspondingly contoured board depending on squeeze movement. Unfortunately one problem arises in the use of the contour board. The contour employed is correct for not only one value of M, contour board movement against the sand. If M changes, due to changes in initial density of the sand in the flask, then the contour board will not give uniform density. Molding with the diaphragm squeeze might possibly overcome this problem. However, all straight squeezing operations are faced with the same problem of non-uniform density caused by the build-up of high sand density cones in confined spaces.[1] Figure 14 illustrates this effect.

A high density cone builds-up in front of the moving squeeze plate and develops a sidewise compressive component that binds against vertical walls. The appearance of these cones is revealed in Fig. 15, a

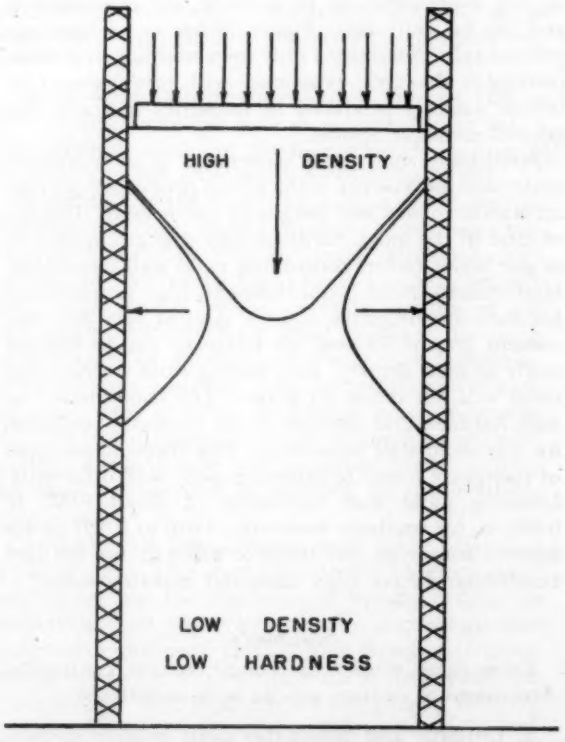


Fig. 14 — Schematic diagram illustrating buildup of high sand density cones and sidewall reaction from squeezing in a restricted volume such as a deep pocket, between pattern and flask, between patterns, etc.

Squeezed Surface



Fig. 15 — AFS 2.0-in. x 2.0-in. diam specimen formed by squeezing and then broken in compression in universal strength testing machine. Note high density sand developed under squeezed surface and lower density away from squeezed surface. Under compression test, high density cone under squeezed surface has resulted in shearing and sidewise movement of cones developed at sidewalls of specimen.

2.0-in. x 2.0-in. diameter squeezed specimen broken in compression in the Universal sand strength tester. The high density cone seen in Fig. 15 builds up in front of the squeeze plate in squeezing. Friction and the sidewise component of force prevents the full squeeze pressure from being delivered through the entire depth of the sand.

In confined spaces, then a hardness differential may develop from squeeze plate to parting. Figure 16 illustrates this effect is dependent on sand depth and size of confinement in 2.0-in. and 4.50-in. diameter

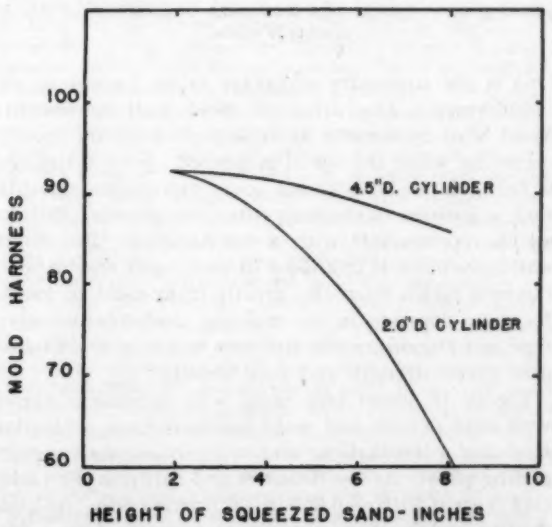


Fig. 16 — Effect of height of cylinder of sand after squeezing on mold hardness at parting at constant squeeze pressure of 110 psi using sands C and D.

cylinders. Sands squeezed are same as referred to earlier. Note the drop in mold hardness at the parting as depth of sand increases. Examples of this in molding are confined spaces such as between pattern and flask, between patterns, deep pockets, green sand cores, etc. Often, pattern drawing problems arise because the sidewise force component from squeezing either grips the pattern with a wedging action or bends it side ways to create a backdraft.

Straight squeezing alone always must contend with problems resulting from the principle of high density sand cones and sidewise force component. Jolting, however, does not produce this effect. In jolting, the dense sand at the parting is stationary and the loose sand above it moves toward the denser sand. The lower density sand comes to rest when the sand has reached the maximum density that can be produced by the power of the particular jolt action applied. Thus, the combination of jolting and squeezing is sound if they are properly balanced in power applied.

Dimensional Accuracy, Casting Size and Weight

Molding sands and machines properly used can compact the sand to optimum density and hardness. How does this affect casting dimensions? Figure 17 shows two important factors involved whenever the dimensions of a casting are compared with the dimensions of the pattern from which that casting is made.

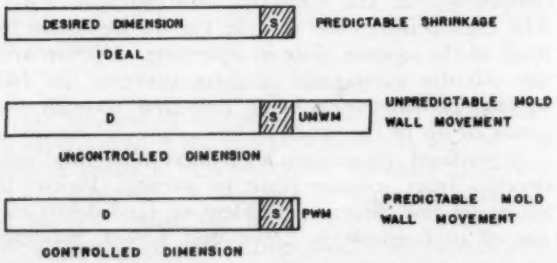


Fig. 17 — Important factors involved whenever dimensions of a casting are compared with dimensions of pattern from which casting is made.

One is the customary shrinkage factor known to all foundrymen. The other is mold wall movement. Mold wall movement is enlargement of the cavity occurring when the metal is poured. Now shrinkage is fairly dependable unless some extraneous variable such as gasiness of the melt enters the picture. Shrinkage is reproducible with a normal melt. But mold wall movement is unknown in most cases and in soft-rammed molds fluctuates greatly from mold to mold. Its value depends on the molding sand mixture, clay type and amount, water and very much on mold hardness, green strength and sand density.

Figure 18 shows how mold wall movement varies with sand density and mold hardness for a particular sand-clay combination and aluminum and copper casting alloys. At low densities and hardness the variation is great with + 0.060 in. occurring easily. At high density and hardness this figure is reduced to a few thousandths of an inch. In other words, the mold is dimensionally stable and controlled. Therefore the

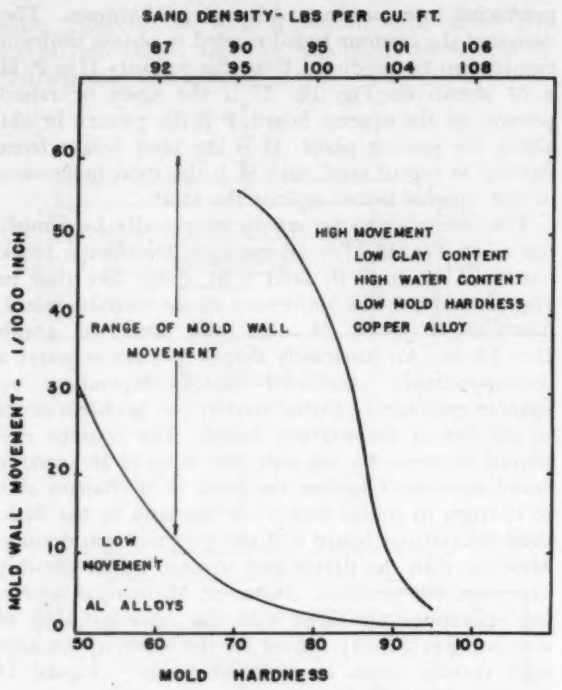


Fig. 18 — Effect of mold hardness on mold wall movement caused by Al-base and Cu-base alloys in molds made of 89 AFS GFN, 4-screen sand, bonded with 5 to 15% southern bentonite and 2.5 to 6.6% H_2O. Sand density depends on clay and water combination.

casting dimensions can be accurate and reproducible and can be controlled. Figure 18 has so far been experimentally determined only for aluminum and brass castings.[2] However, experience with measurement of ferrous castings produced in foundries indicates the general principle applies.

Predictable mold wall movement (also minimum mold wall movement) is the key to producing castings accurately to size and weight in green sand. Density of sand in the mold, hardness, and strength appear to be the major factors controlling mold wall movement. High density alone is not sufficient, high strength and hardness are necessary, too. A sand of very low clay content (raw silica sand for instance) can be molded easily to high density. But, having little strength the mold will not retain its shape. The combination of high hardness and strength is one necessary condition for low mold wall movement. The three major types of clays were found to influence mold wall differently. Limiting mold wall movement of about 0.002 to 0.005 in. for southern bentonite, 0.010 to 0.020 in. for western bentonite, and 0.020 to 0.030 in. for fire clay bonded sand have been measured experimentally.[2]

Summary

From these studies the conditions for controlling dimensions of castings appear to be as follows:

1. Uniform and reasonably high relative density. Uniform density is needed to obtain uniform and predictable expansion when the sand is heated. High relative density is needed to develop the hardness and

strength obtainable from the sand. High relative density does not mean high density per se, but rather high density relative to the maximum for the particular sand. The sand in the mold should be about 94 per cent or more of the maximum density obtainable.

2. High green strength and hardness. High density alone is insufficient. A low density sand will have low mold wall movement if the low density is accompanied by high hardness and strength. Values are difficult to assign but a mold hardness of 90 and over is suggested. A standard AFS green strength test value of over 15.0 psi compression is a minimum desired and much higher values are preferred but this depends in part on moisture level. Clay content of the sand is exceedingly important and should be high enough to produce the necessary strength. However, a high green strength sand cannot be used on under-powered molding machines.

3. The sand should possess reasonably high initial density in order to get the most power from jolting. Over 60 lb per cu ft is suggested.

4. A deep flask, with extra sand at the bottom is preferred to get maximum power for jolting.

5. The sand should possess optimum response to jolting and squeezing so that the necessary density hardness, and strength can be reached most readily.

6. The patterns should be of good quality. Backdrafts, nicks, dents, loose mountings, etc. cannot be permitted if molding is done with sufficient power to obtain high density molds.

7. The sand should be molded at the lowest moisture content possible so that loss of density from excess moisture is avoided. The temper point for maximum green strength is an objective for moisture control.[3]

Three years of experience with a sand of 16.0 to 20.0 psi green strength have shown the author that a high strength sand can produce superior castings if properly handled on suitable molding machines.

In addition to the material specifications above, the molding equipment should produce the necessary uniform and high relative density mold. Fundamentally, jolting and squeezing appears to be a sound combination. High quality results, accurate castings in size and weight, are presently being obtained by using current types of machines at their correct capacity. Overloading of machines both for jolting and squeezing is the most common mechanical cause of oversize and overweight castings. Almost as common are sand practices which do not permit highest relative density and mold hardness to be reached on the available equipment.

Consider again the questions posed at the beginning of the paper. The true capabilities of green sand molding include the possibility of making castings of predictable, and therefore accurate, dimensions and weight. For the most part, current practices are not producing castings that measure up to the capabilities of the process. This is largely because present molding equipment does not generate sufficient power in jolting or squeezing to produce the density and hardness of high strength molding sand needed for accurate castings. However, improvements in molding machines based on the principles presented here suggest themselves. Such machines are being built. With improved equipment, the best green sand castings are still to be made.

Acknowledgment

Data presented were determined by the author as part of a consulting program for the Hill and Griffith Co., Cincinnati, Ohio and are published by their permission.

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DISCUSSION

Chairman: E. C. TROY, Riverton, N. J.

Co-Chairman: W. R. MOGGRIDGE, Ford Motor Co. of Canada, Windsor, Ont., Canada.

Secretary: H. F. BISHOP, Naval Research Laboratory, Washington, D. C.

T. E. BARLOW[1] *(Written Discussion):* The work of R. W. Heine is a tremendous step in the right direction. The full possibilities of green sand molding will be realized only as we give proper recognition to the limitations imposed by improper use of equipment and materials. Too frequently we assume that a jolt-squeezer is a jolt-squeezer. The effects of over-load, improved sand, incorrect rigging, etc., are ignored more often than not. How often we see the sand technician trying to stop a swell or strain with sand when actually the problem is too large a pattern for the flask or for the machine being used. The foundry industry is literally throwing money away every day in oversized castings which could be eliminated with proper sand and equipment.

Pressure molding is green sand molding — or a phase of it. We have carried out a lot of work on that approach to the problem. We can confirm and praise all the work in this paper but no two people place the emphasis in the same place when looking at a collection of data. I would like to discuss the emphasis of certain phases of the work in four sections.

1. In many places, the author infers that density and hardness are not always related. For example "Different sands should not be compared at equal mold hardness. . . ." "A sand of very low clay content (raw silica sand, for instance) can be molded easily to high density. But, having little strength (he could have added; practically no hardness), the mold will not retain its shape". Again, "A low density sand will have low mold-wall movement if the low density is accompanied by high hardness and strength." In spite of these statements, however, the author does say, "Mold hardness is a convenient means of measuring density indirectly."

I believe such a statement, although true when qualified as Heine has done, can be misconstrued and misleading. In fact, in a few cases, the author has drawn conclusions on density when only hardness tests were used with no evidence of actual density tests. Mold strength is the only true resistance to mold wall movement. It is a combination of density and strength. Mold hardness is merely a measurement of *surface* strength and as such, can mislead a foundryman into believing that his hard mold guarantees low mold-wall movement.

The author recognizes the difference but I believe more emphasis is needed to avoid any possible misuse of the facts.

2. In the discussions of density in this paper, and later in mold wall movement statements, the author does not emphasize the possibility of nonuniform density. I am thinking particularly of the rather common mold which has a hard mold face (from jolting) and a hard upper surface from low pressure squeeze, but low density and hardness in between. Such a mold can and will show mold wall movement. The author is working with laboratory conditions in which densities are relatively uniform. Some of the problems which are only minor in the laboratory become all important in a com-

1. International Minerals & Chemicals Corp., Eastern Clay Products Dept., Chicago.

mercial mold. Overloaded equipment becomes the average condition rather than the exception.

3. Along the same line is the work on squeeze as related to jolt. The author mentions the need for high pressure squeeze to accomplish anything after hard jolting. This work should be printed in red ink. We have found that the only way to consistently develop a dense mold in most commercial molds is for the squeeze to overshadow the jolt. If it does not, a non-uniform density develops and the foundryman develops mold wall movement to say nothing of scabs, buckles, rat-tails and what have you. As an example, we just finished some work with 95 psi diaphragm squeeze after other methods of molding. We used a 12 psi green strength sand which is limited to about 88 mold hardness. The following table shows the effect of various prepack conditions.

Molding Method	Horizontal Hardness	Vertical Hardness
Diaphragm only	84 - 86	84 - 86
35 psi flat squeeze followed by diaphragm	65 - 80 84 - 87	60 - 71 83 - 86
80 psi flat squeeze followed by diaphragm	75 - 85 85 - 88	75 - 81 83 - 87
5 jolts (3 in.) followed by diaphragm	42 - 55 83 - 86	26 - 38 83 - 84
Low pressure blow followed by diaphragm	no reading 84 - 86	no reading 83 - 86

Only when the final diaphragm squeeze was sufficient to overcome the pre-pack could we stop mold wall movement in this 12-in. x 16-in. mold, 9-in. deep using a 4-in. x 4-in. x 9-in. step block pattern.

4. The fourth point of emphasis is again related to the preceding one. The author was forced to use typical laboratory equipment and I would like to throw the emphasis on the commercial implications of his valuable data. Notice in Fig. 16 the drastic effect of changing from 2-in. cylinder to a 4.5-in. cylinder. This points beautifully the reason that very early in our high pressure molding studies, we dropped the 2-in. specimen for density conclusions. We could never develop the results in the laboratory that we could get in even a small flask in practice. The only explanation we could give was the drastic effect of side-wall friction. The standard specimen is cursed with one of the worst conditions we could find; much more than the foundryman would permit in practice.

In the 2-in. specimen (2-in. high), we have a surface friction of over 12 sq in. with a sand volume of only 6.28 cu in. In a 4-in. diameter specimen, the ratio is 1 to 1 (25 to 25). In other words, the specimen tube we use in the laboratory is equal to a pattern 2-in. high, only 1-in. from the flask wall. One does not get a good squeeze or mold density under these conditions. The 4.5-in. cylinder is comparable to the surface to volume ratios recommended in good commercial practice. Under these conditions, it is questionable whether or not jolting can be effective unless followed by squeeze of sufficient magnitude to overshadow the original density. Work with diaphragm squeeze indicates that the initial jolting under these conditions has no value except to pre-pack or fill the flask.

We recently finished a series of tests with commercial equipment to find the limiting strength above which we could not go without sand bridging (the author's Fig. 14). We kept moving the pattern closer and closer to the flask until we reached a point where we could develop full density (no bridging) with strengths up to 8.5 psi in ordinary synthetic sand or 14 psi with special high-pressure additives. Beyond these strengths, bridging occurred either with or without jolting.

I am particularly happy about this paper because it is one of the few that confirms our work and statements on desirable sand properties. The ideal sand for low mold-wall movement combines high strength, low moisture, high toughness and high flowability (to get high density). Such a sand will permit the foundrymen to make a really fine product in green sand when molded with sufficiently high squeeze pressures to develop uniform density.

MR. HEINE *(Reply to Mr. Barlow):* For a particular sand, mold hardness is reasonably proportional to density and to strength. The statements in the paper are therefore valid. Non-uniform density is almost guaranteed whenever the average density is low. The standard 2-in. x 2-in. specimen is a reasonable specimen for testing purposes but it does not duplicate the conditions currently found in molds with rare exceptions.

F. S. BREWSTER[2] *(Written Discussion):* The author's subject is certainly timely since many of us seem to be making an especial effort to improve sand castings. The writer has always been in favor of exhausting the possibilities of green-sand molding improvements as well as investigating newer processes; as the author states, uniform density, high strengths, and thermal stability are essential for best results. The caution to watch density as a percent of that obtainable rather than as an absolute figure is well taken.

One way to achieve these results is as suggested to use a flexible squeeze board capable of adjusting itself to all patterns. This is now being done commercially and at the same time eliminating the noise and jar of the jolt which has been one of the undesirable factors in making the foundry a rough place to work.

It is hard to see how high density cones can build up ahead of a flexible diaphragm as shown in Fig. 14 for a flat pressure plate. Unless the writer is greatly mistaken the breaking of the 2-in. x 2-in. specimen is not good proof of existence of such cones.

Is there not a principle stated in strength of materials that brittle substances whose shear strength is less than half of their compressive strength will translate compressive stress into 45 degree shear and fail in that manner? One has only to place a square core in the dilatometer and watch the compressive stress of expansion break the core on a 45 degree plane at each corner. Cores of concrete, refractories and bricks that have not been formed by ramming are seen to break in a similar manner to the sand specimen.

As the author states, density is not enough to insure mold wall stability. In fact the writer feels that it is so unreliable as to warrant its exclusion from a discussion of mold wall stability.

In a study of mold wall movement perhaps we should not ignore the green deformation test entirely. It measures yield under load which is exactly what the mold wall does when the casting is poured. High strength, rigidity, and hardness are the essential features of a non-yielding mold. It might be well to remind ourselves that hardness is actually a test of surface strength. Figure 18 makes these points clear.

The writer wishes to add his wholehearted approval to sands in the 16-20 psi green compression range. Actual mold strength up to 35 and 50 psi are possible with these sands when properly compounded and squeezed.

MR. HEINE *(Reply to Mr. Brewster):* The existence of sand cones or pyramids can easily be observed in molds squeezed under high pressure both with flexible diaphragm and flat squeeze.

O. J. MYERS[3] *(Written Discussion):* The author should be commended for this "critical look at an 'old' process, green sand molding". Only through work such as this can the true capabilities of green-sand molding be appraised.

It is my opinion that a study of molding *materials* is equally important in such a work. Although the author has investigated types of clays, other molding sand additives, however minor, might affect the results of the work appreciably. Lubricants and surface-active agents have been successfully added to molding media to increase the effectiveness of compaction.

MR. HEINE *(Reply to Mr. Myers):* The author agrees that mold materials should be studied particularly with reference to precision green-sand molding.

C. C. SIGERFOOS[4] *(Written Discussion):* Sand technicians and foundrymen in general should welcome the important test data presented in this paper. The author is to be commended for having accumulated data that can be directly applied to

2. Brumley-Donaldson Co., Los Angeles, Calif.
3. Archer-Daniels-Midland Co., Minneapolis.
4. Michigan State University, E. Lansing, Mich.

improve the selection of molding sand and to the development of better machine ramming techniques.

The problem of mold wall movement and its effect on the dimensional accuracy and "apparent shrinkage" of castings has been reported in the literature[1, 2] by the writer and his associates. In comparing this report with the results of the various earlier experiments carried on at Michigan State University, it is pleasing to note that there is general agreement on the sand conditions that tend to promote mold wall stability and dimensional accuracy of castings. For example, there is agreement that high mold hardness, high density and low moisture tend to promote mold wall stability. The writer has also observed that high green strength, low deformation and the presence of certain carbonaceous additives in the sand may also provide mold wall stability. These observations have been for the most part on sands used for gray iron and malleable iron castings.

The writer is in general agreement with the theme of this paper namely that careful selection and ramming of green molding sand tends to provide castings with predictable dimensions and weight. However, a word of caution regarding the predictions of mold wall movement appears to be in order. To the best of the writer's knowledge, the results of the standard green sand laboratory tests can not be relied upon to give completely accurate predictions of mold wall movement.

Using standard laboratory test data to predict wall movement for complex blends of green sand containing sea coal, wood flour or other additives has been found to be especially hazardous. The reason for this difficulty appears to be due to the fact that common green sand physical testing does not measure the effect of the heat shock and the various chemical reactions that occur during pouring. It is therefore the writers firm conviction that mold wall movement predictions for the complex blends of sands that are used in many foundries should be based upon measurements of small test castings. One type of satisfactory test casting for this purpose is 3 in. in diameter and 4 in. high and is made in a mold of controlled rammed density. The making of this test casting is described in the paper written by Sanders and Sigerfoos listed as the No. 2 reference below.

References

1. "Influence of the Mold on Shrinkage in Ferrous Castings", H. Womochel and C. C. Sigerfoos, AFS TRANSACTIONS, vol. 48, p. 591, 1940

2. "Gray Iron Shrinkage Related to Molding Sand Conditions", C. A. Sanders and C. C. Sigerfoos, AMERICAN FOUNDRYMAN, Feb. 1951.

MR. HEINE *(Reply to Mr. Sigerfoos):* The author acknowledges Prof. Sigerfoos' valuable contribution to the subject under discussion. He regrets not having made reference to the articles of Sigerfoos and Sanders. This is due to the fact that this paper was originally given as a lecture before the Wisconsin Chapter of AFS and was not provided with a bibliography as is the author's normal habit. Please accept my regrets.

D. C. WILLIAMS[5] *(Written Discussion):* The author's paper provides valuable information to assist the foundryman in learning about the compaction of sand mixtures.

This discussor would like to suggest that, along with the information in this paper, the foundryman consider the information presented 20 years ago by Loshe.[1] The result will be a more thorough knowledge of what takes place when preparing a mold.

C. A. Sanders for several years has been trying in many ways to get the foundryman to be conscious of the "bulking" quality of sands when water is added to quartz grains. The author states, "Clay and water hold sand grains apart" and shows in Fig. 1 the relation between density and ramming for sand-clay-water mixtures. No data are presented to show this relationship for a mixture composed of only sand grains and water. If this information were given then it would be possible to evaluate how much a clay addition contributed to the bulking and compacting of sand mixtures. Baker and Williams[2] found that clay additions to a sand mixture did not contribute to a significant increase in power requirements during mixing, and that the importance of the water films around the grains contributed the greatest resistance to movement.

This leads to the hypothesis that molding forces must be able to cope with and exceed the tensile forces set up between the adsorbed water layers on two adjacent grains or clay particles. It is for this reason that this discussor believes additional data are needed to complete the density-ramming relationship. Since the details of the author's sand mixtures are not given it is not possible to check on this point.

From Fig. 5 it is found that mixture B compacts to a higher density than mixtuure A. This would be expected since mixture B has less clay, therefore, there is more water available to form the adsorbed layers on the sand grains and clay particles. The forces resisting disruption built up attractive forces in the adsorbed water layers decrease as the layer thickens. Therefore, under given compacting stresses mixture B will become more dense than mixture A. This effect of the adsorbed layer is very well presented by Diran and Taylor[3] and Davies.[4]

The author is to be congratulated in presenting the information on the effect of the length of the jolt stroke. Increasing the length of the jolt stroke will greatly assist many foundrymen in producing better molds. One point should be emphasized. A jolt is not effective unless it is a complete, full, distinct jolt.

In the authors's work he found densities for his 6-in. and 12-in. tests ranging upwards from about 50 lb per cu. ft. To one familiar with the standard test specimen it seems desirable to have some point of reference for evaluating these densities. Therefore, the following is given:

Apparent Density of Rammed Mixture	Density lb/ft³	Weight of Mixture (grams) for 2-in. x 2-in. specimen
1.76	110	181
1.61	100	166
1.44	90	148
1.28	80	132
1.12	70	116
0.965	60	99.5
0.802	50	82.5

It is particularly pleasing to note that the author has endeavored to attack the problem of compacting the sand in a pattern pocket. Combining the information of the author with that of Terzaghi[5] one will be provided with information relative to the difficulties involved in trying to obtain uniform compaction in a pattern pocket. The theory of arching given by Terzaghi[5] also provides an excellent basis for understanding why the sand in many pattern recesses breaks away from the main body of the mold during the removal (draw) from the pattern.

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2. "Factors Involved in Making a Sand Mixture," D. F. Baker and D. C. Williams, TRANSACTIONS, AFS v. 61, 1953, p. 309.

3. "The Nature of Bonding in Clays and Sand-Clay Mixtures," L. M. Diran and H. F. Taylor, TRANSACTIONS, AFS vol. 60, 1952, p. 356.

4. *Foundry Sand Control,* W. Davies, United Steel Companies, Ltd., Chapter 5, p. 50.

5. *Theoretical Soil Mechanics,* K. Terzaghi, John Wiley and Sons, 1944, p. 66.

MR. HEINE *(Reply to Prof. Williams):* Thanks to Prof. Williams for his comments.

According to test results, contribution of sand and water alone to bulking is insignificant when the sand is rammed as in a standard rammer or hard rammed mold. The swelling and stickiness of clay is the cause of bulking carried over into rammed molds. The details of the sand mixture are given as a part of the graphs presented. No emphasis is placed on specific mixtures in this paper since it deals primarily with the problems of making good green-sand molds from a general viewpoint. A paper of impossible length would rave resulted if the molding material subject had been discussed.

The confirmatory references cited by Prof. Williams are gratefully received.

5. Ohio State University, Columbus, Ohio.

DOES SAND TESTING GIVE US THE FACTS?

By

R. W. Heine,* E. H. King,** and J. S. Schumacher†

It is generally agreed that a mold made in the foundry by jolting and/or squeezing does not have sand physical properties that can be directly correlated with the values found in testing by standard AFS test methods. Foundrymen, metallurgists and sand technicians have recognized these variations. The purpose of this paper is to offer a possible solution to allow these values to be more closely correlated.

AFS Foundry Sand Handbook, 1952 Edition, outlines the standard methods of sample preparation. It points out that the sand in the specimen tube is rammed three times by dropping a 14-lb weight 2 in. It has often been observed that the mold hardness developed on the sand specimen after ramming is generally different from the hardness developed by the same sand in the mold. This is true because most foundries do not have molding machines capable of equaling the ramming intensity of the standard AFS rammer. Therefore, it is not only a dangerous assumption to consider that sand test strengths directly relate to mold strengths, but it can be demonstrated that this is not true.

Mold Hardness vs Specimen Hardness

It is assumed that at equal mold hardness (test specimens versus sand in the mold), properties of green strength, permeability, density, etc. of a given sand are similar—whether the hardness is achieved in a mold made on a jolt machine, or in a standard 2-in. AFS sand specimen made on the standard rammer. It can be shown that this assumption is reasonably true for a particular foundry sand. Thus, the mold hardness test should be a good basis for the desired correlation. Unfortunately, when a foundry sand is tested in the laboratory, it is commonly found to have a much higher mold hardness than can be achieved by jolting (and squeezing) on a molding machine.

Effect of ramming and jolting on mold hardness of a typical green sand is shown in Fig. 1. Jolting for Fig. 1 was performed on a machine having a 3-in. jolt travel (this corresponds to the jolt travel on many standard molding machines.) Sand heights of 6 in. and 12 in. were used over a flat parting surface. Mold hardness measurements were made on the flat surfaces. Ramming was performed on the standard AFS sand specimen using the AFS standard ramming device which has a 14-lb weight dropped 2 in.

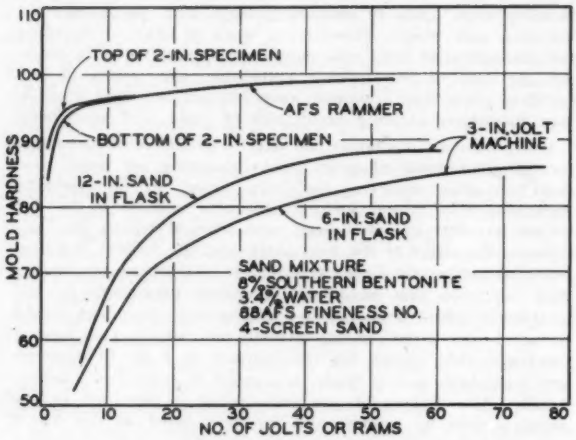


Fig. 1 — Effect of number of jolts on hardness at parting and effect of number of rams with 14-lb weight on hardness of top and bottom of standard AFS specimen.

Figure 1 shows that mold hardness increases with number of jolts, but that a hardness is reached beyond which additional jolting produces little increase in hardness. This is a common experience in the foundry. Figure 1 also shows that the mold hardness produced at the parting surface is greater when jolting with 12 in. of sand than with 6.0 in. of sand. The greater weight of a 12-in. sand height produces the higher hardness during jolting. With reference to Fig. 1 repeated ramming also causes mold hardness to increase to a maximum value.

However, Fig. 1 shows that the mold hardness achieved with the AFS rammer is much higher than that obtained from the jolt machine, regardless of the number of jolts. Each ram is a much more severe molding action than a jolt. The standard test of 3 rams is accompanied by a mold hardness of 93-94 for the sand. This hardness is not reached by jolting

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**Vice President and †Chief Engineer, The Hill & Griffith Co., Cincinnati, Ohio.

even after more than 50 jolts. Thus, the standard 14-lb, 2.0-in. drop, 3-ram test does not produce properties of mold hardness, green strength, permeability, or sand density in the 2-in. x 2-in. specimen comparable to those obtained after the usual 10 to 20 jolts on a molding machine.

Equivalent Ramming and Jolting

The question may then be asked: "What relationship should exist between ramming with the AFS testing device and jolting on a molding machine?" If the same work is done on the sand by both methods, equivalent sand density, mold hardness and other properties should be reached. Simple calculations based on physics can be used to relate ramming power and work done by the rammer weight for a 2-in. drop and the jolting power and work done on a molding machine. For the AFS rammer, maximum work which can be done on the sand per ram is that equal to the kinetic energy of the rammer weight or

$$14.0 \text{ lb} \times 2.0/12 \text{ ft} = 2.335 \text{ ft-lb}$$

However, this work is done to a 2-in. high x 2-in. diam specimen over an area of

$$\frac{\pi R^2}{144} = \frac{3.1416 \times 2x2/4}{144} = 0.0218 \text{ sq ft}$$

Thus, work applied by the standard rammer weight is

$$2.335 \div 0.0218 = 107.5 \text{ ft-lb/sq ft}$$

Similarly, maximum work which can be done by jolting is that equal to the kinetic energy of the *sand* when the jolt table strikes. For calculation, assume a sand column 12.0 in. high x 12 in. square, i.e. 1 cu ft, and a sand density of 80 lb per cu ft. For a 3.0-in. jolt stroke, kinetic energy of the cubic foot of sand would be 80 lb x $\frac{3}{12}$ ft or 20 ft-lb. This work is applied to 1 sq ft of area, so 20 ft-lb per sq ft work can be done per jolt to sand at the parting line. Note that the jolt can do less than one-fifth as much work as the AFS rammer under the conditions described. To make the ramming conditions more like jolting, the weight of the rammer must be reduced. Work which could be done by different weights dropped 2.0 in. is given in Table 1.

TABLE 1 — WORK ENERGY AVAILABLE FOR COMPACTING SAND WITH THE AFS RAMMER

Rammer Weight, lb	Energy on 2-in. x 2-in. Diam Specimen, ft-lb per sq ft
14.0	107.5
6.0	46.05
4.0	30.7
3.0	23.0
2.0	15.35
1.0	7.67

Table 1 shows that a 2- to 3-lb weight dropped 2.0 in. would provide kinetic energy equivalent to jolting 12.0 in. of sand weighing 80 lb per cu ft with a 3.0-in. stroke. However, in molding, the jolting energy is subject to two variables, the jolt stroke and sand weight. The jolt stroke depends on the molding machine and is fixed thereby. Weight of the sand is determined by its height in the flask and its density in pounds per cubic foot before jolting begins. Densities of 40 to 80 lb per cu ft before jolting have been observed in different sands studied by the authors. To give an equivalent molding effect, the ramming weight must provide the same energy as is obtained from the combination of sand weight (sand height and density) and jolt stroke. Ramming energy is expressed by $\frac{WxR}{A}$ ft-lb per sq ft where W is the rammer weight, R is dropping distance of $\frac{2}{12}$ ft and A is the area of the specimen of 0.0218 sq ft. Jolting energy is expressed by DxHxJ, where D is the density of the sand in lb/cu ft, H is the height of the sand in feet, and J is the jolt stroke in feet. For equal molding work

$$\frac{WxR}{A} = DxHxJ \text{ or } \frac{Wx2}{0.0218x12} = \frac{DxHxJ}{12x12}$$

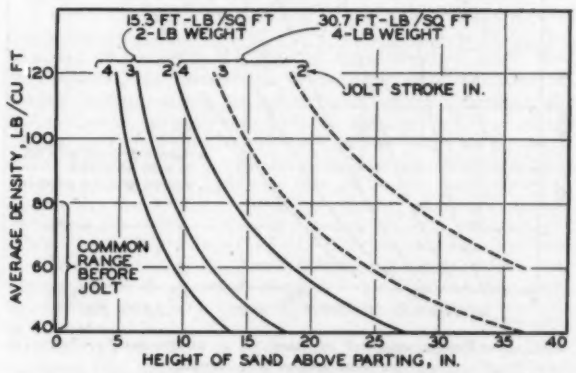


Fig. 2 — Density and height of sand in flask imparting same energy per square foot of parting surface by jolting with 2-in., 3-in., and 4-in. stroke and by ramming with 2-lb and 4-lb weight in AFS ramming device.

Relationship of these factors is graphically shown in Fig. 2. For a given jolt stroke, the curves relate density of molding sand and its height in a flask which will provide the same molding work to the sand in a mold as 2- and 4-lb rammer weights dropped 2.0 in. will give to the standard sand specimen. For example, sand weighing 60 lb per cu ft with a sand height of 12.0 in. will have the same work done on it in a mold by a 3.0-in. jolt as would be done on a 2-in. x 2-in. diam sand specimen by the 2-lb rammer weight, dropping 2.0 in. in the standard rammer. If the sand weighed only 40 lb per cu ft, a 4.0-in. jolt and 13.75 in. of sand would be needed for work equal to the 2-lb, 2.0-in. ram. About 7 in. of sand at 80 lb per cu ft and 4.0-in. jolt would give work energy equal to the 2-lb, 2.0-in. ram.

Power is a measure of doing work in foot-pounds per second. A jolt or ram is applied instantaneously at the moment of impact. However, time is required to compact the sand. Thus a comparison of the

power developed by jolting and ramming is in order. In either case, the power is defined by the momentum (mass and velocity) of the ram weight or sand weight per unit area at the instant of impact. For a 2-lb weight dropped 2.0 in., power is calculated as follows:

$MxV \div A$ = Power for sand compaction per unit area

M = 2 lb

V = Velocity calculated from the acceleration due to gravity = $\sqrt{2\,gd}$

$V = \sqrt{2 \times 32.2\ \text{ft/sec}^2 \times 2/12\ \text{ft}}$

V = 3.28 ft/sec

A = 0.0218 sq ft surface area of 2.0-in. diam specimen

$$2\ \text{lb} \times 3.28\ \text{ft/sec} \div 0.0218\ \text{sq ft} = 301\ \text{ft-lb per sec per sq ft}$$

Similar calculations can be made for each rammer weight, or jolting combination of sand weight and jolt stroke. Power factors are summarized in Fig. 3.

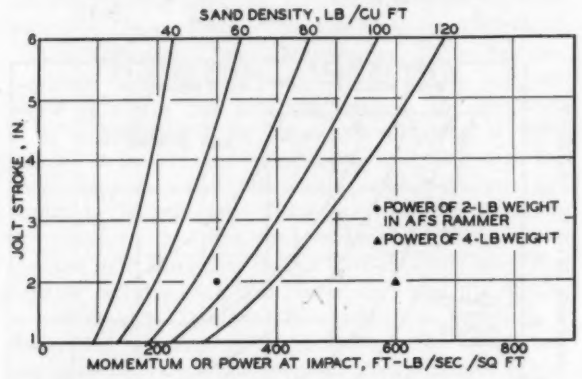


Fig. 3 — Power applied to base of a 12-in. sand column as a function of jolt stroke.

From Fig. 3 it is evident the 3.0-in. jolt stroke acting on 12.0 in. of sand weighing 80 lb per cu ft develops slightly more power than the 2-lb, 2-in. ram. Jolting a sand weighing only 60 lb per cu ft under these conditions would develop less power than the 2-lb, 2.0-in. rammer. Power developed by the 14-lb, 2.0-in. ram is completely outside the limits of the graph.

Considering the principles described above, it is no wonder that good correlation of properties of molding sand between those obtained with the standard AFS 14-lb, 2-in. ram and by jolt molding in the foundry has not been possible. The 14-lb weight has a ramming work and power of 5 to 7 times that encountered in practice. For exact correlation of laboratory and foundry properties, the rammer weight-distance combination should be selected to do the same work as the jolt stroke-sand weight combination. This may be done with the aid of Figs. 2 and 3. For convenience, however, it is desirable to standardize on one or at the most a few rammer weights. The 2- and 4-lb weights seem to cover most conditions and are therefore recommended for the ramming device.

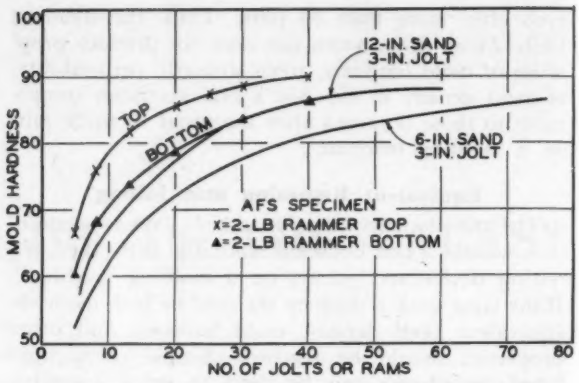


Fig. 4 — Effect of number of jolts on hardness at parting and effect of number of rams with 2-lb weight on hardness of top and bottom of specimen.

Mold Hardness and the 2-lb Rammer

Effect of ramming with the 2-lb weight on mold hardness and comparison with mold hardness obtained by a 3.0-in. jolting is shown in Fig. 4. The same sand was used as for Fig. 1. Correlation of mold hardness with number of jolts and rams is much better for a 2-lb, 2.0-in. ram than for the 14-lb, 2.0-in. ram. This is demonstrated by a comparison of Figs. 1 and 4. However, even the 2-lb weight produces a somewhat higher hardness than is obtained with a 3.0-in. jolting. This is necessarily so in this case, because of the low starting density of this sand, 49 lb per cu ft. Reference to Figs. 2 and 3 show that 12.0 in. of sand at a density of 49 lb per cu ft will not do as much work or provide as much power with a 3.0-in. jolt as the 2-lb, 2.0-in. ram. Therefore, a slightly lower hardness would be expected from 3.0-in. jolting. Four-inch jolting, however, should produce higher hardness than 2-lb, 2-in. ramming, which it did.

Sand Density, Ramming, Jolting

Average sand density is another means of studying compaction of sand which occurs during molding. Figure 5 shows how density of a molding sand is affected by ramming and jolting. A comparison is

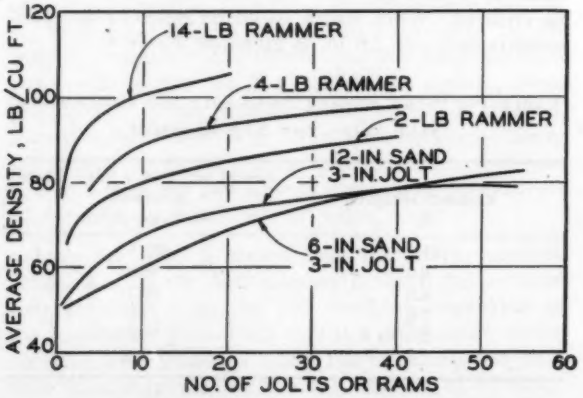


Fig. 5 — Relation of average sand density to number of rams with 14-lb, 4-lb, and 2-lb weights and to number of jolts (3-in.) with 12-in. and 6-in. sand columns. (Same sand as in Fig. 1 and Fig. 4.)

made of the effects of 2-, 4- and 14-lb ram weights, and jolting with a 3.0-in. stroke a sand column 12.0 or 6.0 in. high of 49.0 lb cu ft initial density. The sand employed was the one also used for Figs. 1 and 4. Again, the 2-lb, 2-in. ram provides a much closer correlation of density with jolting. The 14-lb, 2-in. ram is completely unrealistic in this comparison. Even the 2-lb weight produces higher average density than 3.0-in. jolt with 12.0 in. in height of this sand. Again, this is caused by the low starting density of the sand and a consequently lower applied power and resultant work.

Other sands with higher starting density show even closer correlation of jolting with 2-lb, 2-in. ramming. Moisture content, clay type and amount, and additions, all influence the starting density of the sand before jolting. Experience with a variety of sands shows that the 2-lb, 2.0-in. ram provides a reasonable comparison with the 3.0-in. jolt and a 4-lb, 2.0-in. ram compares with a 4- to 4.5-in. jolt.

With respect to density there is a fundamental difference between the AFS ramming tests and jolting. The ramming procedure produces a fairly uniform density because the sand is confined in a tube. Still, density (and mold hardness) are maximum at the surface of impact. The higher the average density of the 2x2.0-in. specimen, the more uniform its density is; while the lower its average density, the greater the difference in density from top to bottom. This is the reason for the greater difference of top and bottom mold hardness with the 2-lb, 2-in. ram (Fig. 4) compared with the 14-lb, 2.0-in. ram (Fig. 1). Jolting produces a density gradient with high density at the parting (impact) surface and low density near the surface away from the parting. Figure 5 shows only average density values.

Since the 2-lb, 2.0-in. ram produces more of a density gradient (and top and bottom hardness difference) it again more closely simulates a jolting effect. Properties such as permeability, green strength, dry strength, hot strengths, etc. are affected both by average density and density gradient. It should be obvious that a laboratory sand specimen rammed to a high and uniform density by a 14-lb, 2.0-in. ram has much higher green strength and far lower permeability than the same sand jolted on a mold to a lower average density with the density decreasing away from the parting. Thus, for correlation of sand properties obtained in the laboratory with those existing in a mold, 2-lb, 2.0-in. ram and 4-lb, 2.0-in. ram procedures are recommended. The number of rams should correspond to the number of equivalent jolts used in the foundry.

For comparison purposes, a standard of 15 rams and jolts is suggested, although individual foundry practices may indicate a higher or lower figure.

Conclusion

The foregoing does not imply a rejection of the 14.0-lb, 2-in. ram procedure. The latter provides a useful means of comparing the properties of sand at high density, where the data are most reproducible, are least affected by non-uniform density, and the ultimate bonding properties and minimum permeability of the sand may be determined. However, the 14-lb, 2.0-in. ram does not produce molding conditions comparable to those of the jolt-squeeze machine, and should therefore not be used for correlation of sand properties in the laboratory and foundry. The 2- and 4-lb weights with 2.0-in. drop are recommended, to better correlate foundry results with laboratory control data.

DISCUSSION

Chairman: E. C. TROY, Riverton, N. J.

Co-Chairman: W. R. MOGGRIDGE, Ford Motor Co. of Canada, Windsor, Ont., Canada

Secretary: H. F. BISHOP, Naval Research Laboratory, Washington, D. C.

J. B. CAINE[1] *(Written Discussion):* Agreed that a 2-lb rammer weight as advanced by the authors should be used, the question is what to do until such a weight is agreed upon and the rammers in use are changed. This will take years. The authors' premise that test specimens should be compacted to densities and hardnesses representing those of the mold is too important to wait for this changeover.

What are the authors opinions as to the use of the present 14-lb weight, but at a lower number of rams than the present three? Test specimen green hardnesses from 40 to 80, about the range to which vertical surfaces are rammed in production can be obtained with the 14-lb weight with a press (0 ram), 1 and 2 rams.

If green hardness is to be a test of major importance a very important point is how and where the hardness is to be determined on molds in the foundry. This discusser would like to emphasize the necessity of determining green hardness on vertical surfaces, almost to the exclusion of readings taken on flat horizontal surfaces, especially the parting. It is not at all unusual to find vertical surfaces rammed to under 60 hardness when the parting shows over 80 green hardness. Admitted that hardness readings are harder to take on vertical surfaces, vertical surfaces are by far the more important. Mold wall movement and penetration are most common on such surfaces.

MR. HEINE *(Reply to Mr. Caine):* One- and two-ram tests are useable for extending the density range of specimens which can be rammed on the standard rammer. Green hardness should be measured at every point deemed important that can be reached in a mold.

H. W. DIETERT[2] *(Written Discussion):* The authors make the statement that correlation with casting quality has not been possible with the present AFS high density ramming. This statement must be in error as is shown by the following correlations.

Referring to Fig. A, one may note that excellent correlation is obtained between dry compressive strength and the number of castings lost per 10,000 castings on hydraulic test. Reference page 252, AFS TRANSACTIONS, vol. XXXIV (1926).

AFS Committee 8-J in its report of 1951 showed that dry compressive strength influenced the degree of scabbing. The graph, Fig. B, shows this correlation. Dry strength above 100 psi was inducive to production of castings with large scabs. Below 100 psi dry strength, scabs were either absent, or were small.

In AFS TRANSACTIONS, vol. 63, p. 138 (1955), is Fig. 32 showing that sands with green compressive strength under 11 psi produced castings with and without scabs. When the green strength was above 11 psi no castings were produced with scabs. This graph is reproduced herein as Fig. C.

One more proof of correlation is illustrated on page 206, AFS TRANSACTIONS, vol. 63 (1955), as Fig. 5. This graph is illustrated herein as Fig. D. Wyman shows a correlation between the dry compressive strength and hot tearing of steel castings. An excellent correlation is secured. Thus it is incorrect to state that no correlation exists between AFS rammed specimens and casting quality, since all tests referred to were rammed with the AFS sand rammer.

Undoubtedly, the authors intended to state that possibly better correlations could be obtained when the sand specimen in the laboratory was rammed to the same hardness as the molds in which the castings were made.

1. Consultant, Cincinnati, Ohio.
2. Chairman of the Board, Harry W. Dietert Co., Detroit, Mich.

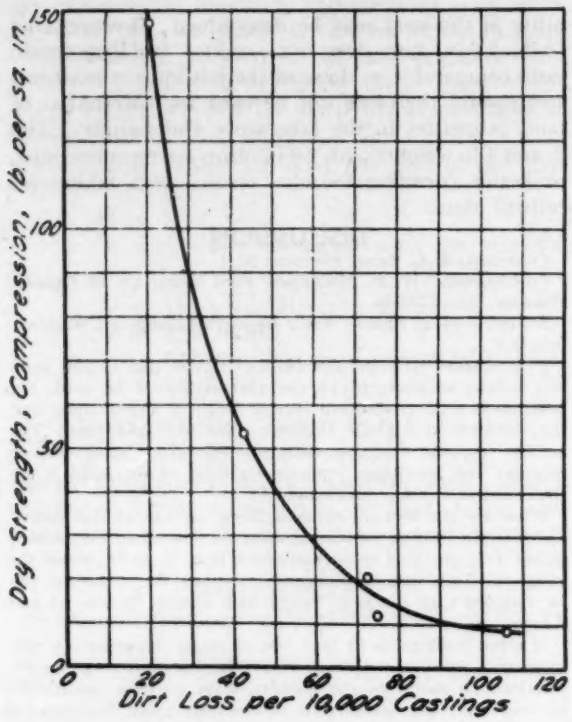


Fig. A — Close correlation between dry strength and dirt loss per 10,000 castings on hydraulic test.

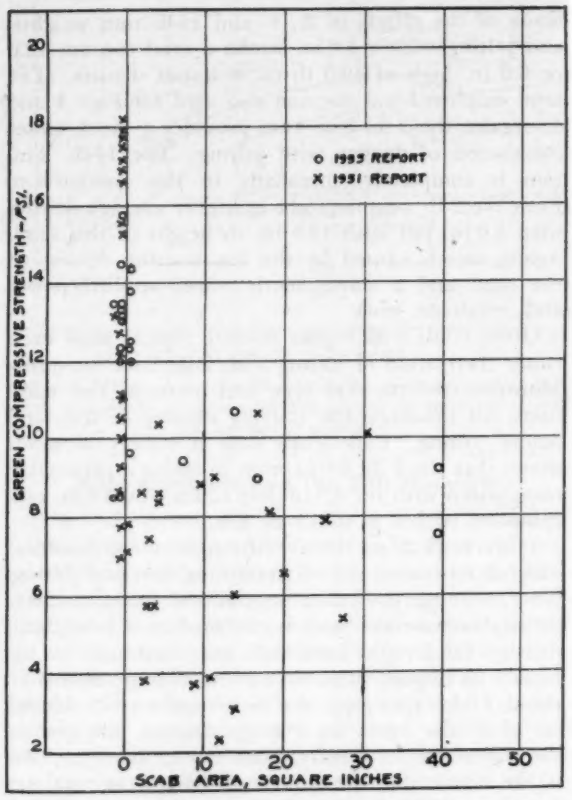


Fig. C — Green compressive strength related to scabbing tendency.

The authors recommend that the AFS rammer be changed to give the same work energy as the molding machines employed in the foundry. High frictional resistance of the relatively small specimen tube as compared to mold area prevent this.

They also recommend that the density of the sand specimen and mold be identical. This reasoning is good. However, obtaining density values of sand in the mold is time consuming.

Density is not a good overall measure of load-carrying ability of a sand in a mold. Density varies from sand to sand due to composition, grain size and grain distribution.

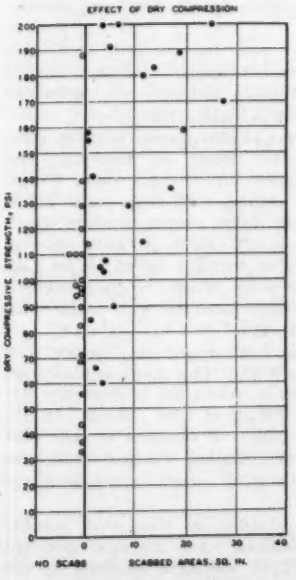


Fig. B — Relationship between dry compressive strength and degree of scabbed area.

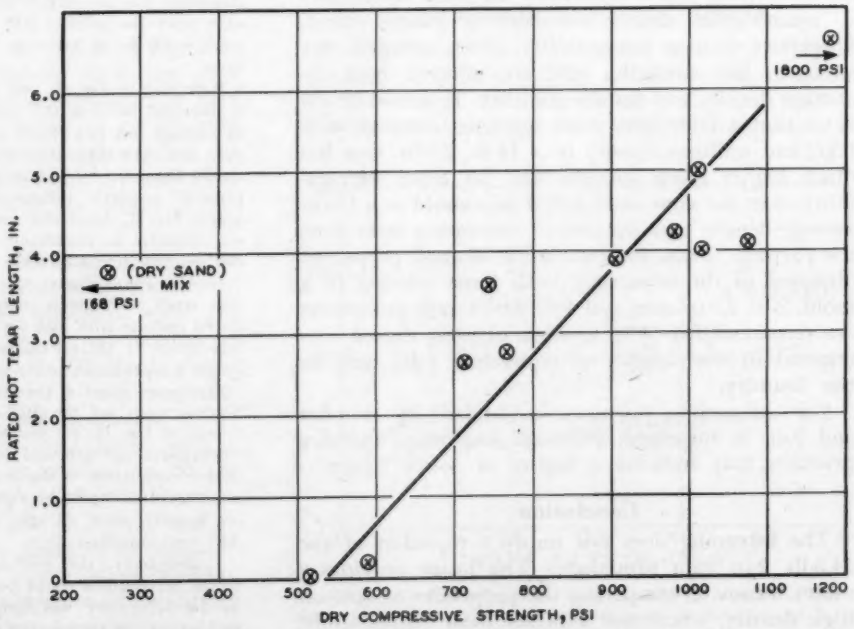


Fig. D — Degree of hot tear increases with an increase of dry compression.

A convenient method is to ram the specimen to the same hardness as the mold is in the foundry. With this method the higher frictional resistance of the specimen tube can be ignored.

Evaluating all sand test values on the basis of hardness is good practice together with standardization of the hardness of the molds in the foundry.

The authors further recommend a new AFS sand rammer with a 2-lb weight. The foundries that they visited had abnormally soft rammed molds. It is their suggestion that a 2-lb weight be employed. This will require up to 50 drops for the usual hard-rammed molds in a steel foundry.

A better practice for those wishing to ram the sand specimen to the same hardness as the molds in the foundry is to use the present AFS sand rammer and drop the weight from a distance less than the standard 2-in. drop. An inexpensive hardness control sand rammer accessory is illustrated in Fig. E. This accessory can be set to vary the height of rammer weight drop so that it will produce any hardness desired. This is accomplished by loosening two screws and changing the position of the cam support plate. By lowering the cam in this manner, any portion of the total cam rise may be used. The support plate can be raised for the AFS standard cam.

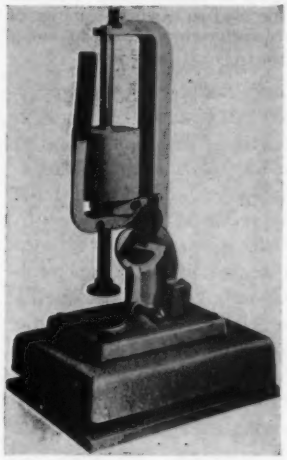


Fig. E — Accessory for AFS sand rammer to conveniently obtain specimens at any selected hardness.

MR. HEINE *(Reply to Mr. Dietert):* The authors do not, as Mr. Dietert says, make the statement that "correlation with casting quality has not been possible with the present AFS high density ramming." Instead, the authors say that ". . . a mold made by jolting and/or squeezing does not have sand physical properties that can be directly correlated with the values found in testing by the standard AFS testing methods." Therefore, there is no disagreement with the items cited by Mr. Dietert. The statement was made that the 2-lb, 2.0-in. ram makes it possible to achieve comparable densities in mold and sand specimen.

J. F. WALLACE[3] *(Written Discussion):* The authors have clearly demonstrated the reasons for the difference in the mold hardness and density of molding sand compacted in a jolt machine compared to the AFS sand specimen made on the standard rammer. While the total work energy will probably be similar when the greater number of jolts utilized in jolt machine practice is compared to the three rams for the AFS standard test, it is apparent that the method of force application and momentum considerations are sufficiently dissimilar to cause different properties.

The discusser has also observed a similar difference between the hardness of AFS rammed specimen and molds produced in jolt and squeeze machines, although the hardness of slinger and hand rammed molds, particularly in steel molding, are considerably closer to the standard test.

This paper has very ably demonstrated the advantages of employing 2-lb or 4-lb weight in preference to the 14-lb standard for ramming the AFS specimen when testing the properties of molding sands for jolt machines. The problem of reproducibility of the test has also been mentioned. This is an important requirement for any significant test. Do the authors have any results that indicate the reproducibility of green and dry compression and permeability tests of the standard size rammed with 2-lb and 4-lb weights for a total of 15 to 20 times?

MR. HEINE *(Reply to Mr. Wallace):* Reproducibility of the test values from 2 and 4 lb weights is likely to be less than for the 14 lb weight. This is due to the fact that low density specimens always allow more density variation within the specimen causing the test results to fluctuate somewhat. This, of course, is also true of low density molds in the foundry.

T. W. SEATON[4] *(Written Discussion):* The work done by Messrs. Schumacher, King and Heine is most commendable. The time for us to take a critical look at present sand testing methods has certainly arrived. Much emphasis certainly has been placed on the reproducibility of laboratory tests and little or no effort placed on a correlation of laboratory tests with actual foundry conditions.

This paper certainly points out the error in believing that a laboratory test on a 2x2-in. specimen closely duplicates what we are doing in the foundry with the same sand. Obviously, a sand jolted and squeezed to a surface mold hardness of 75 will have different physical characteristics than the laboratory specimen of the same sand rammed to a hardness of 95.

I understand that this work was carried out with gray iron foundry sands. I feel that foundries in the steel, malleable and non-ferrous fields should run tests similar to those indicated in this paper for the purpose of emphasizing that this condition occurs with all types of sands. A close evaluation of these sands in this manner will certainly give the foundry a more accurate picture of its own sands.

It is also my feeling that the first sentence in the conclusion of this paper is incorrect to the extent that the data presented certainly does indicate and should imply that the use of the 14-lb rammer is useless except in foundries that actually work at the mold hardness figures obtained with the 14-lb rammer. There is a very little point in testing sands in the laboratory at higher mold hardnesses and densities than we actually achieve with the equipment and sands in actual operation.

It would certainly appear, in the light of what this paper has shown, that one of the AFS Sand Division Committees should be assigned the project of furthering this investigation and given the authority to change the present method of ramming specimens. Certainly our present method of ramming should be modified so that we can ram to any desired mold hardness. The use of lighter weight rammers is certainly a step in the right direction.

The closer our laboratory tests correlate with actual foundry operating conditions, the greater will be the value of the sand tests and certainly the greater will be the understanding of sand and its importance in the cast metals industry.

MR. HEINE *(Reply to Mr. Seaton):* The 14-lb weight has value for testing because it rams to densities that make possible the evaluation of clay bonds, sieve analyses, additives, in the range where test results are most reproducible. Also, molds should eventually be rammed to these densities.

V. M. ROWELL[5] *(Written Discussion):* This observer has noted that in many foundries producing castings of good quality the molds are rammed to hardness values approximating the hardness of the AFS standard test specimen. Thus it can be assumed that the properties of the sand in the molds will resemble those of the standard test specimen. Of course some foundrymen ram their molds harder and some ram softer than the standard specimen.

Molding and casting quality problems can be predicted and cured by virtue of the correlations that have been established using the standard rammed specimen. Even better correlation can be expected by ramming a test specimen to the same hardness as individual molds.

The sand used to compile the data in the paper is not described but is probably a sand of inferior rammability. A sand with good rammability will be easily compacted to optimum density with the expenditure of a minimum of ramming energy. A sand possessing poor rammability will require more energy. The former type of sand will not change properties materially under the influence of excess ramming energy but the second

3. Professor, Case Institute of Technology, Dept. of Met. Engr., Cleveland, Ohio.

4. Sales Manager, American Silica Sand Co., Inc., Ottawa, Ill.
5. Executive Vice President, Harry W. Dietert Co., Detroit, Mich.

sand requires more ramming energy in order to demonstrate its potential.

Upon investigating the true amount of kinetic energy applied to molding sand by the 14-lb ramming weight, the 2-lb ramming weight, and jolt ramming a new picture shapes up. The authors disregarded the inertia of the rammer head, shaft and crosshead. Due to mechanical considerations we cannot eliminate these parts. These parts weigh 3½ lbs and are stationary until set in motion by the falling weight.

It can be shown that some kinetic energy is expended in overcoming this inertia. This amounts to a kinetic energy loss of 21.5 per cent for the 14-lb weight and 66 per cent for the 2-lb weight. The net amount of kinetic energy received by the sand from one blow of the 14-lb weight is 107 ft-lb minus 21.5 per cent or 84 ft-lb per sq ft.

The net amount of kinetic energy received by the sand from one blow of the 2-lb weight is 15.3 ft-lb minus 66 per cent or 5.2 ft-lb per sq ft. A 12-in. by 12-in. by 12-in. cube of sand weighing 80 lb would have a kinetic energy of 20 ft-lb upon falling 3 in.

To equal the ramming energy supplied to the sand by dropping the 14-lb weight 2 in. 3 times (252 ft-lb) we would have to jolt our hypothetical 1-ft cube of sand 3 in., 12.6 times. This presumes perfect efficiency in the jolt machine. To equal this ramming work we will have to drop the 2-lb weight 2 in., 48.5 times. Several sands were rammed in the standard manner and by the 2-lb weight to the same average density. 40 to 50 blows were required. This variation in the experimental data with different sands is probably due to variations in the internal friction and elasticity of different sands.

Since we can equal the ramming energy of 12.6 jolts for this above-mentioned sand with 3 blows of the 14-lb standard weight dropping 2 in. or 48 blows of the 2-lb weight falling 2 in. it is certainly more practical to simulate mold conditions with the AFS standard rammer.

MR. HEINE *(Reply to Mr. Rowell):* In cases where the sand in a mold is rammed to the same density as the standard specimen made with the 14-lb weight, the sand properties in each should be comparable. The lighter weight, or shorter drop with the standard weight if you will, can be used to extend the range of testing to the lower densities frequently encountered in molds in the foundry.

JOLT TEST FOR SAND

By

R. W. Heine,* E. H. King,** and J. S. Schumacher†

Introduction

This article points out discrepancies between laboratory testing and the results developed in the mold by standard molding procedures.

In another article (see pages 408-14) entitled: "Does Sand Testing Tell Us the Facts?" the same authors illustrate that the laboratory preparation of the AFS standard sand specimen imparts properties to the sand that are not found in actual molding practice. The article further describes a method of sand preparation that develops the same strengths in the specimen as found in the mold. This was found to require (instead of the 14-lb standard ramming weight) a weight of 2 or 4 lb, depending upon the molding methods used. The present article further develops this idea and compares the test data obtained with 2x2.0-in. diameter specimens found with the 2-lb weight to sand properties achieved on conventional jolt molding machines.

A comparison is made by taking a metal tube approximately 2 in. in diameter by 12 in. high, and a second tube 2 in. in diameter by 6 in. high, filling them with sand and jolting on a molding machine table to produce sand samples for testing. The two different heights are used to simulate flask height above the pattern. Details of the test are given in the article.

Jolt Test for Sands

The jolt test is performed on the common jolt molding machine using a few accessories for forming the sand sample. A section view of the sand sample tube and cup is shown in Fig. 1. The sample container consists of a hollow steel or brass tube 2.0-in. I.D., 2.375-in. O.D., by 12-in. long. The tube is sealed in a base cup of the same size as that used for the specimen tube of the standard AFS sand specimen. The base cup rests on a steel block 2x4¼x4¾ in. The tube, cup and block are assembled on the jolt table of the machine as shown in Fig. 1.

The tube is filled to overflowing with molding sand that has been put through a 6-mesh riddle. It is then rapped with a small mallet three or four times at each quadrant about 4 in. from the bottom of the tube. The tube is then refilled and struck off. Thus, the specimen consists of a sand column 12 in. long x 2.0 in. in diameter. Reproducibility of the weight of sand in the column to within 2 per cent for the author's technique. The tube is then jolted five times on the molding machine. Upon jolting, the sand settles in the tube. The distance from the top of the tube down to the sand level is measured, and subtracted from 12.0 in. to obtain the height of the sand column after five jolts. The tube is then removed from the cup and several mold hardness readings are taken on the bottom sand surface (equivalent to parting plane of a pattern).

Occasionally, the first five jolts will not prevent the specimen from sliding in the tube from the pres-

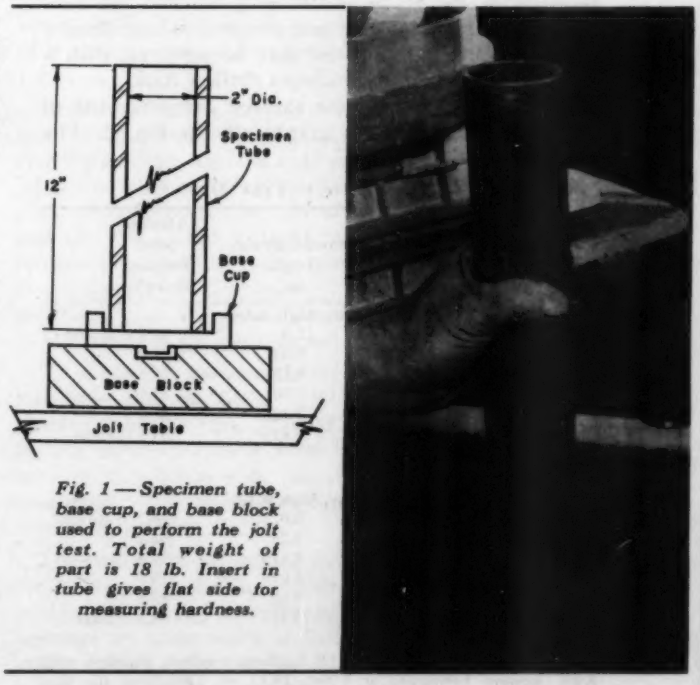


Fig. 1 — Specimen tube, base cup, and base block used to perform the jolt test. Total weight of part is 18 lb. Insert in tube gives flat side for measuring hardness.

*Professor, University of Wisconsin, Madison, Wis.; **Vice-President and †Chief Engineer, respectively, The Hill & Griffith Co., Cincinnati, Ohio.

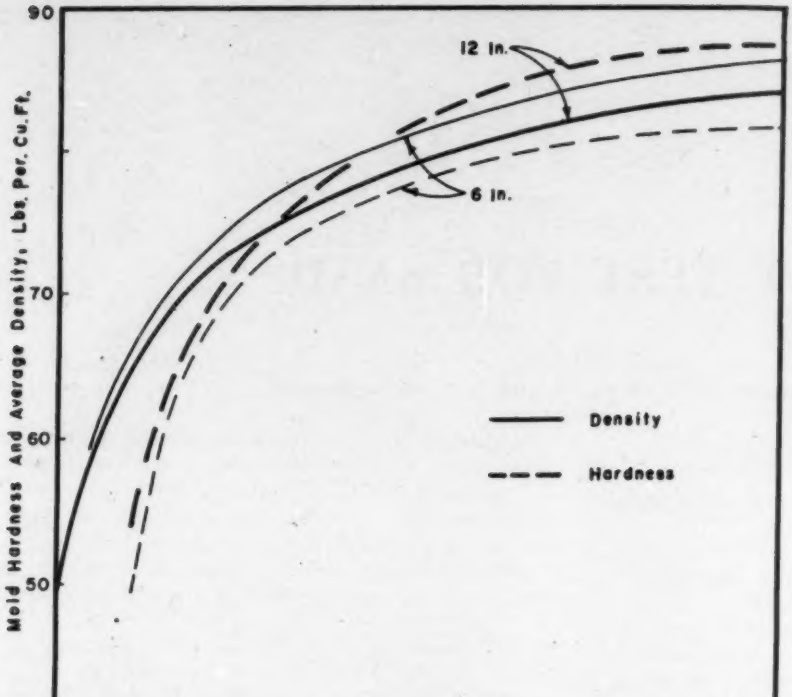


Fig. 2 — Effect of number of jolts on mold hardness at the jolt surface (parting) and average density of the sand in the tube 12-in. and 6-in. sand columns and a molding machine with 3-in. jolt stroke. Sand mixture is given in Table 1.

sure of the hardness measurement. Force applied to the top side of the specimen will prevent specimen movement. The tube is then replaced in the cup and jolted an additional five or more times. The above measurements are repeated and data are taken at about 5, 10, 15, 20, 30 and 40 jolts. Finally, the sand is removed from the tube and weighed so that density can be calculated. The test may be repeated with a 6.0-in. sand column to simulate a shallow flask.

A sample set of data for the test are presented in Table 1 and are shown graphically in Fig. 2. The figure shows that increasing the number of jolts causes a progressive increase in mold hardness and average density. However, after about 30 jolts, mold hardness and average density values reach a level where further jolting produces only minor improvements. Figure 2 also reveals that hardness is lower and average density is higher for a 6.0-in. sand column than for a 12.0-in. sand column for this particular sand.

TABLE 1 — JOLT TEST DATA*

Number of Jolts	Mold Hardness	Sand Column Height, in.	Average Sand Density, lb/cu ft
	For 12-in. high tube		
0	—	12.0	48.3 — initial
5	54	8.94	64.9
10	69	8.125	71.5
20	79	7.50	77.5
35	86	7.125	81.6
50	88	6.875	84.5 — final
Weight of sand	501 grams	—	—
	For 6-in. high tube		
0	—	6.0	48.3 — initial
6	49	4.25	68.2
11	70	3.875	74.8
21	77	3.625	79.9
36	81	3.50	82.8
52	82	3.375	85.8 — final
Weight of sand	250 grams		

*For a 4-screen sand, 89 AFS fineness number, bonded with 8.0% western bentonite at 3.20% H_2O, no additions, the particular specimen tube was 2.0468 in. I.D. x 12.0 in. long.

The lower hardness at the base of the 6.0-in. sand column means that the density is also lower at the column base. Likewise, the higher hardness at the base of the 12.0-in. column means that the density is higher at the column base. However, the 6.0-in. column has a higher average density than the 12.0-in. column because there is less density gradient from top to bottom in the former. The greater density gradient in the 12.0-in. column is due to the greater amount of internal friction of sand grains in the longer column. With another sand of better response to jolt molding (higher flowability) the 12.0-in. column will have higher average density than the 6.0-in. column.

The lower mold hardness at the base of the 6.0-in. column when compared with the 12.0-in. column results from the lesser weight of sand in the former. Lower weight produces lower power input to the sand from jolting.

Correlation of green sand properties at the jolt or parting surface may be partially accomplished through the mold hardness value and 2.0-in. x 2.0-in. diameter sand specimen. These specimens are tested in the usual way for permeability, mold hardness and

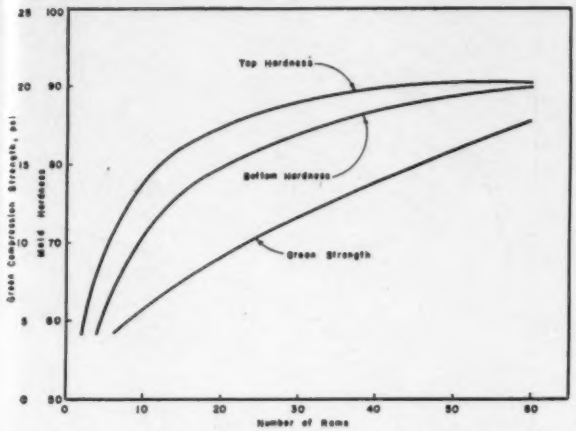


Fig. 3 — Effect of number of rams with a 2-lb weight dropped 2 in. on green compressive strength and mold hardness of the 2-in.x2-in. diameter specimen. Sand same as that in Table 1 and Fig. 2.

green compressive strengths, but they are prepared on a sand rammer having a 2-lb rammer weight, rather than the standard 14-lb weight.* The 2-lb weight is necessary to obtain sand densities and mold hardness of the same magnitude as those imparted by a jolt machine. The relation of number of rams to mold hardness and green strength of the sand described in Table 1 and Fig. 2 is presented in Fig. 3.

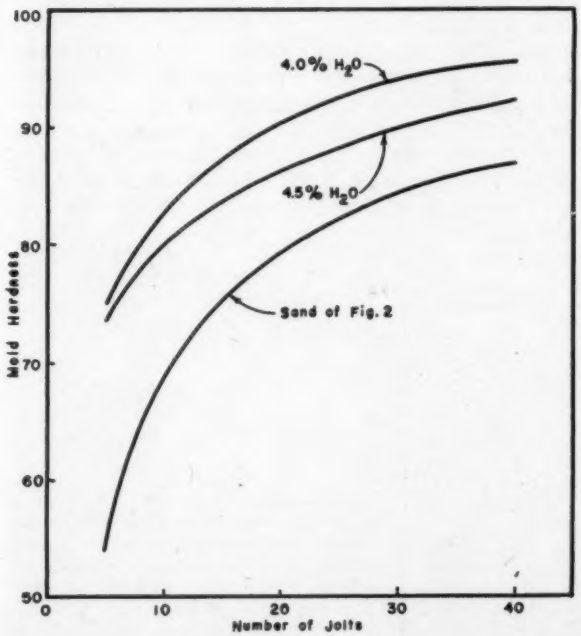


Fig. 4 — Effect of number of jolts on mold hardness at the jolt or parting surface. The foundry sand employed is a 4-screen sand of 72 AFS GFN, 9.5% AFS clay, western bentonite bonded, 7.5% combustibles from additives, 4.0 and 4.5% water. The curve from Fig. 2 is reproduced for comparison of the two sands. The same molding machine with 3-in. stroke was used in each case.

*See "Does Sand Testing Give Us the Facts" by the same authors in this volume. See pages 408-414.



Fig. 5 — Taking mold hardness readings.

Permeability is not considered since it was over 90 in all cases. The top of the 2.0 x 2.0-in. specimen receives the ramming impact, so that it is similar to a jolted surface. Using Figs. 2 and 3, a mold hardness of 80 at the top of the specimen and the bottom of a 12.0-in. sand column is accompanied by about 6.5 psi green compressive strength. The figure is not exactly accurate, since in each case there is a density gradient in the sand. However, it is a far more respresentative value than would be obtained with the standard sand testing combination of 14-lb weight, 3 rams and 2.0-in. drop; in the case of this 12.0 psi and 90-92 mold hardness.

For the jolt test to be useful, it should be capable of revealing any differences which may exist in response of a number of sands to jolt molding. For comparison with Fig. 2, the effect of jolting on mold hardness of a second sand at two moisture levels is shown in Fig. 4.

Mold hardness is higher at a lower moisture content. Mold hardness of the second sand is substantially higher than the first sand at any given number of jolts, the same molding machine being used in each case. After 20 jolts, the first sand has 79 mold hardness and the second 86 to 91 mold hardness. The difference between the two might be described as due to a change in "flowability." It can be said more accurately that the second sand has superior response to jolt molding, i.e. jolt compactability.

To study jolt compactability more specifically, the jolt tube may be fitted with one flat side for determining penetration of mold hardness up from the parting surface. This is done by inserting a strip of metal with a $\frac{7}{8}$-in. wide flat side and the other side contoured to fit the tube. The strip is held in the tube with a screw.

After jolting as before, the specimen is stripped from the tube by first pulling out the metal strip and then pushing it out on a stripping post. Mold hardness readings are then taken at 1-in. increments from the jolted surface (Fig. 5). Data are then plotted as a function of distance from the jolted surface as shown in Fig. 6.

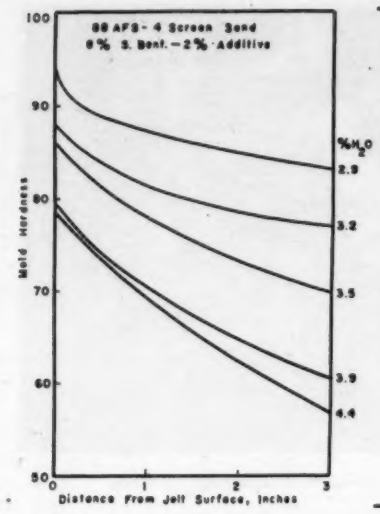


Fig. 6 — Curves show that moisture has a marked effect on hardness up flat side of jolted sand test specimen.

Figure 6 shows how moisture content of a sand affects the penetration of hardness up the vertical side wall from the jolted surface. When performed in this manner, the jolt test is truly capable of measuring flowability of molding sand under jolt molding.

The jolt test is also an excellent means of measuring the jolting efficiency of different molding machines. Furthermore, the effect of the jolt table load on jolting efficiency can also be measured. When jolt effectiveness is being studied, a single sand of known density and predetermined sand column height should be used throughout the tests.

Jolt Test for Molding Machines

Hardness and density of sand developed by jolting depends on the effectiveness of the jolt. The jolt stroke is a mechanical characteristic of the machine. Length of the jolt stroke and the weight of the sand determines the power applied from momentum of the sand at the instant of impact. Sand weight per unit area of projected parting surface depends on sand density and sand height (flask height). The relationship of jolting power to jolt stroke and sand weight is shown in Fig. 7.

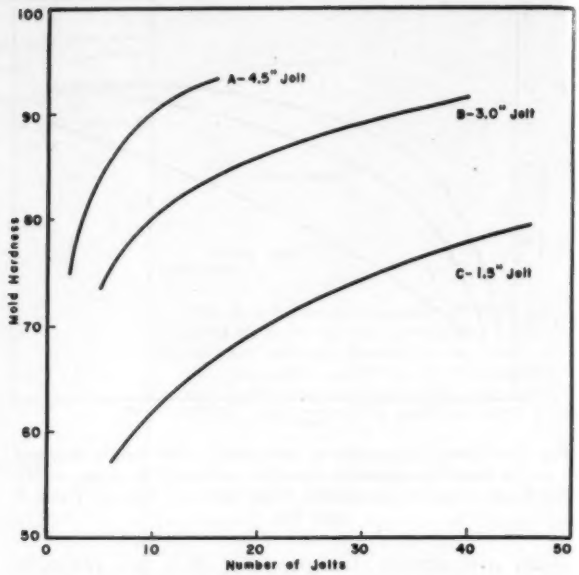


Fig. 8 — Effect of jolt stroke on mold hardness of a sand. Curves A and B are for two different molding machines. Curves B and C are for same machine where 200 lb additional table load reduces 3-in. stroke to 1½ in.

From this figure, it is evident that different mold hardness and density will be obtained from any given sand and column height, depending on the jolt stroke. Included in Fig. 7 is the ramming power of a 2.0-lb ram weight dropped 2.0 in. on a 2.0-in. x 2.0-in. diameter sand specimen. Power supplied by the 2.0-lb ram weight is of similar magnitude as the molding machine.

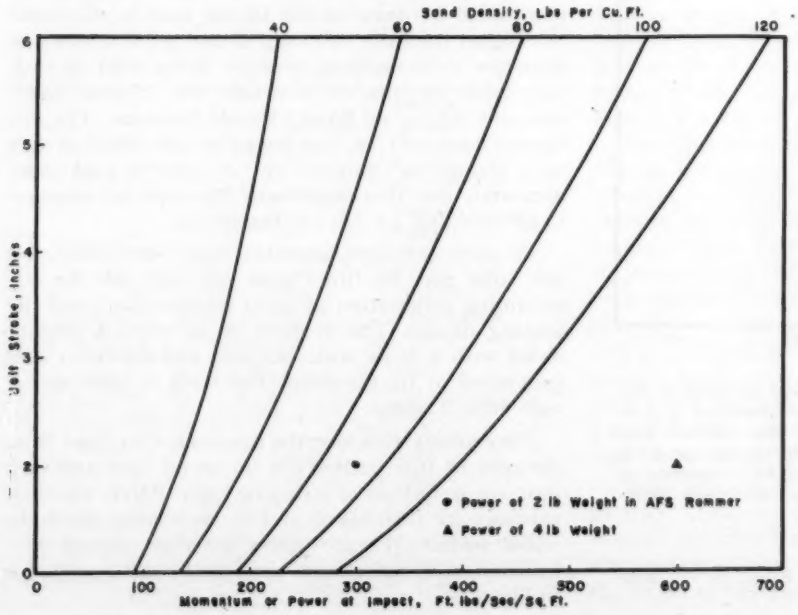


Fig. 7 — Power applied to base of a 12-in. sand column as a function of jolt stroke. Multiply by x ÷ 12 for a sand column x inches high. Sand density increases with impact (jolt stroke, sand weight).

The effect of jolt stroke on mold hardness is shown in Fig. 8. The sand referred to in Fig. 4 was also used for the tests reported in Fig. 8. Curves A and B show data obtained on two different molding machines of 4.5 and 3.0-in. jolt stroke respectively. Curves B and C were obtained from the same machine at jolt strokes of 3.0 and 1.50 in. respectively. In the latter case, the jolt stroke was reduced from 3.0 to 1.50 in. by putting 200 lb additional weight on the jolt table to simulate a large flask and heavy sand and pattern load. The effect of table load on jolt stroke of three different molding machines is shown in Fig. 9.

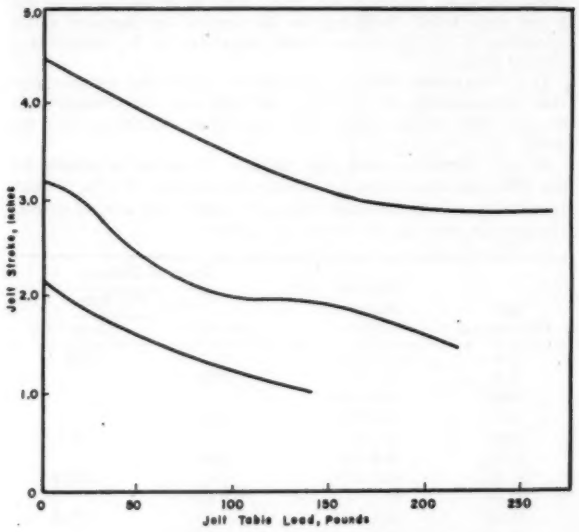


Fig. 9 — Effect of table load on jolt stroke for three different molding machines.

It is quite evident from Figs. 6 and 7 that it is inherent in the machine to produce low mold hardness (and density) when the table is overloaded. The sand should not be blamed for a condition peculiar to the mold. Obviously, overloading molding machines will lead to inferior molding practices.

Summary

Differences in the response of molding sands to jolt molding can be measured with the jolt test. Through mold hardness and density measurements, the jolt test is capable of revealing the effects of sand type, moisture content and additives on jolt compactability, or flowability. Correlation of green sand properties with jolt molding requires the use of a 2-lb rammer weight. A single molding machine of fixed jolt characteristics and a fixed sand column height, 12.0 in. are required for comparison, when the molding sands are being studied.

The jolt test is also an excellent means of measuring the jolting efficiency of different machines. Furthermore, the effect of the jolt table load on jolting efficiency can also be measured. When jolt effectiveness is being studied, a single sand of known density and predetermined sand column height should be used throughout the tests.

Particularly observe the effect caused by overloading the molding machine, that is using too large a flask on a molding machine. It is interesting to note that the length of jolt stroke is shortened by increasing the weight on the jolt table (the sum of the weight of flask, pattern and sand in the flask). These should be taken into consideration in the purchase of a molding machine, as it is the length of the jolt stroke and the solid impact that governs the ram of the sand in the flask.

In conclusion, the interaction of sand and molding machine should be recognized. The sand cannot be expected to do what should be accomplished by the molding machine. Likewise, sands with the best jolt compactability will produce superior results from any particular machine.

DISCUSSION

Chairman: E. C. TROY, Riverton, N. J.

Co-Chairman: W. R. MOGGRIDGE, Ford Motor Co. of Canada. Windsor, Ont., Canada.

Secretary: H. F. BISHOP, Naval Research Laboratory, Washington, D. C.

F. S. BREWSTER[1] *(Written Discussion):* This paper appears to be a continuation of the paper "Does Sand Testing Give Us The Facts?" published on pages 408-414 by the same authors. Therein the statement was made that the standard AFS rammer does not produce a molding condition comparable to that of a jolt squeeze machine and should not be used for correlation of sand properties between the laboratory and foundry. The proposed test appears to be a jolt test only and as such valuable for studying the reaction of sands to jolting. The effect of wall friction is so predominant in a long 2-in. cylinder that the jolt tube test should be useful to show ease of jolting in confined places.

If the standard dropping weight rammer cannot be used for jolt squeeze work, would it not also be improper to use only a jolt test in predicting the results of a jolt-squeeze, air rammer, hand rammer, slinger, high-pressure diaphragm, blow, or blow-squeeze action? Should such a test gain general acceptance it would be well to standardize each part of the equipment used in the test so that we don't revert to the days when everyone had a different test. Perhaps the standard rammer could be fitted with easily changeable weights to give greater flexibility. There is a foreign rammer design, allowing quick removal of the weight, that might be adapted to our use.

The authors have a well-taken point in emphasizing the need to make correlation between laboratory and foundry. This has also been brought out by Dietert and Graham in the 1953 AFS TRANSACTIONS, page 490, "Ramming of Molding Sands." Here a piece of laboratory equipment was proposed which could be used to form specimens of reasonably uniform and controllable density throughout at any mold hardness desired. The method, basically a double end squeeze, minimized wall friction, and could easily be standardized for universal use.

Some years ago a flowability test was proposed where in a square foot of sand 1 in. thick was jolted on edge after which one side of the mold was removed and mold hardness taken over the remaining depth of sand. It is also possible to produce a series of 2-in. specimens in the standard rammer at random increments of mold hardness so that a property vs mold hardness chart can be drawn. Thus one can determine actual mold properties knowing only the hardness of the mold itself.

The authors are to be complimented on emphasizing the need for more accurate information on mold characteristics.

Apropos of this paper's title, better molding practice is currently being accomplished by using higher pressure to eliminate the low hardnesses and densities quoted.

MR. HEINE *(Reply to Mr. Brewster):* As stated, this test is useful for testing response of sands to jolting, not to other methods of molding although some correlation to other testing methods exist. Obviously, when the squeeze pressure for molding is low, the mold is mainly made by jolting. Response to jolting then is a major property in flowability of sand. Wall friction does not seem to be so important since tubes larger than 2.0 in. were studied with no particular difference noted.

1. Director of Research, Brumley-Donaldson Co., Los Angeles, Calif.

V. M. ROWELL[2] *(Written Discussion):* In order to accurately predict the influence of molding sand on castings quality it is necessary to examine the physical properties of the sand as exhibited by a test specimen rammed approximately the same as the mold. The AFS standard test specimen was developed in 1923 and was intended to represent a mold properly rammed by a skilled molder. Obviously individual molding practices, influenced by the character of the sand and castings, result in molds rammed harder as well as softer than the standard specimen.

A large percentage of molds are rammed to properties approximating those of the AFS 2-in. by 2-in. cylindrical test specimen. The test described in this paper would be difficult to reproduce because of several factors not mentioned by the authors.

1. The term "common jolt machine" is vague. Molding machines vary considerably in jolting efficiency. These variations include speed of air release, internal friction and the possibility of air leaks as well as the loading factor described in the paper.

2. The conditions existing in a tube of 2-in. inside diameter are far different than those prevailing in a flask containing sand to the same depth. The large difference in ratios of mass of sand to surface area of the container mean that the sand jolted in the 2-in. diameter tube is bound to be rammed less effectively than in a flask. The friction between the sand mass and the long specimen tube will be high even though the interior of the tube is smooth. This factor probably accounts for the relatively minor differences noted between the 6-in. and 12-in. columns of sand. The smooth surfaced, short, standard specimen tube minimizes this effect. (A worn specimen tube does not produce consistent or accurate specimens.)

3. It was stated that after about 30 jolts the mold hardness and density changes are minor. Usually the changes brought about by successive rams of the standard 14-lb weight falling 2 in. are minor after the standard 3 blows. Some peculiarly unbalanced sands may continue to change in properties up to 35 or 40 rams. Continued jolting would undoubtedly show changes upon continued jolting in the same manner.

4. The statement that a 2-lb rammer weight is necessary to obtain sand densities and hardness values prevailing in foundry molds is open to question. Few, if any, foundrymen make molds of this depth with a jolt operation alone. Although the subsequent squeeze operation is most effective in the sand close to the squeeze board, the end result is to obtain, in most cases, a mold hardness at the parting line comparable to that of the standard specimen. The only exceptions would be on those sands that are very low in permeability, high in volatile matter and used on light castings.

One exceptional sand observed personally was rammed into a standard specimen and rammed by the molding equipment used in the foundry. The standard specimen exhibited the following properties: density, 101.2 lb per cu ft; green compressive strength, 12.0 psi; and permeability, 90. The foundry-rammed specimen was rammed to a density of 120.0 lb per cu ft; green compressive strength, 35.0 psi; and permeability of 6.27. These amazing results are not typical but serve to illustrate that the standard 2-in. diameter by 2-in. tall test specimen rammed by a 14-lb weight falling 2 in., three times, as established in the 1920's, is still hard to beat as a universal standard.

MR. HEINE *(Reply to Mr. Rowell):* It is true the jolt action must be standardized to use the jolt test. A 3-in. jolt is common in foundries. Actually, the authors made standard tests using a dead-weight jolt tester where a 3-in. jolt was guaranteed in addition to tests on a variety of machines. Tube size was investigated and a 2.0-in. diam was found satisfactory. In considering mold density, the type of clay in the sand must be specified. A fireclay-bonded sand always rams to high density. However bentonite is more widely used and it causes the low density problem.

T. E. BARLOW[3] *(Written Discussion):* In all three papers in this volume involving the work of R. W. Heine on jolt tests, squeeze tests, etc. I get the impression that 8 per cent southern bentonite or 8 per cent western bentonite was the basis of comparison in almost all cases. This is an unusually large amount of bentonite to be used for standard molding sands. I can't help but wonder whether the same conclusions would be drawn if the more normal value of 5 per cent or 6 per cent bentonite was substituted for the 8 per cent used in this work.

Some years ago, I saw a number of incorrect conclusions drawn because the work was based on sands containing only 2 per cent or 3 per cent bentonite. It was quickly proven that the relationships which existed at that level did not exist at the normal level of 5 per cent or 6 per cent.

The work that has been done by Heine is extremely valuable and I think it would be helpful to know whether the conclusions can be drawn to the same extent and in the same relation with more normal sand practice.

MR. HEINE *(Reply to Mr. Barlow):* A number of clay percentages other than 8 per cent were investigated, including 4 and 6 per cent. The authors were simply more interested in the 8 per cent level. Included in the papers are foundry sands operating at this level and which certainly can be classed as a normal level.

D. C. WILLIAMS[4] *(Written Discussion):* While the authors provide information on the effect of different jolt heights they do not state which height was used when obtaining data for Table. 2.

It is of interest to note that the rate of change in density for the 6-in. and 12-in. high tubes with the number of jolts is quite comparable. Using the data shown in Table 2 the rate of density change per jolt can be shown as follows:

Jolt Increments	Density Range, lb/ft³	Rate of Density Charge Per Jolt	
		6-in. tube	12-in. tube
0-5	64.9-48.3		3.32
0-6	68.2-48.3	3.32	
5-10	71.5-64.9		1.32
6-11	74.8-68.2	1.32	
10-20	77.5-71.5		0.6
11-21	79.9-74.8	0.51	
20-35	81.6-77.5		0.273
21-36	82.8-79.9	0.193	
35-50	84.5-77.5		0.193
36-52	85.8-82.8	0.188	

If, for the 6-in. tube, one subtracts the rate of density change, for the 0-6 jolt increment, from the density at 6 rams (68.2) one finds the density after 5 jolts in 64.88 lb/ft³ and this is sufficiently close to the density of 64.9 lb/ft³ for the 12-in. high tube jolted five times. If mold hardness reflects the density condition and the densities of the sand in the two tubes is the same it becomes difficult to see why the hardness curves of Fig. 2 should not coincide at the 5-ram level. Perhaps the authors would expand their information on this point.

It is difficult to evaluate the data in the article not knowing the details of the sand mixture composition, the sand mixing equipment employed, and the order of addition of ingredients to the mixture. If this information can be revealed it would be of great value.

The data plotted in Fig. 4 is very interesting, however, one vital piece of data is missing. For the mixture tested at 4.0 and 4.5 per cent water what is the water content for this mixture at which maximum green strength occurs? Along this same line the authors recommend that the weight of the rammer weight be decreased from 14 lb to 2 lb. No data have been supplied to show that this change does or does not influence the type of curve one obtains for the water content-green compressive strength relationship.

In Fig. 3 the absissa is given in number of rams. At this time it might be appropriate in the label to include "using a 2-lb rammer weight." Also in Fig. 3 the slope of the curve for the ram-green compressive strength is almost uniform throughout the length. It seems important to know if the same slope would be maintained if the number of jolts is extended above 60.

It is indicated in the paper that in order to prevent a molded mass from slipping from the tube a force was applied

2. Executive Vice President, Harry W. Dietert Co., Detroit, Mich.
3. Sales Manager, International Minerals and Chemicals Corp., Eastern Clay Products Dept., Chicago.
4. Professor, Ohio State University, Dept. of Met. Engr., Columbus, Ohio.

to the top side. This is comparable to squeezing which produces an added friction component at the tube wall near the top. No information was provided as to what effect such an applied force had upon density of the test specimen or what effect was obtained on further jolting as compared to tests made without the applied force at the top.

The authors state, "With another sand of better response to jolt molding (higher flowability) the 12-in. column will have a higher average density than a 6-in. column." This discussor was unable to locate any data in the article to substantiate this positive claim. Perhaps the authors would provide such data.

The authors state, "However, the 6-in. column of sand has a higher average density than the 12-in. column because there is less density gradient from top to bottom in the former." If this is true, data and method by which it was obtained should be readily available and should be given.

In the authors' next sentence it is indicated that the internal friction of sand grains is greater in longer columns of molded sand. How this can be escapes this discussor and perhaps the authors would explain this point for this is a very important consideration.

MR. HEINE *(Reply to Prof. Williams):* The paper should state the jolt height employed for Table 2 as 3.0 in. The hardness values at the parting do not coincide for 6- and 12-in. heights at the 5-jolt level because the mold hardness test is a surface hardness measurement. The 12-in. column delivers more power to the surface and it is therefore harder. This paper does not deal with ingredients of sand mixtures because its purpose is to describe a testing procedure, not to discuss the many variables which may be studied with the test. The latter can be done in future papers.

THE GENERAL MOTORS BLOW-HOT PRESS AUTOMATIC SHELL MOLDING MACHINE

By

R. S. Amala,[1] J. H. Smith,[2] A. L. Boegehold[3] and R. F. Thomson[4]

ABSTRACT

This paper describes a method of shell molding which employs blowing techniques for making large uniformly contoured shell molds and a hot pressing process for curing these molds. The new molding method is called the blow-hot press shell molding process.

Also described in this paper is the design and construction of production model molding machines which operate on the blow-hot press cycle. Advantages of high productivity and improved dimensional quality of castings have been proven in volume production using automatic blow-hot press shell molding machines to make shell molds for automotive camshafts.

The shell molding concept has considerably accelerated the evaluation of the foundry from a craftsman's workshop to a highly mechanized production facility. A new type of shell molding machine which promises to hasten the trend has been invented, designed, built, and placed into production through the cooperative effort of the Central Foundry Division and the Research Staff of General Motors Corp.

The machine, to be described in this paper, was developed to produce more shell molds per pattern equipment per hour than previously available machines. It has been used in production to make camshafts and rocker arms, and experimentally has produced satisfactory shell molds for automotive crankshafts and fractional horsepower motor end frames.

The general operation of the blow-hot press (BHP) machine is shown schematically in Fig. 1. The operation consists of five steps:

1. The heated pattern is placed below a blow-head with a pattern-matching contour as shown in Fig. 1 (A).
2. A resin-sand mixture is blown into the cavity between the pattern and blow-head.
3. The pattern is retracted, and with an adhering sand resin mixture is moved into contact with a heated contoured head, Fig. 1 (B), where pressure is applied to facilitate curing.
4. After curing the pattern and shell are retracted from the press-head and the shell mold is stripped from the pattern, Fig. 1 (C).
5. The pattern is returned to the blow station for a repetition of the cycle.

1. Supervisor, Metallurgical Engineering Department, Research Staff, GMC, Detroit.

2. General Manager, Central Foundry Division, GMC, Saginaw, Mich.

3. Assistant to Vice President, Research Staff, GMC, Detroit.

4. Head, Metallurgical Engineering Department, Research Staff, GMC, Detroit.

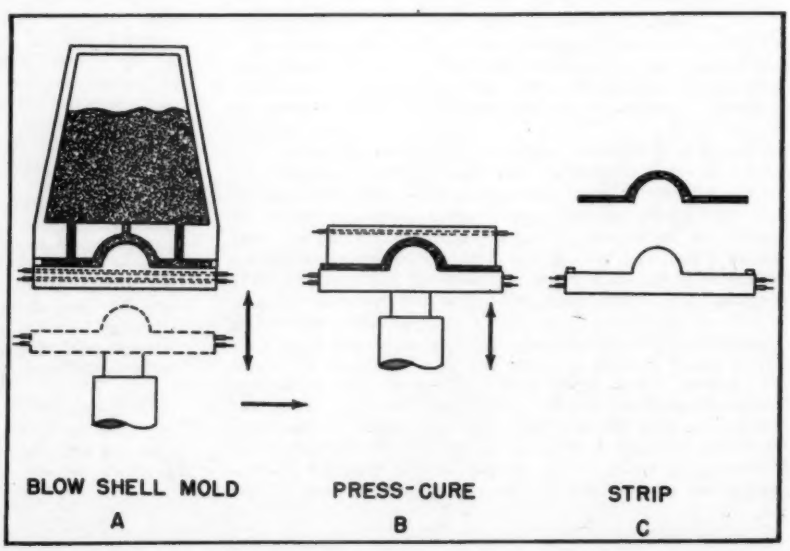


Fig. 1 — Schematic drawing showing essential operations of the BHP shell molding machine.

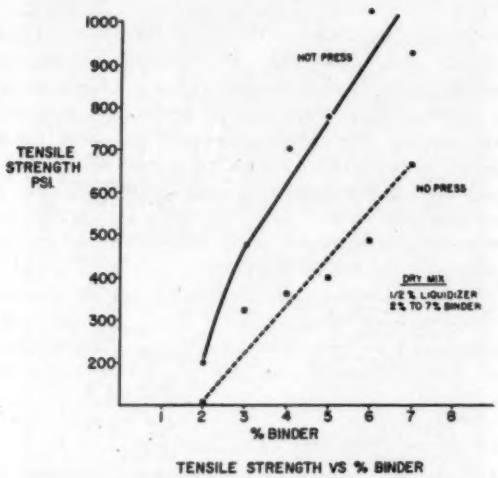


Fig. 2 — Effect of resin content on tensile strength.

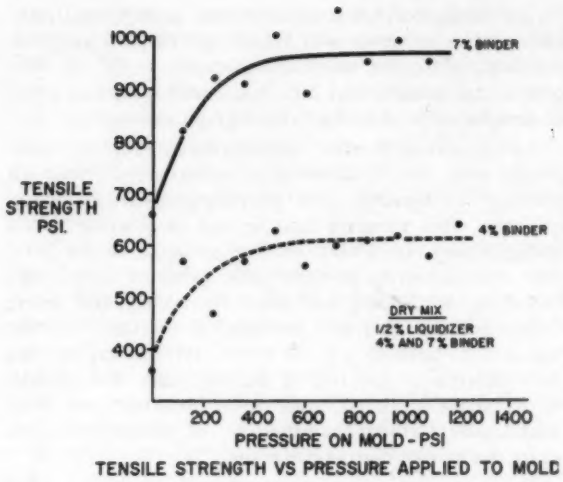


Fig. 3 — Effect of hot pressing pressure and binder content on shell-mold tensile strength.

Shell Molding Mixtures

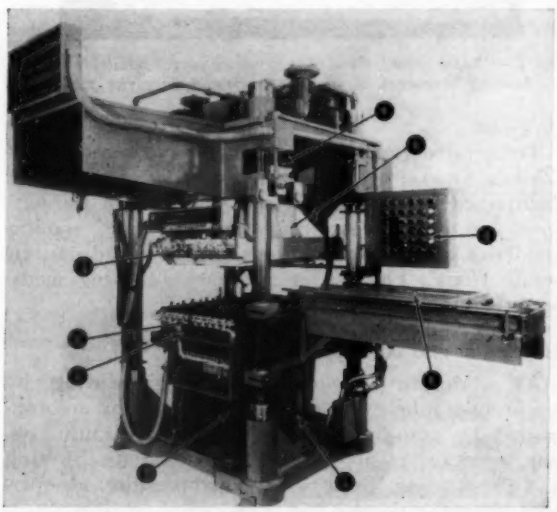
Important to the successful operation of the BHP shell molding machine has been the development of a shell molding mixture compatible with blowing and press-curing. Increased tensile strength was used as an early criterion for bench testing. The effects of resin content and pressure are shown in Figs. 2 and 3.

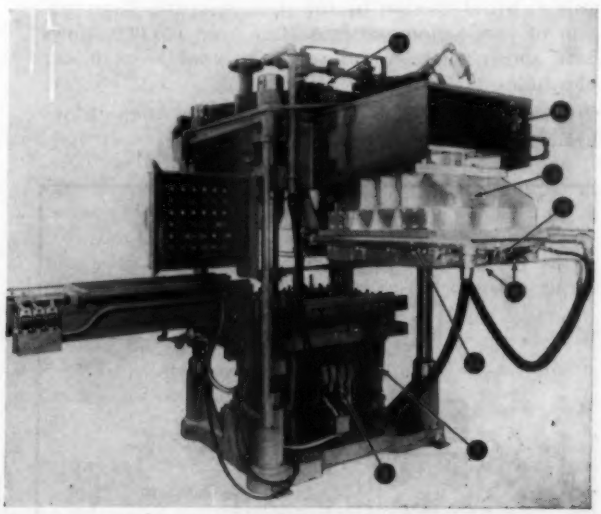
Resin-coated sands, properly adjusted to provide green strength, have performed most successfully in blowing and press-curing. Shells have been made at the rate of 120 per hour. Thermal shock problems, however, have delayed their acceptance as a BHP shell-molding material. Thermal shock which is characterized by mold cracking on pouring is believed related to differential expansion of hot and cold sections of the mold as it is being poured.

Factors which govern thermal shock are the expansion coefficient of the sand particles, bonding film thickness, points of contact where bonding exists, degree of cure of the resin, and the geometry of the mold cavity. Additions of cushioning agents such as wood flour, expanded volcanic ash and other fibrous or cellular materials to shell mixtures have had little effect on the reduction of mold cracking due to expansion. Resin plasticizers in shell mixtures have reduced thermal shock cracking.

This problem of thermal shock cracking causes most difficulty in molds for long castings, where thermal expansion is additive over a considerable length of shell mold, such as in camshaft molds.

It was soon determined that an intimate mixture of sand, resin, and other ingredients possessed su-





BHP shell molding machine: (1) Press-head and press-head heater plate, (2) Pattern heater plate, (3) Stripper plate, (4) Hydraulic elevating and pressing cylinder, (5) Auxiliary elevating cylinder, (6) Mold carry-out, (7) Control panel, (8) Press-head holder, (9) Press-head carriage, (10) Pattern table limit switches, (11) Pattern table guide posts, (12) Vent blow back manifold, (13) Blow-head (cope and drag), (14) Blow-head water cooling coils, (15) Blow magazine, (16) Carriage conveyor cylinder, (17) Blow valve.

Fig. 4 — Front View

Fig. 5 — Side View

perior thermal shock resistance over coated sand mixtures. A formulation was finally developed which is satisfactory for automotive camshafts made by this process using bank and lake sand with sufficient resin to provide desired strength for a given casting.

In appearance this formulation, properly prepared, may be classified as a semi-coated type. In addition to bonding and providing essential curing qualities, this mixture also served to provide green strength and blowability. Green strength in the mixture is required to prevent blow tubes or holes from emptying or flowing free after each blow and, more important, to provide mechanical strength to the mixture to permit it to be blown on vertical pattern details without sagging or falling after the pattern with partially cured sand resin mixture has been withdrawn. Catalysts are used in the mixture to accelerate the rate of cure of the resin.

Dry type shell mixtures can be adjusted with addition agents to make them blowable, but cure times by the BHP process have been prohibitively long.

Machine Construction and Operation

The final design and construction of the machine is shown in Fig. 4 and 5.

Blow and Press-Heads

The aluminum blow and press-heads were cast in dry-sand molds faced with zircon sands. Separate sets of patterns were required to cast the blow and press-heads to compensate for the expansion of the press-head on heating to operating temperatures. No machining of the pattern contours was necessary except for polishing. One-half inch blowholes were drilled into the blow-heads and counterbored in the back to reduce friction losses on blowing. A ratio of slightly less than one blowhole to one vent was used for a total of 50 blowholes and 60 vents per head. Blowholes are located in positions away from pattern details to avoid erosion by the shell mixture. Examination of production patterns after over 100,000 blows have shown little or no wear on stool areas under blowholes.

Illustrated in Fig. 6 is a sectional view of the blow-head, blow plate and magazine assemblies.

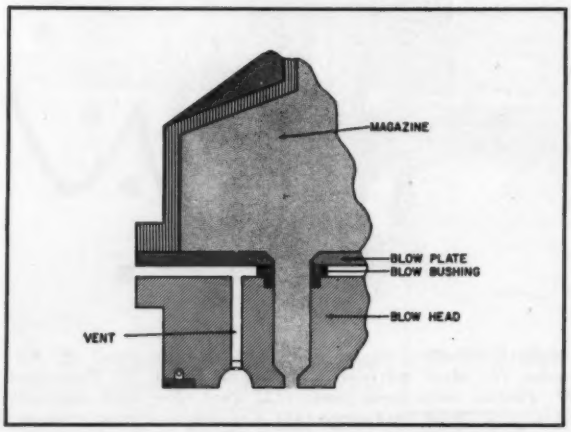


Fig. 6 — Schematic cross section through blow-head and magazine.

Pattern Plates

Four-on gates of camshafts are mounted on a 1¼-in. thick cast iron stool. The pattern stool is bolted and doweled to the heater platen which is bolted to support rails on each side of the pattern table. Bolt holes in the heater platen are elongated to allow for expansion on heating to operating temperatures. Guide pins between the cope and drag patterns locate the blow and press-heads with the pattern during their operations. A metal frame around each pattern controls the blown shell's thickness and acts as a seal between the blow-head and pattern during blowing. Cartridge heaters located in the sprue are required to maintain set temperatures during production operation. Temperature control is afforded by indicating-controlling pyrometry.

Pattern and Press-Head Heating

To provide adequate heat to sustain a designed production capacity of 90 molds per hour, tubular heating elements were cast integrally with the pattern and press-head heater platens. Figure 7 shows a mold for these castings with tubular elements in place. This cast-in construction provides maximum heat transfer rate at lower element sheath temperatures.



Fig. 7 — Sand mold used for making press-heads, showing heating elements in place prior to closing the mold.

Total rated capacity of the heater platens is over 125 kva. Operating temperatures of 500 to 600 F are maintained by thermal switches mounted in the patterns and press-heads. Insulation and water cooling maintains units adjoining the press-heads and pattern below 170 F. The use of alternative heating methods is being investigated.

Hydraulic and Air Operated Units

All motion in the molding machine is provided by air or hydraulic cylinders whose operations are controlled by auto-timed circuitry. The hydraulic system, used for raising the pattern table for blowing and press-curing, consists of a dual-pressure, air-operated hydraulic booster and a hydraulic cylinder. With this combination, 100 psi oil pressure is available for rapidly raising the pattern table and 3000 psi oil pressure is available for clamping for blowing and for press-curing. Auxiliary air cylinders aid in raising the table to the blow and press-cure position.



Fig. 8 — Electrical console of BHP machine showing timers, relays and wiring circuits.

Air motivated units are controlled through plumbing circuits consisting of solenoid-operated valves, pilot-controlled valves, and quick opening and closing valves. Regulating valves control the operating speed of large components such as the blow and press-head assemblies. Magnesium was used in the construction of the blow magazine and the press-head mount to minimize inertia effects.

Duration and sequence of operations are controlled at the electrical console located away from the molding machine. Thirteen operations in the molding cycle are time regulated by adjustable timers mounted on the door of the electrical console (Fig. 8).

Proper sequence of operations is governed by mechanical and electrical interlocks located in the electrical console and on the molding machine. Limit switches insure proper positioning of movable units for each succeeding operation.

Blow Valve

Standard foundry blow valves were used in the construction of the BHP machines. Modifications to the blow valves vent or pressure release valve were made to provide a rapid pressure drop in the blow valve body after blowing. Additional quick opening vents were also placed in the throat and lower sections of the blow magazine. Opening and closing of vents are automatically timed with blow cycle.

Cycle Mechanics

The three basic steps of the BHP molding cycle are (1) blow to form the shell mold on the pattern, (2) press-cure with a heated contoured head, and (3) strip to remove shell from the pattern. On the production model, all operations of the machine can by cycled manually at the control panel or the entire molding cycle can be set to "automatic" to be repeated after each mold is discharged.

In operation the pattern is elevated to contact and seal the magazine blow assembly. The shell sand mixture is then blown into the space between the contoured blow-head and the pattern plate (500-550 F). After blowing, and the magazine has been vented, the pattern is lowered, the press head, similar in contour to the blow-head, is conveyed on a roller carriage to the press-cure position. The pattern with the uncured shell is raised and pressed (300 psi) against the heated head (500-550 F) for curing.

After curing, the pattern is lowered to an intermediate position where the stripping pins lift the cured shell from the pattern. Mold conveyor bars are then inserted between the pattern and the shell after which the pattern lowers to the start position. The conveyor then removes the mold halves for assembly. Shell sand mixture is automatically fed into the magazine from a hopper during the press-cure cycle.

A typical production cycle is as follows:

Operation	Time, sec
Raise pattern for blow	5
Duration of blow	1
Delay for venting	5
Lower pattern	4
Transfer presss-head	4
Raise pattern for curing	5
Press-cure	25
Lower pattern for strip	2
Strip	2
Insert carry out	3
Lower pattern to start	2
Remove carry out and molds	3
Return blow-head to start	4
	65

55 molds per hour

The time cycle of the various steps can be modified as required to meet production requirements related to the machine. The total cycle may be, on occasion, somewhat more or less than this, as required to produce satisfactory shell molds.

Production Experience

Several of these machines have been built and placed in production. Figure 9 shows one of three of these machines in a production line at Central Foundry Division. The camshaft shell molds are shown in the foreground on the conveyor.

More than 120,000 camshafts were produced on one machine in 1955 (partial year's operation). Two machines have consistently produced between 40 and 50 molds per hour per pattern for 16-hour operation during the past three months.

Figure 10 shows shell molds made in production and a casting from these molds.

Limitations

Because the process requires auxiliary contoured equipment (blow and press-heads) it has been generally regarded as suitable only to large volume production applications. Some casting or pattern designs may be such that the production of each head may be readily cast or machined without incurring too great additional cost.

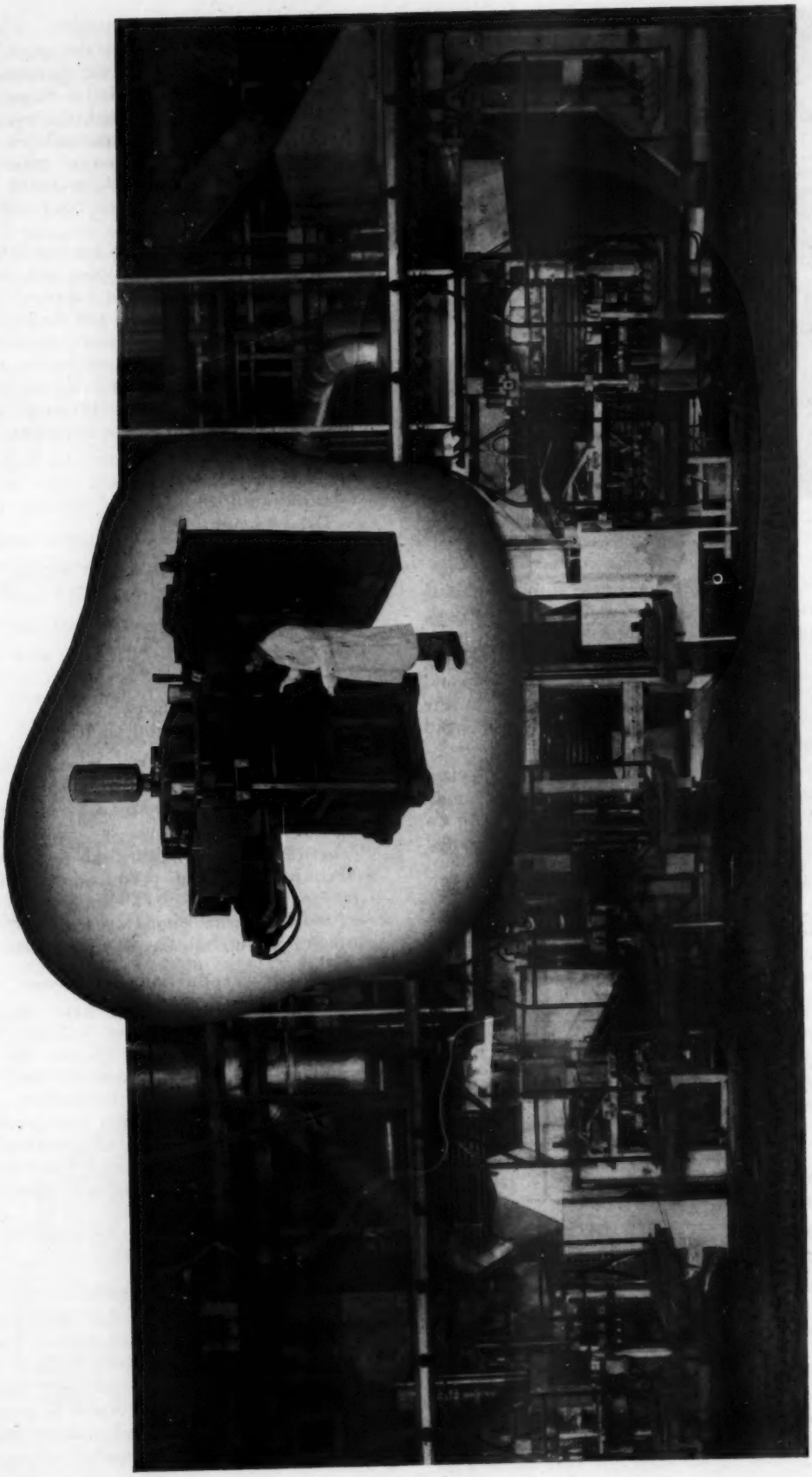


Fig. 9 — Installation of BHP machine in production line at Central Foundry Division of General Motors Corporation.

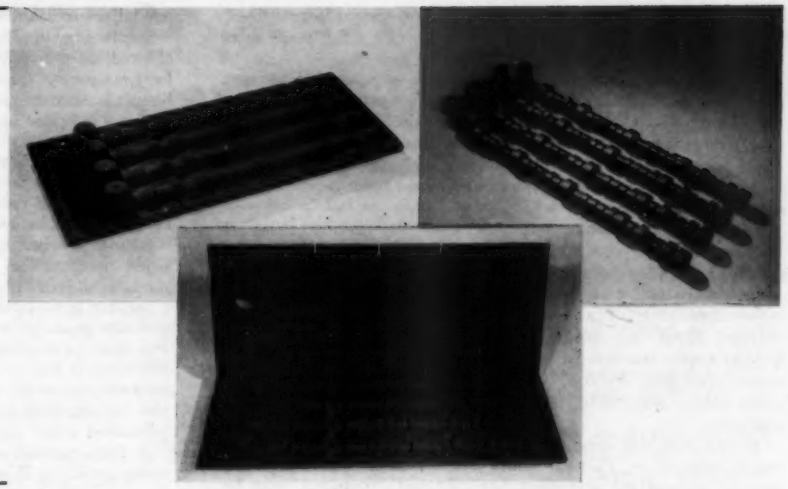


Fig. 10 — Shell molds and castings produced by BHP machine.

Press-curing is most effective on shell molds or pattern designs whose geometry are generally flat or moderately contoured. Since curing is effected by heat conduction through contact, vertical sides receiving less contact pressure would cure at lower rates than high pressure contact positions.

Conclusions

1. A significant production advantage of rapid mold processing has been realized by the application of the BHP principle to the making of automotive camshafts. The shortened cycle has accrued from these two steps:

a. Shell mixture blown to form the mold in one to three seconds. Conventional practice requires 12 to 45 seconds for the investment.

b. The "blown" mold is cured by pressing a heated contoured head to contact the mold uniformly. Cure time reduced to 12 to 25 seconds from the conventional 35 to 75 seconds.

2. Casting dimensional qualities have been improved. Warpage of cast camshafts has been reduced because the molds are uniformly supported on leveling pads during pouring which are blown as an integral part of the drag mold. These pads prevent mold sagging during pouring.

3. Thinner shell molds, resulting in decreased resin costs, are possible because of the uniform thickness of the shells produced by the contoured blow and press-heads.

Acknowledgment

There are many people in General Motors Corp. to whom much credit is due for the successful development of this machine. Special cognizance should be taken of the work of M. J. Diamond, W. S. Hackett, C. A. Koerner, and F. J. McDonald, all of Central Foundry Division, and H. I. Duryee, C. F. Froberger, D. J. Henry, N. I. Musser, W. E. Nixon, and R. E. Schroeder, of the Research Staff. The authors wish to acknowledge the excellent assistance of Messrs. E. R. Cprek and W. M. Gruszczynski in preparing photographs and charts for this paper.

DISCUSSION

Chairman: JAMES THOMSON, Continental Foundry & Machine Div., Blaw-Knox Co., East Chicago, Ind.

Co-Chairman: K. N. SMITH, Caterpillar Tractor Co., Peoria, Ill.

G. A. CONGER[1] *(Written Discussion):* I would first like to congratulate the gentlemen who have engineered and perfected this machine. Their contribution to the knowledge of shell technology is appreciated by all foundrymen.

1. I assume that conventional pressures were used for blowing, 30 to 120 psi, and the need for a release agent should be similar to usual shell procedures. With the application of pressure at the press head, the density of the shell should be increased as witnessed by the higher tensile strengths of Fig. 4. One might then expect that the shell would have a greater tendency to seize the pattern producing a stripping problem. This should be more pronounced at higher press head pressures.

Is it therefore necessary to take additional steps to insure release such as special pattern treatments, additives in semi-coating or stripping green?

2. The beneficial effects of hot pressing as measured by tensile strength are illustrated in Fig. 2. At *a* resin level the tensile strength can be doubled by hot pressing (the pressure for which was not indicated) as:

	Tensile Strength, psi	
% Resin	No Press	Hot Press
2	100	200
4	350	700
6	475	950

It is also apparent from Fig. 3 that the tensile strength increases with increasing hot press pressure up to approximately 500 psi. Over 500 psi the tensile strength does not change appreciably.

Does the shell density vary in a similar fashion? Pressing should further densify the shell-pattern interface thereby minimizing wrinkles and porous areas to produce a superior cast surface on vertical pattern surfaces, inherently difficult in dumping procedures. Further, on pressing with a *hot* head, the time of cure has been reduced by one-third to one-half that of "conventional" practice. We also have found this to be the case, when blowing into a heated enclosure, by virtue of more rapid heat transfer by conduction simultaneously from all heated surfaces to the shell. Curing in a complete enclosure, stripping only when curing has been completed should minimize warpage and distortion.

Is there a further shortening of the cure time attributed to the applied pressure, perhaps a lowering of the melt point of the resin?

Also, might a shell defect occur similar to a mold "ram-off"

1. C & S Products Co., Inc., Detroit.

upon the application of pressure and are special precautions taken in design of the pressure heads to move the shell mix toward the vertical pattern surfaces?

3. Reference is made to thermal shock or cracking of the mold prior to solidification such that fins are produced on the casting or in more severe instances, the mold cracks open permitting a run-out. The paper indicates that this difficulty is in some fashion linked with a coated sand. A number of factors effecting thermal shock and requiring a degree of control are mentioned.

A statement is made, "It was soon determined that an intimate mix of sand, resin and other ingredients possessed superior thermal shock resistance over coated sands . . . etc." It occurs to me that this might well be the keynote of this paper and perhaps a major contribution to the process as thermal shock has plagued shell operations from its birth. Undoubtedly thermal shock is more pronounced in coated sands primarily because coated sands usually are made with silica sands noticeably free of cushioning agents and other additives.

Would not the same formulation mentioned, if coated, be more resistant to thermal shock than coated sand? Can the authors expand the statement ". . . properly prepared, may be classified as being a semi-coated type."? I believe a great deal of interest prevails in the treatment of sand with resin commonly termed "coating". I find a new term has been coined, "semi-coated". One might deduce that the resin dispersement and physical form is midway between a pure mechanical mixture as found in dry blending and a continuous resin film about sand particles found in coating.

Can the authors give a typical semi-coating procedure?

4. The blowholes used were ½ in. in diameter. It was also stated that a ratio of slightly less than one blowhole to one vent was used.

What size, what type and what was the effective area of the vents employed? Are vents used both on the blow head and pattern plate and might one say that venting of the box is similar to venting conventional blower boxes?

5. Heating of the equipment is accomplished by casting around tubular type electric resistance heating elements. It should be noted that this provides for total immersion of the element such that heat is conducted more or less uniformly away form the elements. This procedure is consistent with maximum efficiency, longest life and proper engineering application of resistance heaters. Shell core or mold pattern installations employing electric resistance heaters mounted in a fashion whereby heat is conducted away from only a portion of the heaters surfaces usually fail prematurely. An optimum situation is illustrated by the total immersion of a heater in a fluid where all effective heating surfaces have equal opportunity to give off heat.

It is stated that over 125 kva is required to provide heat to maintain temperatures of 500 to 600 F. What is the size of the heater platens employed or more basically, what watt density is supplied? Also, how much time is consumed in initial warm up from room temperature to operation temperature?

It is further noted that care must be taken in the original pattern equipment design such that all the components of blowing into a hot pattern are accounted for. Placement of blowholes, vents, pattern surfaces, strip pins and heater elements must be considered simultaneously in the original engineering as one lacks the opportunity to move the components after the heaters have once been positioned in the platens.

I believe the process and equipment discussed in this paper constitute a major refinement of the shell molding process.

MR. AMALA *(Reply To Mr. Conger)*: Our replies are numbered to correspond to the questions or remarks numbered in the above discussion.

1. With properly designed and maintained pattern equipment no special treatments to the pattern or shell mix are required to avoid "shell sticking." As like most shell mixes which are to be blown the release agent, which is wax, is included in the mix formulation.

2. Shell density does increase with increases in pressing pressures. Little or no improvement in casting finish is noted, however, because shell surfaces adjacent to the pattern become partially cured and rigid before pressure is applied.

Shortening of the cure by the application of pressure is effected only by the improvements in heat transfer from the heat source to the shell. In the opinions of the authors very little pressure is exerted on the resin since during its cure it becomes a liquid and is free to flow into cavities between sand grains.

Only minor "ram-off" defects have been noted of the castings being made by the BHP process. A "ram-off" condition might become serious if the design presented high vertical sides which could not be beneficially press-cured.

3. The absence of thermal shock in the particular shell mix being used can be credited to the use of impure or bank-type sands and resin plasticizers. "Semi-coated" may be regarded as a generic term used to describe mixes in which the resin exists as finite particles bonded to individual sand grains. The bonding agent in this instance is a liquid resin and a resin plasticizer. Procedures for preparing semi-coated sands are similar to methods currently being used to prepare dry mixes except that the liquid additions are made prior to the addition of the powdered resin.

4. Conventional venting practice was used on the blow heads using ½-in., ⅝-in. and ¾-in. brass slotted vents (0.014-in. slots). No vents are used on the pattern plates.

5. The heater platens used are approximately 35 in. x 35 in. x 2 in. The watt density for all radiating surfaces of the platens is about 22 watts per square inch. Over one half of platen area is insulated to provide a maximum heat input to the pattern. In production about 25 minutes are required to heat the patterns to operating temperatures.

G. KOREN, R. MITCHELL and J. JAPUNTICH[2] *(Written Discussion)*: The authors are to be highly commended on reporting the extensive work done in development of this phase of the shell-mold process.

The authors state that "the machine was developed to produce more shell molds per pattern equipment per hour than with previously available machines" and this certainly supports our basic feeling that the shell mold process finds its best adaptation in the field of high production repetitive jobs with a definite limitation on the size and shape of casting.

The authors state that "Important to the successful operation of the BHP shell molding machine has been the development of a shell molding mix compatible with blowing and press-curing."

In this connection the following questions arise:

1. What constitutes a semi-coated sand?

2. What is the nature of the ingredients mentioned, other than the sand and resin which produce the improved thermal shock resistance over the coated sand mixes?

3. Assuming that plasticizers are used, do they provide the necessary green strength and how much?

4. What effect does the resin plasticizer have on the cure time?

5. What is the net result in the final cost of the mixture of the use of plasticizers and catalyst?

In connection with some of the mechanical and design features, the following questions arise:

1. The blowhole shown in Fig. 6 is a straight-through, rather than an offset hole. Has an advantage been found in the straight-through hole over the off-set blowhole which is used in coated sands without green strength?

2. Is there any problem of short life involved in the use of the integrally-cast heaters?

3. Is the heater the limiting factor in the 500-550 F temperature which is used for press-curing?

4. Is the 500-550 F temperature adequate to get the shortest and thus most desirable cure time?

5. How is the reduction to "12 to 25 seconds" from the "35 to 75 seconds" accomplished with the 500-550 F temperature?

6. Are we correct in assuming that the 300 psi pressure used is a compromise, inasmuch as the curve in Fig. 3 shows considerable advantage of 450 or 500 psi over the 300 psi used?

7. What provision is made for the release of the gases produced during the press-cure cycle?

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2. Pettibone-Mulliken Corp., Beardsley & Piper Div., Chicago.

MR. AMALA *(Reply to Messrs. Koren, Mitchell and Japuntich):* Our replies are numbered to correspond to the questions numbered in the above discussion. Replies which have been omitted have been covered in the reply to Mr. Conger.

3. Plasticizers are used in the mix to provide green strength. For the application described in the paper about 1.5 to 1.8 pounds green compression as measured on standard biscuit, is required. The amount of plasticizer required to produce this green strength will vary depending upon grain fineness and sieve distribution of the sand being used and on the amount of resin used.

4. Resin plasticizers used in quantities required to produce green strength decrease the cure rate of the shell mold.

5. The net final cost of a shell mix which employs plasticizers and other addition agents depends upon the casting requirements of the part to be made. In general, for similar applications, blow hot press type mixes are of a lower cost than mixes used in the dump process.

In connection with some of the mechanical and design features:

1. Shell mixes with green strength cannot consistently be blown through offset holes because of packing. Straight holes offer less friction, consequently lower blow pressures can be used. Also, cleaning of straight holes by "rodding" is more easily accomplished.

2. Integrally-cast heaters have a greater service life than do loose fitted heaters if the heater is properly designed and properly used. Some difficulty in production has been experienced where carbon-rich gases have condensed inside the element sheath to cause grounding.

3. The heater is not the limiting factor in using 500-550 F press-curing temperatures. Higher temperatures have been used but because of the low heat conductivity of the mold material, surfaces in contact with the press heads and patterns would over-cure and char before the midsections of the mold would be even partially cured.

4. More important than short cure is uniformity of cure. Greater uniformity is obtained at 500-550 F than at higher temperatures.

5. Reduction of cure time is possible by press-curing because optimum heat transfer is afforded by contact between the press-head and shell mold.

6. Pressing pressures not exceeding 300 psi were selected because of the mechanical requirements imposed by the use of higher pressures.

7. No provision is made for the release of gases from the mold during pouring. Enough permeability exists in the mold to allow gases to escape at the mold edges where no pressing is done.

PAD WASHING WITH CARBON ARC PROCESS

By

Frank Newberry*

Carbon arc-compressed air pad washing has reduced grinding and cleaning costs in the finishing department of Oklahoma Steel Castings Co., Inc.

In our plant the castings are brought from the foundry to the cleaning room in tote boxes by five-ton crane. The loose sand and cores not removed by the shakeout, especially internal cores, are removed by an air hammer or by striking the castings manually with a hammer. The castings are then placed on a four-table automatic cycle blast for additional cleaning.

After blasting, the castings are placed on plates on conveyor lines for removal of gates and risers by oxyacetylene torch. After the gates and risers are removed, the castings pass into the burn-out booth for inspection and defect removal. They then move along the conveyor line into the weld booth. After welding the castings are ready for pad removal.

We use the carbon arc process with copper-clad carbon electrode for removal of gates, riser pads, fins, nails, and excess weld deposit. For this operation we use two 600-amp rectifier welding units tied together, and an electrode holder fitted with a ¾-in. head to take ¾-in. copperclad carbon electrodes. The welding machines are about ten feet from the booth. The welding cable passes through a compressed air hose and attaches to the holder.

We use 80 psi compressed air at the torch. Air is turned on before making contact with the metal to be removed so that the melted metal will be blown away. This allows the operator better to observe the contour of the casting. The operator then moves the electrode back and forth in a rhythmical fashion.

We cut our gates and risers on the average of about $\frac{3}{16}$-in. high, which enables us to remove all the metal at one pass. It is seldom necessary to make a second pass to clean the casting. Since the air is on at the time the arc is struck, metal is blown away as rapidly as it is melted and there is no difficulty with carbon pick-up.

Here's what we use in the way of equipment, and some of the costs:



Four-table shot blast removes last of sand prior to torching gates and pads.

*General Foreman, Finishing Dept., Oklahoma Steel Castings Co., Inc., Tulsa, Okla.

Castings on roll conveyor are gas torched to remove gates and risers.



Copper-clad carbon electrode	3/4-in.
Rectifier welding machine	1200-amp
Electrode cost	$0.294 each
Amperage setting	750
Minutes contact time per electrode	7.25
Pounds metal removed per electrode	4.90
Electrode cost per pound metal removed	$0.06
Electrode cost per sq in. pad area, avg. pad thickness 3/16-in.	$0.0032
Power cost per hour operation	$0.30
Contact time	About half the total working time

While the pounds of metal removed as shown on the standards data is 4.90 per electrode, it must be remembered that such things as pad washing, fin removal, and nails are included. On a large piece of metal we found that we could remove ten pounds of metal per electrode.

The electrodes are 12 in. long, and we use them down to approximately three inches. A fixture which we attach to the stub lets us use more of the electrode, thus reducing the cost. As a comparison of costs, the standards data show that 3.2 mills is the cost of electrodes per square inch of pad that has an average thickness of 3/16-in. Our swing frame grinding cost is approximately 4 mills per square inch on the same thickness.

From the curves in our standards department, we find that the carbon arc process with 3/4-in. copper-clad carbon electrode removes the metal approximately 20 per cent faster than swing frame grinding. This reflects the savings in labor and materials. The operator can do a better job holding the contour of the casting than with the swing frame grinder where he is working some distance from the metal.

We would like to use much higher current than 750 amps, since the speed of metal removal is much greater as the amperage is increased. But at present the holders available to us will not stand it.

We make many high pressure castings on which the customers do not allow welding, and we have had complaints from these customers because the area where the metal is removed by the carbon arc process appears to have been welded. However, we have had no difficulty in proving this was not the case.

We use carbon arc equipment immediately following weld repair, and in this way the operator can remove all pads, fins, and nails as well as excess weld deposits. Following this operation we use a hand grinder to slick up the area. Actually, the operator makes no attempt to remove metal with the hand grinding operation. Many of the castings are shipped to the customer without any hand grinding.

We have found it more economical to use copper-clad electrodes than plain carbon or copper-coated electrodes. Based on pounds of metal removed, we have found that the cost of copper-clad electrode is 15 per cent less than either of the other types.

The copper-clad electrode does not heat more than about one inch up from the work. The plain carbon and copper-coated electrode heated up to the holder, causing maintenance on the holder to be greatly increased. There is no breakage on the copper-clad electrode, while breakage on the other types was very high. Since the copper-clad electrode lasts one-third longer, labor costs are reduced in time spent changing electrodes.

Some rectifier units are not suitable for continuous operations with carbon arc equipment. We have had much difficulty with some rectifier units; however, our oscilloscope observations show that those we are now using are well designed and capable of continuous operation. We are well pleased with the equipment and plan to install two more units in our cleaning lines.

PROPERTIES OF REFRACTORIES AFFECT AIR FURNACE BOTTOM SERVICE

By

C. C. Lawson* and L. R. Jenkins**

Refractory life in air furnace bottoms is greater the lower the impurity content of the brick according to a three-year study made by the authors. Service life of eight refractories with different physical and chemical properties was compared with manufacturers' laboratory tests.

The tests were run in two production air furnaces which are part of a cupola-air furnace duplexing system. The furnace bottoms are totally enclosed by a steel shell and the space between the service bottom and floor is filled with sand. The furnaces are old style with bottoms which are not air cooled. Inside dimensions are 7 ft wide by 29 ft long. They are fired with single pulverized coal burners. Cupola iron enters just inside the bridge wall. The tapping blocks are located on one side about 7 ft from the burner wall and 20 ft from where the cupola metal enters.

At the beginning of a refractory life improvement program in February 1953, thermocouples were placed beneath the service bottom in the hottest zone of the furnace. This was found, by experiment, to be in the center of the furnace just past the tapout blocks. The hot junction of the thermocouple was in contact with the heel of the service bottom brick.

The normal practice at this time was to fire each furnace for five consecutive days and then tear it down for repairs. This usually consisted of an entire bottom replacement to insure against failure and resultant production stoppage. Several super duty refractories were tested with no significant improvement. Different methods of laying, regardless how tight, showed no significant life improvement. The failures appeared to be due to joint penetration. The brick were still full size.

The thermocouple study indicated that at the end of the second day the temperature at the heel of the service bottom brick had increased to 1940 F. This was 46 hours after the furnace was first fired off. At the beginning of the 3rd day, the bottom had cooled to only 1860 F and at the end of the 3rd day the temperature had increased to 2080 F. At the beginning of the 4th day, the bottom had cooled to only 1920 F and at the end of this day had increased to 2090 F. At the beginning of the 5th day, the heel of the service bottom had cooled to 1930 F and at the end of the day the temperature had increased to 2090 F again. The firing period is normally 22-23 hours per day.

It was then decided that the two furnaces should be fired on alternate days. This decreased the rate of temperature rise and shortened the time the heel of the brick was at high temperature. With the practice of alternate day firing, the furnace was operated for 16-22 days (354-470 firing hours) before the heel of the brick reached 1950 F. After an extended period of operation (30 days), the temperature occasionally reached 2300 F without apparent danger. The actual time at this temperature was short. The average temperature peak during this period was 2000 F. There were no more production stoppages in mid-day due to the bottom floating.

Caution! The thermocouple protection tube can become an uncontrollable tapping spout in case of bottom or sidewall failure in the vicinity of the tube. Provision should be made for plugging this hole in the side of the furnace.

The next step in the program was to experiment with heavier duty refractory brick. Those which indicated a reheat expansion were selected because it was felt that a bottom would tighten up rather than become loose and allow joint penetration to take place. This was quite evident in that as time passed, the shell of the air furnace was pushed out, and bent the vertical rail buck stays.

In a normal day of production, the furnace is fired off at about 11 pm the night before it is used. Iron from the cupola enters the furnace at about 4 am and the furnace is tapped first at 6:30 am. Cupola iron runs into the furnace continuously until about 9 pm. The furnace is intermittently tapped until it is empty at 11:15 pm.

There is a considerable period of time during

*General Superintendent and **Metallurgist, Wagner Malleable Iron Co., Decatur, Ill.

COMPARISON OF AIR FURNACE BOTTOM LIFE WITH REFRACTORY PROPERTIES

Refractory	A	B	C	D	E	F	G	H
Melting Point	3335 F	3310 F	3335-3389 F	3308-3335 F	3173-3200 F	3335 F	3300 F	3172-3200 F
PCE	38	37	38-39	37-38	33-34	38	36-37	33-34
Density lb/cu ft	154	157	162-168	144-150	140-147	152-157	140-142	140-142
Permanent Reheat Change	0.1% exp	0.5% exp	0.0-2.5% exp	0.0-2.5% exp	0.5% cont 1.0% exp	0.2-2.5% exp	2.5-3.6% exp	0.5% exp 0.9% cont
Hot Load Strength								
Temperature	3200 F	3200 F	2640 F	2640 F	2640 F	2640 F	2640 F	2640 F
Deformation	0.45%	3.2%	0.5-2.0%	1-3%	3-5%	1.5-3%	2.5-2.3%	3-5%
Panel Spalling ASTM C122-47	0.0%	0.0%	1-3%	1.3%	0.4%	3-10%	1-3%	1.5-4%
Modulus of Rupture ASTM C133-39	1200 psi	2800 psi	1000-1300 psi	900-1200 psi	1000-1600 psi	1200-1600 psi	550-800 psi	600-1000 psi
Apparent Porosity ASTM C20-46	24%	15%	23-27%	23-27%	12-16%	20-24%	19-22%	12-16%
Water Absorption ASTM C20-46	9.8%		9-11%	11-13%	5- 7%			
Chemical Analysis								
Alumina	76.17	61.4	78.0-82.0	68.0-72.0	42.5-45.0	68.9	60.5	41.6
Silica	20.08	35.4	11.0-15.0	21.0-25.0	51.5-53.5	24.3	32.1	52.8
Titania	2.51	2.0	2.5- 3.5	2.5- 3.5	1.2- 2.2	2.9	2.9	2.3
Ferric Oxide	0.87	0.9	1.5- 2.5	1.5- 2.5	1.0- 2.0	1.9	1.8	1.5
Calcium Oxide	trace	trace	0.6- 1.3	0.6- 1.3	0.2- 0.8	0.2	0.4	0.3
Magnesium Oxide	trace	trace	0.0- 0.5	0.0- 0.5	0.0- 0.5	0.3	0.5	0.4
Alkalies	0.08	0.2	1.0- 2.0	1.0- 2.0	0.5- 1.0	1.5	1.8	1.2
Service Life								
Days	44	37	38	30	16	26	26	21 *
Tons	5607	7757	8381	4832	2071	4879	4900	2270
Hours Fired	953	851	874	690	361	598	530	315

* Single shift (15 firing hours)

which the bottom is exposed to a flame without the protection of an iron layer. The normal depth of iron at the center of the furnace near the tap holes is 7 inches. The flame temperature (optical pyrometer) from the burner is about 3160 F to 3200 F. The iron temperature at the spout is 2860 F to 2880 F.

The table lists several types of refractories with their comparative laboratory test properties and observed service life. The data for this table were furnished by the respective refractory manufacturer.

Refractory *A* spalled in areas of about 2 to 3 ft in diameter and in various locations at different times. About one inch of brick depth would break away clean and flat. The thickness of this bottom in the vicinity of the tap holes was reduced to about 2 in. prior to removal and replacement. The joints were still tight and there was no evidence of iron under the bottom. This brick showed a more rapid increase in temperature at the heel which indicated greater thermal conductivity. When used in the sidewalls, the brick spalled excessively.

Refractory *B*, which shows a 61 per cent alumina content, proved to have comparable life to refractory *C* which has an 80 per cent alumina content. Refractory *B* showed superior resistance above all others to erosion when used for lining the cupola front slagger. The reason may be that refractory *B* contains less impurities than the other refractories.

Refractory *D* showed fairly good bottom life. This refractory had superior resistance to flame erosion in the air furnace sidewalls.

Refractories *E* and *H* proved to be inferior to the higher alumina materials. They are kept in stock for rapid, short time, bottom repairs where the expected remaining service life is within that of the patching material.

Refractory *F* showed considerable spalling in various locations and was removed at about the same time for each bottom life.

Refractory *G* appeared to give satisfactory service; however, each time a bottom was removed there was iron under it. This led to the opinion that the operation was being carried out near the danger point.

It is interesting to note that refractories *A*, *B*, *C*, and *F*, were not patched, and when removed, showed indications that they still had additional life left. The service lives listed in the table do not include any additional service after patching.

Conclusions

It seems possible that the laboratory tests may be used to predict the action of a refractory in service. The impurity content of the refractory appears to have a strong effect on its general properties.

In laying up a bottom, care should be exercised to see that the heels of the bricks in each ring are as tight as the tops. The best brick available will fail prematurely if it is improperly set. Each brick is dipped to cover all contacting surfaces with an even layer of a sillimanite mortar cream. The completed new bottom is given a thin sillimanite mortar wash to fill any existing joint cracks. Sand is never swept over the bottom.

Standard 9-in. brick are preferred because they are more easily obtained, and they form a smoother contour for the bottom. The few extra joints as compared to the large brick, do not seem to affect the service life of the bottom.

High alumina refractories appear to give longer, more dependable, service life than those of lower alumina content.

Impurities, such as iron oxide and alkalies, may tend to shorten the expected life of a given alumina content. This is especially noticable in the linings of cupola front slaggers. In service, 60 per cent alumina brick of high purity, had as good bottom performance as 80 per cent alumina brick of lower purity.

The present practice is to use either the 60 per cent alumina brick of high purity, or the 80 per cent alumina brick of lower purity, because it is felt that they are more dependable than the other brick listed in the table.

DISCUSSION

Chairman: W. R. JAESCHKE, Whiting Corp., Harvey, Ill.

Co-Chairman: J. P. HOLT, Basic Refractories, Inc., Clayton, Mo.

CHAIRMAN JAESCHKE *(Written Discussion):* The authors are to be highly complimented on this fine study of their refractory problems and the excellent manner in which it has been reported.

It is indeed gratifying to see how closely their results tally with conclusions that could be drawn from a close study of all the properties of these refractories. The exception to this lies in the difficulties encountered with spalling of some of the higher grades of brick. Inasmuch as spalling generally occurs at the line of vitrification in the brick, and that line in the case of higher grade brick is much closer to the hot end, I wonder if some benefits might not be obtained if a damper were used at the bridge wall after a heat to avoid too rapid cooling. Such dampers have been used in several different furnace installations and have appreciably retarded the cooling rate of the furnace.

The authors mention that several super-duty brick tested, failed due to joint penetration and the brick were still full size. It has been my experience that when such brick failed prematurely, close examination disclosed that appreciable shrinkage had taken place in the brick and that only the lower end was still full size. Because the lower ends were not laid tightly, the furnace bottom failed early. The data in this paper show that the lower end of the brick reach 2090 F at the end of 4 days' 2-shift operation. Considering that this class of brick softens at 2325 F, a thermal gradient exists in that brick shows only about 2¼ in. of the lower end of the brick below the softening point. There is no wonder that such a brick will fail early if this lower end is not laid tightly.

Again, I wish to thank the authors for this excellent contribution to our studies of refractories in service.

CENTRIFUGAL CASTING OF UNUSUAL SHAPES IN NON-FERROUS ALLOYS

By

J. P. Krishon*

The permanent mold foundryman quoting on a new job often finds the most difficult problem is that of establishing a manufacturing technique to fit the quality and maintain a competitive price. Normally centrifugal process is completely ignored because of:

1. Probable high tool cost considerations,
2. Technical aspects of a relatively unknown process.

Actually the centrifugal process is basically simple. The foundryman who ignores its possibilities is overlooking cost cutting potentials that can be large for some types of work.

An investigation of textbook and handbook data on the centrifugal process will not reveal its entire potential. Some work done in Germany during the last war and published in the FIAT reports in 1945 was concerned with centrifugal casting of pure aluminum windings in electrical motor rotors. This process has been improved upon in this country. However, until recently most centrifugal work was limited to bushings in brasses, bronzes and ferrous alloys, tubing, soil pipe and babbit liners in medium and large bronze bearings.

An example of an aluminum permanent mold casting successfully made by the centrifugal method is the adapter plate (Fig. 1) being produced for one of the major automotive companies. A general purpose aluminum alloy with the following composition is used:

Copper	3.0-4.5	Zinc	1.0 max.
Iron	1.0 max.	Nickel	0.5 max.
Silicon	6.5-7.0	Titanium	0.2 max.
Manganese	0.8 max.	Others	0.5 max.
Magnesium	0.5 Max.		

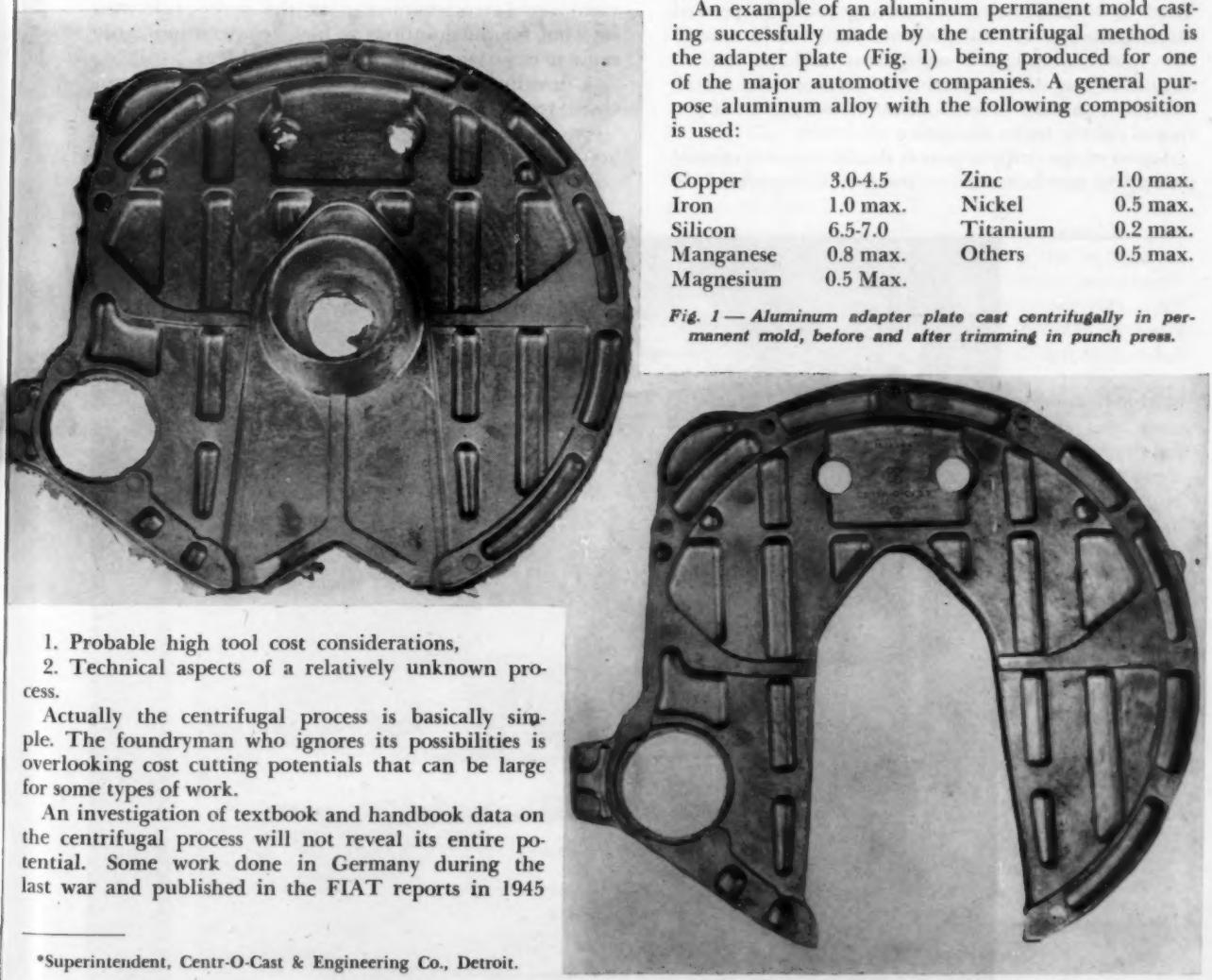


Fig. 1 — Aluminum adapter plate cast centrifugally in permanent mold, before and after trimming in punch press.

*Superintendent, Centr-O-Cast & Engineering Co., Detroit.



Fig. 2 — Cope (left) and drag of metal mold for aluminum adapter plate.

The casting is poured in a horizontal position in a mold consisting of a top and bottom cavity (Fig. 2) machined from a hot worked No. 3 pressure die steel. There is no doubt that the casting could be poured successfully in a vertical position. This would necessitate more complicated equipment and, as experimentation has shown, would not improve the metallurgical soundness or unit manufacturing cost of the type of casting under discussion.

Action of the stripper pins is similar to conventional permanent molds and their position is important. In general, the pins should be located at the outer periphery of the casting (Fig. 1) with consideration being given to deep pockets and cored out areas. Elevated mold temperatures near the sprue rule this area out for pin locations as high temperatures cause rapid deterioration of pins and pin bushings.

A description of the mechanical details of the machine required to operate this mold is beyond the scope of this article. Basic features of the machine are:

1. A horizontal centrifugal table (Fig. 3) on which

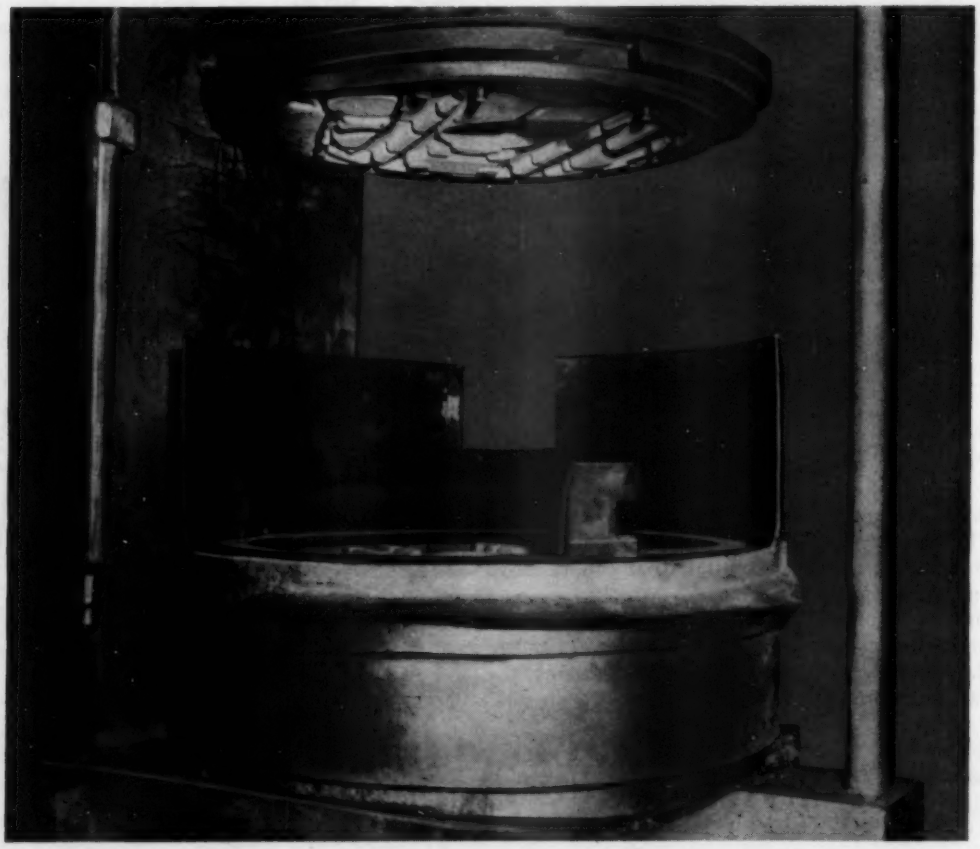


Fig. 3 — Centrifugal molding machine with cope half of mold raised.

the bottom cavity is affixed by bolts or clamps. The table may be operated either by hydraulic or electric motor. Hydraulic is preferable because of the range of speeds available.

2. A superstructure with a hydraulic cylinder lift of 18 in. or greater for the purpose of opening and closing the molds. The superstructure must contain bumper plates or posts to activate the stripper pins.

3. A hood for the spinning mold.

In the manufacture of an unusual part, the centrifugal speeds used for manufacture of bushings and sleeves do not apply. Figure 4 shows a cross section of the adapter plate cut directly through the sprue. The sharp changes in direction undergone by the molten metal as it travels through the mold, in addition to the variations in wall thickness, create some special problems.

Relatively heavy sections are isolated from the sprue by thin sections. Normally, in static permanent molding, chilling of these heavy sections would minimize shrinkage; however, experimentation showed that the unusual conditions encountered when casting the part centrifugally made chilling ineffective.

The mold was initially prepared for operation by preheating with gas torches, and applying a conventional China clay spout about 0.008 in. thick.

Experimentation revealed that three factors required control which were so interrelated that they must be discussed as one. These are: metal temperature, mold temperature, and operating cycle.

Initially, a speed of 250 rpm and a metal temperature of 1350 F were selected. Metal was poured into the mold in a very thin stream. Under these conditions, an operating cycle of four to five minutes was required to allow the sprue sufficient time to solidify. Several hours of operation showed that the mold would retain an operating temperature of approximately 700 to 750 F. Fractures and X-ray examination of castings revealed extensive shrinkage in the heavier sections.

To eliminate the shrinkage, the thickness of spray was increased on thin sections and removed from the thick sections to provide chill. Investigation indicated that pouring metal in a thin stream in addition to altering spraying technique produced castings which were sounder in the areas affected by shrinkage; however, evidences of lamination occurred which were due to the slow pour.

The metal temperature was therefore dropped to 1300 F and the pouring rate increased. This brought about a slight improvement in the shrinkage condition and the mold was allowed to operate for two weeks.

Frequent checking with a surface pyrometer indicated mold surface temperatures in excess of 750 F. Signs of erosion of the mold cavity appeared and severe galling of stripper pins and stripper pin bushings was in evidence. The mold was disassembled and the eroded areas welded and hand-ground to size and contour as determined by use of a templet.

The mold was prepared again with 0.008 in. spray and the metal temperature was reduced from 1300 to 1250 F to reduce the mold temperature. Under these conditions, shrinkage was still evident in the heavy sections and chilling made little improvement. Therefore, speed was reduced to 160 rpm and the sprue size increased to make possible rapid pouring.

The lower pouring temperature permitted shortening the operating cycle from approximately four minutes to three minutes. The areas of shrinkage showed marked improvement and the mold was put into operation for several weeks on a 24-hour per day basis. Periodic checks with the surface pyrometer revealed that mold temperatures reached 700 F. Signs of erosion appeared after the first week and by the middle of the second week, stripper pins again showed galling.

The mold was repaired again, resprayed and put into operation. The metal was now poured at 1200 F. On a 24-hour per day basis, at an operating cycle of three minutes, several weeks passed with no erosion evident. Stripper pins remained in good working condition and mold temperatures remained under 700 F. The shrinkage previously evident in heavy sections disappeared. X-rays and sectioning of parts indicated sounder structure than that obtainable by gravity methods.

It was then felt that the process was ready for production and with some modification the equipment was rigged to permit one man to operate two molds. On two molds, operating three shifts per day, each operator was able to produce 300 castings or more per 8-hour day. The operator is provided with relief time each hour.

Using the final technique developed, six molds in operation three shifts per day, five days per week, have run more than 420,000 pieces with no erosion evident in the mold cavity and only occasional down time to replace stripper pins or stripper pin bushings. For the past 12 months scrap has been less than 6 per cent. Molds are taken down once each week and the cavities sand blasted and re-sprayed.

Colloidal graphite spray was found to be beneficial in aiding ejection of the casting and extending the life of the China clay spray. In addition, the mold is blown off with an atomized water spray after each casting is ejected. This has the dual effect of clearing metal flash from the parting lines in addition to assisting in keeping mold temperatures within the desired range.



Fig. 4 — Cross section of adapter plate showing sprue in center and variations in thickness of the casting.

The thin gates and minimum amount of flash lead to the use of a trim die rather than band-sawing and grinding. In a conventional crank-type press equipped with an air clutch drive, one operator can trim 1400-1600 castings in an 8-hour day. No subsequent grinding is necessary. The casting, including sprue and flash, weighs 9.0 lb while the finished casting weighs 7.3 lb. This high yield of 81 per cent, together with low scrap, fast operating cycle, and ease of trim has rendered the process far more economical than conventional gravity methods for applicable castings.

The low pouring temperature minimizes the danger of gas porosity and lessens oxide formation due to agitation. However, a problem is created in fluxing and skimming practices because adequate drying of dross skimmings is almost impossible at this low temperature. As a result, skimmings will be found to contain a high percentage of good metal, and smelting operations for salvage may be necessary.

The cost of purchasing or creating equipment to produce castings centrifugally is higher than for conventional static permanent molding. But the expense is quickly amortized by reduced operating cost.

Without exception, sounder castings are obtained by this method than can be produced by gravity casting methods for the same type of work. In addition, for those parts converted from conventional to centrifugal methods, an average unit cost saving of more than 40 per cent has been realized. This is attributable largely to two-mold per man operation, high yield, and extensive savings in the cleaning room.

The foundryman who has no access to engineering facilities would be well advised to contact a reputable manufacturer of permanent mold equipment before embarking on a centrifugal casting program.

DISCUSSION

Chairman: B. L. MEREDITH, Federated Metals Div., American Smelting & Refining Co., South Plainfield, N.J.

Co-Chairman: W. BABINGTON, Bell Telephone Laboratory, Murray Hill, N.J.

W. W. EDENS[1] *(Written Discussion):* Mr. Krishon and his associates are to be complimented for their ingenious use of the centrifugal casting process to produce the casting described in this paper. At first glance it may seem like an obvious thing to do but if we will stop to consider what was actually done, this becomes a very interesting casting job.

First of all, this casting fits the definition of a true centrifugal casting in that it is spun on its own axis during casting. However, it also has some of the characteristics of a centrifuge casting. A centrifuge casting by definition is one which is positioned in a mold away from the central sprue and which uses a low order of centrifugal force to provide feeding. The subject casting is rotated very nearly around its own axis but at far less speed than is normally used in true centrifugal casting practice. A low order of centrifugal force is used as provided by a low rotational speed, to accomplish proper feeding. The resulting casting is indeed a hybrid in centrifugal casting techniques.

The mold used to produce the castings followed general permanent mold practice and design. In order to give the part more symmetry and to provide a central sprue, a web was provided which filled in an open area in the casting. This web carries a heavier gate section which helps the distribution of metal in the mold, primarily to the somewhat heavier rim section. This helps to minimize turbulence which if it were present could result in air entrainment, oxide inclusions, and drossy surfaces.

Systematic variation of centrifugal casting conditions as to pouring temperature, mold temperature, pouring rate, and speed of rotation has led the author to the general optimum conditions for centrifugal casting work. These conditions will vary with different casting shapes and casting volumes but optimum conditions usually include the following.

1. Mold temperatures as might be considered normal to the rate of repetitive pouring without adding or removing heat other than imparted by the solidifying metal.

2. Rotating speeds only sufficient to assure that metal reaches all corners of the mold, and conforms to the mold shape, with minimum turbulence.

3. Pouring temperatures as low as possible to prevent mold erosion and wear without incurring cold laps or seams in the casting.

4. Pouring rates as high as possible without too much turbulence. A high pouring rate materially helps to achieve low pouring temperatures.

This wedding of a fairly conventional type of metal mold, complete with knockout pins, and the centrifugal casting process suitably modified with centrifuge casting overtones, is a unique accomplishment. Again I would like to compliment this organization for having the imagination to conceive this combination, their obvious technical competence, and willingness to publish and share their know-how with others. This serves the highest interests of the foundry industry whose members must always realize that their main competition is not other foundries but rather other processes of fabrication.

1. Allis-Chalmers Mfg. Co., Milwaukee, Wis.

CERAMIC-MOLD PROCESS FOR STEEL CASTINGS

By

D. C. Ekey* and E. G. Vogel**

Quality and cost advantages of Lebanon Steel Foundry's ceramic-mold process (used under a licensing agreement) can be evaluated now after a year of production. We can also report on application of the process to the problems of design and materials engineers.

In the process,† a mixture of fine and coarse sillimanite, liquid hydrolyzed ethyl silicate, and a liquid catalyst are mixed in quick fashion and poured over a prepared pattern. The mix subsequently sets in a few minutes to form a gel and then a rigid mold. During the gel stage the mold is moderately flexible which facilitates stripping from the pattern.

Following stripping, the mold is flamed with a torch, assembled, fired at red heat in a furnace, and is then ready for pouring either while still hot or later when cool, as requirements dictate. The chemical set of the mixture involves two distinct steps: during the life of step one the mold retains a flexible quality, and as step two occurs a complete rigidity ensues.

Advantages

The process has significant advantages for producing steel castings of medium and small size. Thin sections down to $\frac{1}{16}$ in. can be cast. When we go after thin sections we can use molds as hot as 2000 F. Gating systems are extremely important in running thin sections. Higher metal temperatures are, of course, helpful. The difference in various alloy fluidities is to be considered. We have not as yet employed pressure and centrifugal systems.

Surface smoothness is quite evident. We are claiming 125 microinch finish and better. This quality depends on many things, among which are hydraulic head, fineness of refractory grain, metal and mold temperatures, section thickness, and metal fluidity.

Design complexity far exceeding conventional sand castings is possible. The material is relatively strong, smooth, and yet flexible, making possible release from patterns with no draft and producing a mold with faithful registration of fine detail.

Tolerances from ± 0.003 in./in. are possible where the measurements are made on parts cast wholly in the same section of the mold. An additional tolerance of 0.006 to 0.010 in. across the parting line must be accommodated.

The good tolerances and smooth casting surfaces obtained indicate substantially less machining than is necessary with conventional sand castings.

Low pattern cost is a definite factor when the number of pieces to be cast is less than 500. This figure is not the result of an accurate survey but rather a result of cursory study and would, of course, be affected by many factors.

Simple pattern construction and molding procedure insure short lead time. Patterns of bulky design may be made in wood and as such are suitable for short runs of 50 to 100 molds. The best functioning patterns are of mounted design, cast in aluminum and polished.

The ceramic molds permit the use of a great variety of steel alloys without difficulty. Occasionally we encounter the measle defect in the 12 per cent Cr steels.

The process is adaptable to a wide range of casting sizes up to weights of several hundred pounds. As the casting weight increases and design becomes more massive, the surface loses smoothness.

The internal metal sections are unusually clean and free of any gas and other inclusions. This is directly due to the extreme resistance of the mold material to spalling or erosion during the casting of molten steel.

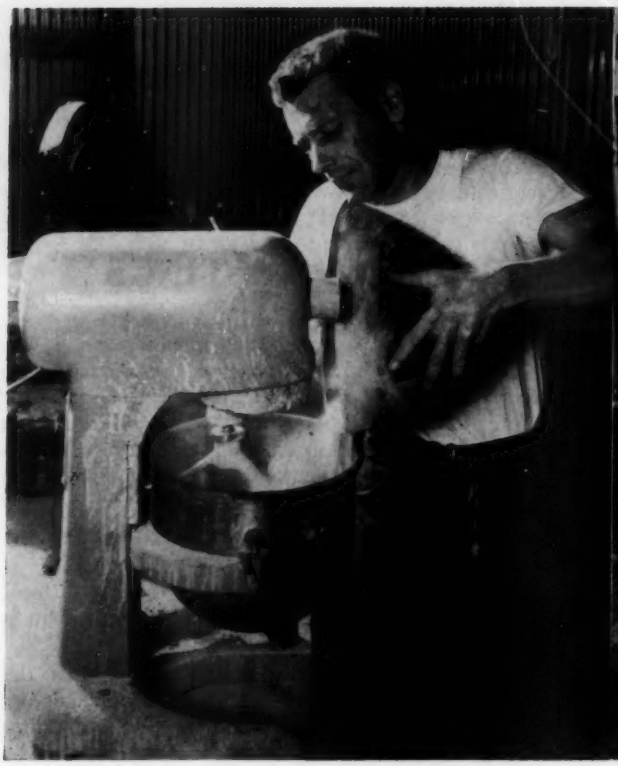
Disadvantages and Limitations

While the process has many significant advantages, it also has its own inherent limitations, and the design and materials engineer must carefully evaluate these limitations so that he may make the best possible use and application of the process.

Since the process invades the field of precision casting, a deciding factor in selection or rejection of

*Director of Research and **Research Engineer, Lebanon Steel Foundry, Lebanon, Pa.

†*"Investment Precision Casting Without Expendable Patterns,"* A. Dunlop, AMERICAN FOUNDRYMAN, June 1954, pages 64-69.



Preparing mold mixture; ceramic used is sillimanite.



Clean pattern gives smoothest casting, best results.

the process is the number of pieces to be cast from one design. When this quantity is approximately 500 or less, the short lead time and low pattern cost of the ceramic-mold process give it definite cost advantages. As quantity requirements exceed 500 or more, then the high first cost of competing precision casting processes is more easily justified. It should be pointed out that the process has very definite advantages in prototype work, where design changes can be made quickly and inexpensively.



Ceramic mixture pours freely into mold.

The accuracies possible with the ceramic-mold process must be qualified when compared with those produced by the lost wax method. The dimensional history of the ceramic-mold product at Lebanon is competitive with the lost pattern type in some cases and definitely noncompetitive in others. For large quantities of small intricate parts, the lost wax process would probably be selected. For components up to about two pounds in weight, either lost wax or ceramic-mold casting might be selected. In large sizes, however, the new process without expendable patterns has the advantage.

Casting Designs Adaptable to the Process

While it is conceivable that the ceramic-mold process is adaptable to almost any steel casting design, there are certain designs which do not lend themselves ideally to the production concept. These include rangy flat components, relatively large print areas, fine or complex core placements, and deeply recessed mold pockets. Although it is possible to transcend all of these limitations in the process, the high labor cost involved because of extraordinary attention or supervision makes such production impractical in all but isolated cases.

Production and Materials Costs

The nature of the process does not lend itself to high speed production. A typical time analysis of the molding of a simple coreless (one to ten pound casting) mold is as follows:



Molds are torched before cleaning and baking dry.



Sand cope, ceramic drag are used for some castings.

	Man Minutes
Pattern cleaning	1.50
Mixing slurry	2.00
Pour and set	1.00
Pattern draw	1.50
Flaming	.75
Closing and luting	3.00
Fire	.50
	10.25

This indicates we can get about five to six molds per man hour.

In the category of materials, we are confronted with high costs. A listing of the major materials and their cost per pound is as follows:

Hydrolyzed ethyl silicate	$0.46
Catalyst	0.03
Sillimanite	0.075

Calculations show the following mold weight and volume costs:

Dry mold cost/lb	$0.19
Dry mold cost/cu in.	0.012

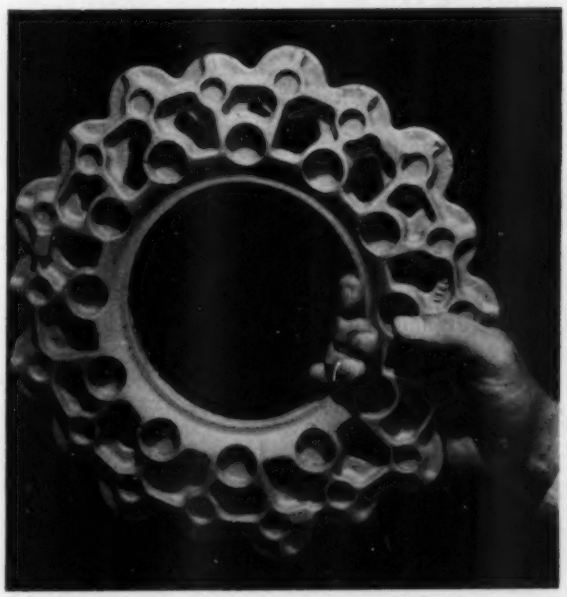
A comparison of the ceramic-mold process with green sand, investment casting, and shell mold casting is shown in the table.

Applications

For service in the aircraft field, jet engine manifolds and blades have been produced from highly alloyed materials while bearing supports and elbows have been cast from low alloy steels. Other aircraft components include landing gear struts, shock absorber housings, brake backing plates and certain structural elements where it is possible to use ferrous alloys.

Engine applications also include blades and vanes for gas turbine vehicular propulsion, steam turbine guide vanes, and diesel engine heads, the latter produced from heat resisting alloys.

A variety of industrial applications includes stainless steel pump impellers and filter press frames produced for service in the process industries, bottling



Typical casting made by the ceramic-mold process.

CERAMIC-MOLD PROCESS COMPARED WITH GREEN SAND, INVESTMENT CASTING, AND SHELL MOLD CASTING:

Casting Requirements	Ceramic-Mold Castings	Investment Castings	Green Sand Castings	Shell Mold Castings
Surface smoothness	80-125 micro-in.	40-125 micro-in.	200-500 micro-in.	80-165 mircro-in.
Intricacy	Excellent, approaching but not equalling precision castings	Excellent	Fair	Excellent but not equal to Ceramic-Mold
Thinness of metal sections	Excellent	Excellent	Fair	Good
Tolerances	Good to excellent	Excellent	Fair	Good
Machining costs	Machining greatly reduced, sometimes but not always eliminated	Minimum machining required	Relatively high cost	Slightly higher than Ceramic-Mold
Lead time	Very short lead time	Longest lead time	Shortest lead time	Long lead time
Adaptability to various sizes	Casting size not restricted except above 100 lb for the top size of the casting	Restricted to small castings	No size limitations	Comparable to Ceramic-Mold
Adaptability to various metals and alloys	No limitations	No limitations	Almost no limitations	Almost no limitations except low carbon steels
Pattern costs	Very low cost; jobbing wood or metal patterns may be used	Very high cost	Very low cost	Very high cost but less than investment castings
Prototype adaptability	Low cost	High cost	Not applicable	Generally, not applicable

This table is compiled for comparison only and is not to be considered absolute. For exact data, it is necessary to consult the foundry engineer on the particular design involved.

cap guides, and drop forging dies. Among large parts which have been produced are chromium steel glass molds for 21-in. television tubes. These weigh as much as 300 lb.

Extrusion die inserts of a cobalt base alloy have been highly successful in the extrusion of copper, the highly alloyed inserts have served for the extrusion of more than 2000 billets, compared with a previous limit of about 250 billets with stainless steel. Dies of chromium steel produced for plastics injection molding have also given excellent results.

Summary

The ceramic-mold process has a very definite place in steel casting design and production. As a process it offers quality advantages in the end product substantially exceeding those of conventional sand castings but generally less than those offered by precision methods. The process is adaptable to a wide range of applications requiring close tolerances and dimensional accuracies at lower costs than precision methods where quantities do not exceed approximately 500 pieces per design.

The process may, be considered an important supplement to existing methods. It adds much needed flexjbility to casting design and production by offering a quality range between the limitations of conventional sand castings and the unique accuracies of precision castings. It provides many advantages of the latter at lower cost.

USE OF GLASS SPHERES FOR CALIBRATING SAND TESTING SIEVES

By

R. E. Morey* and J. H. Schaum**

Accuracy of testing sieves is a serious consideration to users, particularly when the decision to accept or reject a shipment of sand is based on sieve analysis of particle size distribution. According to the sample and sieves used by the vendor the material may be within the agreed specifications whereas the buyer, using a different sample and set of sieves, may find the material outside the specifications and wish to reject the shipment. The question naturally arises as to whether the samples are comparable and representative of the size distribution of the shipment as a whole, and whether the nests of sieves used by buyer and seller actually have openings of the size stated on their name plates.

It is usually easier to get a representative sample of sand (or any particulate material) from the production or manufacturing end than from the users end because, in shipment, vibration causes the smaller particles to settle to the bottom and the larger particles remain on top. Sampling may be done at some step in the production process such as where a conveyor discharges into a car by taking the full flow of the conveyor for several very short intervals of time.

This produces a gross sample of 100 to 500 lb. This sample may be split in a sample splitter to produce several representative samples of 5 to 10 lb for strength tests and samples of 50 to 100 grams for sieve analysis. These samples may be furnished by the producer, or, if this is not satisfactory to the user, they may hire an independent party to take the samples and supply what might be called an official sample to both producer and user. In this particular study, the variables in sampling procedures were not investigated.

The problem of determining whether the actual opening of the sieves is the same as stated on the nameplate is, unfortunately, not quite so simple. In general, sieve manufacturers do an excellent job in making sieves to close tolerances. Sieves which are faulty in manufacture sometimes escape detection and when the sieve is used it is subject to abrasive wear and mechanical damage. It is, therefore, not surprising that sieve analyses made by different companies do not always agree perfectly.

This problem is not peculiar to the foundry industry but is shared by all who use sieve analyses for close control of their manufacturing processes. The sugar refining industry, for instance, uses granular bone char as an absorbent for removing the impurities from sugar and the size and distribution of the particles are important to the process. Several interests including the United States Cane Sugar Refiners and Bone Char Manufacturers, a greater part of the refining industry of the British Commonwealth and Belgium, and the National Bureau of Standards, jointly sponsored a research project which has resulted in the glass sphere method for the evaluation of the effective opening of testing sieves. Most of the work on this project has been done by F. G. Carpenter and V. R. Deitz (1), (2).

To evaluate the glass sphere method of sieve calibration in the foundry industry, the Joint AFS-NISA Grading, Fineness and Distribution Committee has undertaken a cooperative project in which the following companies participated: American Colloid Co., Aberdeen, Miss. and Belle Fourche, S. Dakota, American Manganese Steel Div., Chicago Heights, Ill., The Ayers Mineral Co., Zanesville, Ohio, Carpenter Brothers, Inc., Milwaukee, Wis., Harry W. Dietert Co., Detroit, Mich., Great Lakes Foundry Sand Co., Detroit, Mich., Hill & Griffith Co., Cincinnati, Ohio, Industrial Silica Corp., Youngstown, Ohio, Manley Sand Co., Rockton, Ill., New Jersey Pulverizing Co., New York, N. Y., New Jersey Silica Sand Co., Millville, N. J., The Nugent Sand Co., Muskegon, Mich., Ottawa Silica Co., Ottawa, Ill., George F. Pettinos Inc., Philadelphia, Pa., Sand Products Corp., Cleveland, Ohio, Standard Sand Co.,

*Metallurgist, Naval Research Laboratory, and **Metallurgist, National Bureau of Standards, Washington, D. C.

NOTE: This is a Committee Report sponsored by the Joint AFS-NISA Grading, Fineness and Distribution Committee (8-F).

(1) "Methods of Sieve Analysis With Particular Reference to Bone Char," F. G. Carpenter and V. R. Deitz, *Journal of Research of the National Bureau of Standards*, Vol. 45, no. 4, Oct. 1950.

(2) "Directions for Using Glass Spheres for the Evaluation of the Effective Opening of Testing Sieves," Bone Char Research Project, 333 Medford St., Charleston 29, Mass.

September 22, 1954

To: Committee on Grading and Fineness and Companies Participating in Study of Reproducibility of Sieve Analyses

From: J. H. Schaum and R. E. Morey

Subject: Calibration of Sieves.

Sixteen companies including sand producers, foundry suppliers and foundries have ordered samples of glass beads for the calibration of their sieves and agreed to furnish certain data to the Committee on Grading and Fineness to check the effectiveness of calibrated sieves in giving reproducible sieve analyses. The samples of glass spheres and instructions have been forwarded to each participating company. Also, under separate cover, a special test sample of sand has been distributed.

Important. Do not open the sample of glass spheres until the directions for their use have been read and are thoroughly understood. They are a precision measuring tool and their size distribution is known accurately. Loss of spheres or contamination of the sample destroys its accuracy.

For the purpose of the Committee it is desired that the participants select a set of sieves in good condition and clean it thoroughly. The glass beads are then to be sieved thru the nest of sieves and the weights recorded, preferably as cumulative weighings. Each sieve should be cleaned thoroughly after the glass beads have been sieved so that all possible beads are returned to the sample. The special test sample of sand is then to be sieved on the same set of sieves for the same time and on the same sieve shaking machine. Data are to be reported on the enclosed sheet. Each sieve in the set should have an identifying mark placed on it. Report the two sieve analyses, along with the make of sieving machine and the sieving time used to J. H. Schaum, National Bureau of Standards, Washington 25, D. C.

If the tests are made and reported promptly, they will be analyzed and submitted as a part of the Committee report at the next AFS Convention.

Should a company desire to test more than one set of sieves, it is requested that they do not submit data for more than three sets.

Figure 1

Grand Haven, Mich., and Whitehead Bros. Co., New York, N. Y.

Each company purchased a set of calibrated glass spheres from the Bone Char Research Project Inc. A set of identical sand samples were also prepared and distributed to all cooperating companies along with the data and instruction sheet shown in Figs. 1 and 2. Each sand laboratory then made a sieve analysis of the glass sphere sample and the sand sample, using the same set of their own sieves for each analysis. The results were reported on the data sheet and returned for tabulation. Table 1 shows the data for one sieve set from one company. The data were plotted (Fig. 3) and a corrected sieve analysis was obtained from the plot. Tables 2, 3 and 4 contain the summarized data from all companies.

Table 2 is a summary of the sieve analyses of the duplicate sand samples distributed to all the cooperating companies. In this table the data are based on the nominal sieve openings and the weights of sand retained on each screen were converted to cumulative per cent retained. Selected examples were taken from the data covering all sieves for all companies so as to

Grading, Fineness and Distribution Committee Data Sheet for Reproductibility of Sieve Analysis

Participating Company:_____
Address:_____
Manufacturer of Sieves:_____
Are Sieves Bureau of Standards Certified?_____
Type of Sieving Machine:_____
Sieving Time_____
Name of Person Conducting Test:_____
Date of Test:_____
Glass Sphere Sample Number:_____
Original Weight of Glass Sphere Sample:_____

Sieve No.	Nominal Opening in Microns (from side name plate)	Glass Spheres Cumulative Weight Retained on Sieve	Sand Sample Cumulative Weight Retained on Sieve
6			
12			
20			
30			
40			
50			
70			
100			
140			
200			
270			
Pan			
Other Sieves			

Figure 2

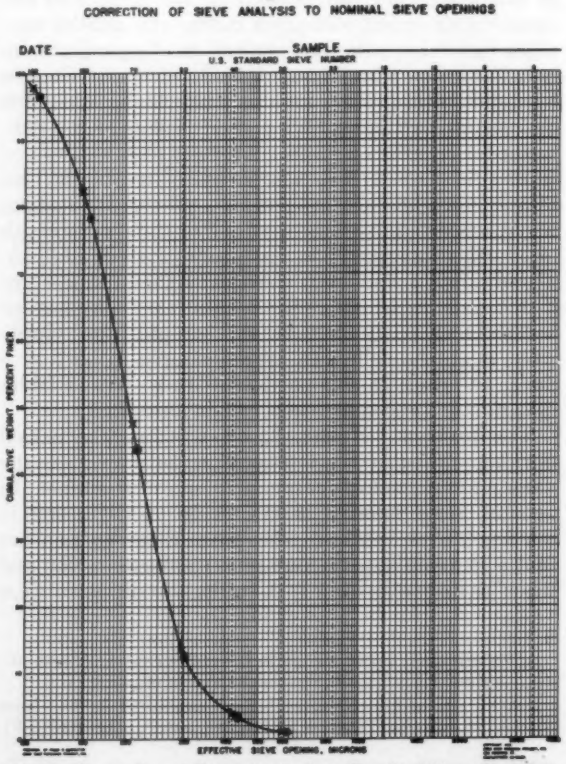


Fig. 3—Correction of Sieve Analysis to Nominal Sieve Openings

Table 1—Example of Data Sheet for One Company Used in Determining Cumulative Curve Shown in Fig. 1

Sieve No.	Nominal Opening Microns	Glass Spheres Cumulative Wght. Retained on Sieve Grams	% Finer than Sieve	Calibrated Screen Opening Microns	Sand Sample Cumulative Wght. Retained on Sieve Grams	% Retained on Sieve	Corrected Cumulative % Retained on Sieve
30	590	31.9	70.0	610	0.8	1.0	1.2
40	420	47.3	55.4	436	2.6	3.3	3.7
50	297	64.7	39.0	300	10.3	12.9	13.3
70	210	75.6	28.8	216	35.0	43.9	47.7
100	149	93.4	12.0	158	62.3	78.2	82.2
140	105	102.1	3.8	108	77.3	97.0	97.9

Table 2—Summary of Sieve Analyses of Sand Sample Based on Nominal Sieve Openings

Sieve No.	Max. Ret. Cum. %	Min. Ret. Cum. %	Range %	Avg. Ret. Cum. %	Std. Dev. %
30	3.8	1.0	2.8	1.3	0.7
40	4.3	3.3	1.0	3.6	0.9
50	14.9	12.0	2.9	13.3	0.7
70	51.4	40.9	10.5	46.0	2.8
100	84.9	78.2	6.7	82.3	1.8
140	98.4	94.1	4.3	96.8	1.2

Table 4—Summary of Sieve Analyses of Sand Sample Based on Calibrated Sieve Opening

Sieve No.	Max. Ret. Cum. %	Min. Ret. Cum. %	Range %	Avg. Ret. Cum. %	Std. Dev. %
30	1.9	1.1	0.8	1.4	0.3
40	4.7	3.3	1.4	4.0	0.4
50	15.7	12.0	3.7	13.8	1.1
70	50.5	42.5	8.0	46.4	2.2
100	83.2	80.0	3.2	82.0	0.9
140	98.6	96.0	2.6	9.75	0.7

Table 3—Summary of Data on Sieve Sizes as Determined by Glass Sphere Method

Sieve No.	Nominal Opening U.S. Series, Microns	Nominal Opening Tyler Series, Microns	Maximum Calibrated Opening, Microns	Minimum Calibrated Opening, Microns	Range Max. to Min., Microns	Average Calibrated Opening, Microns	Standard Deviation, Microns
30	590	589	629	595	34	612	10
40	420	417	445	416	29	429	8
50	297	295	311	291	20	300	6
70	210	210	216	203	13	211	4
100	149	147	158	142	16	148	4
140	105	104	122	103	19	111	8

demonstrate the extreme of maximum and minimum results obtained.

Columns 2 and 3 (Table 2) contain these maximum and minimum values for cumulative per cent sand returned on each screen and the range of these values is indicated in column 4. The average of all cumulative per cent values for each screen are listed in column 5 and the Standard Deviation from the average may be found in column 6.

The range values are an indication of the extent of disagreement in sieve analysis arising between the various companies represented in these tests. The greatest difference between maximum and minimum cumulative per cent values occurred at the level of the No. 70 sieve where 10.5 per cent existed in values reported.

Table 3 is a summary of the data obtained in determining the effective size of sieve openings by running a sieve analysis with a standard sample of glass spheres. In each instance the glass sphere sample was run through the same set of sieves through which the sand sample had been sieved. In the computations only the sieves ranging from No. 30 to No. 140 in the standard AFS series were used, since the sand sample and glass spheres were contained, for all practical purposes of accuracy, in this range.

Since both the U. S. Series and Tyler Series of sieves were used, the nominal opening is shown for each. It will be noted that the nominal opening is not the same for the two series in the case of the No. 30, 40, 50 and 140 sieves and this influence on the results can be noticed. The effective size of sieve opening was calibrated by sieving the standard sample of glass spheres, recording the weight retained on each sieve, and reading the effective size of the sieve openings on the calibration curve. The calibration curve has been established by previous investigators and accompanies each standard glass sphere sample.

From a table containing all the calibrated values for sieve openings as determined from data submitted by all cooperating companies, Table 3 was compiled. By comparing the maximum calibrated sieve opening with the minimum, one can appreciate the extent of variation that existed between 19 supposedly duplicate sieves for each sieve number. For example, in the case of the No. 30 sieves, there was a 34 micron range

between the sieve with the maximum calibrated opening—629 and the one with the minimum—595.

It is obvious that two companies comparing sieve analyses of duplicate sand samples will disagree by an amount proportional to the difference in effective size of openings in their respective screens. This difference is more serious if it involves the screens in the steep part of the cumulative curve where a small difference in screen opening is magnified into a large difference in cumulative weight per cent of sand retained. For example, a difference of 13 microns between two extremes of the No. 70 sieves resulted in a difference of 10.5 per cent (see Table 2) in the cumulative weight of sand retained on these two sieves. However, where the curve is relatively flat, as in the case of the No. 40 sieve, a difference of 29 microns brought about only 1 per cent disagreement in the sieve analyses of the sand sample.

The arithmetical average of all sieves for each size is shown in column 7 (Table 3). It will be noted that in all but one instance the average calibrated size of the sieves is larger than the nominal—a logical trend since the sieving of abrasive particles tends to wear the wire mesh and thereby enlarge the sieve openings. Glass spheres are a useful tool for observing this increase in effective sieve opening resulting from continused use. Any sieve damage would become immediately apparent when calibrated with glass spheres since the effective opening would exhibit a marked increase. By determining the actual effective sieve size one can adjust all sieve analyses to a common base of a theoretically perfect set of sieves.

The Standard Deviation for each sieve size is shown in column 8. These values show the micron range of sieve openings on either side of the average as determined by mathematical analysis. Approximately 67 per cent of all sieves used in these experiments would be included in the size range covered by this deviation. To include 95 per cent of all sieves, the range of deviation would have to be doubled.

Table 4 is exactly like Table 2 except that the data are based on *corrected* sieve analyses where Table 2 was based on *uncorrected* sieve analyses. The average range for the uncorrected data is 4.03 per cent. For the corrected data it is 2.37 per cent, which is a significant decrease in the spread of the results. The average of the standard deviations for the uncorrected data is 1.16 per cent while for the corrected data it is 0.8 per cent, again a decrease in the probable spread of the results.

Summary

It was hoped that the participating companies would use at least a few well worn sieves so that the usefulness of the glass sphere method could be given an effective demonstration. It appears that each company, not wanting to appear careless in their testing, used their most accurate sets of sieves for these samples and the sieve errors found were generally very small.

There is a general shift of calibrated openings to sizes larger than the nominal opening, indicating sieve wear.

Any set of sieves can be measured and assigned calibrated opening sizes which will make it comparable in accuracy to a master set of sieves. Periodic recalibration will maintain a high level of accuracy.

Average range and average standard error were less for corrected data than for uncorrected data.

CONTROL OF MAKE-UP AIR IN INDUSTRIAL PLANTS

By

K. E. Robinson*

Whenever air is exhausted from a building, make-up air (or air supply as it is sometimes called) is necessary. Most engineers will agree to this, but their failure to realize the overall importance of makeup air usually results in its necessity being ignored.

Along with knowing how much to supply, one must also know where, and at what velocity it should be introduced, and how it should be controlled. Failure to properly provide makeup air can result in reduced production, personnel problems, illness and even complete plant shut down.

Makeup air is only one leg of an often ignored, but very troublesome, triangle. The other two factors are heating and exhaust ventilation. The need for heating has been accepted for years, and yet many so-called "modern" plants are limping along with outdated, "horse-and-buggy" methods and equipment.

Industrial exhaust systems, as they are considered today, are relatively new.

Comfort is usually considered to be the condition that regulates the heat loss of the body at such a rate that we are not conscious of heat or cold. It wasn't until about 1912 that we have any record of anybody seriously interested in what this condition should be. At that time a medical doctor was invited to talk to the American Society of Heating and Ventilating Engineers. He had stated that work should be done to find the proper *air temperature, air movement, humidity and radiation* for the comfort of man. It wasn't until the early 20's, however, that work was actually underway.

The original work was carried out by placing seminude people in rooms which were maintained at various saturated air temperature. Their reactions were recorded and it was determined at what saturated air temperature most people were comfortable. Therefore, when we speak of 68-degree E. T. (effective temperature) we are speaking of a combination of wet and dry bulb temperatures which has the same effect on the body as 68-degree saturated air.

Referring to Fig. 1, dry bulb temperature is plotted on the horizontal lines and wet bulb temperatures on the vertical. The diagonal lines running toward the top right-hand corner are for relative humidity. We may ignore these lines as they are derived from the wet and dry bulb temperatures that will be used. We are most interested in the E. T. lines which are curved diagonal lines running toward the lower right section of the chart.

If we have a reading of 70-degree dry bulb and 70-degree wet bulb, we will have a saturated condition and the two lines will intersect at the 70-degree E. T. line which comes downward at an angle. Any combination of wet and dry bulb temperatures which falls on this line results in an E. T. of 70-degrees and will be in the comfort zone.

Most people are comfortable at about 71-degrees in summer and 68-degrees in winter. These figures are for still air only. There are charts available that give

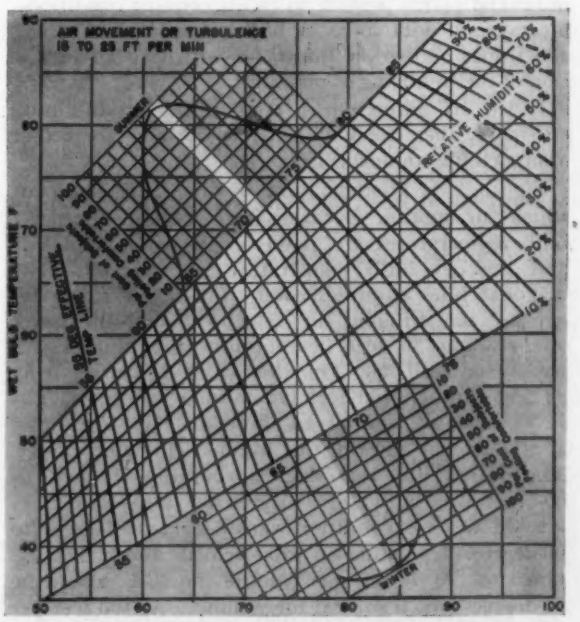


Fig. 1 — A.S.H.V.E. Comfort Chart for Still Air.

*Ventilation Engineer, Research Staff, General Motors Corp., Detroit.

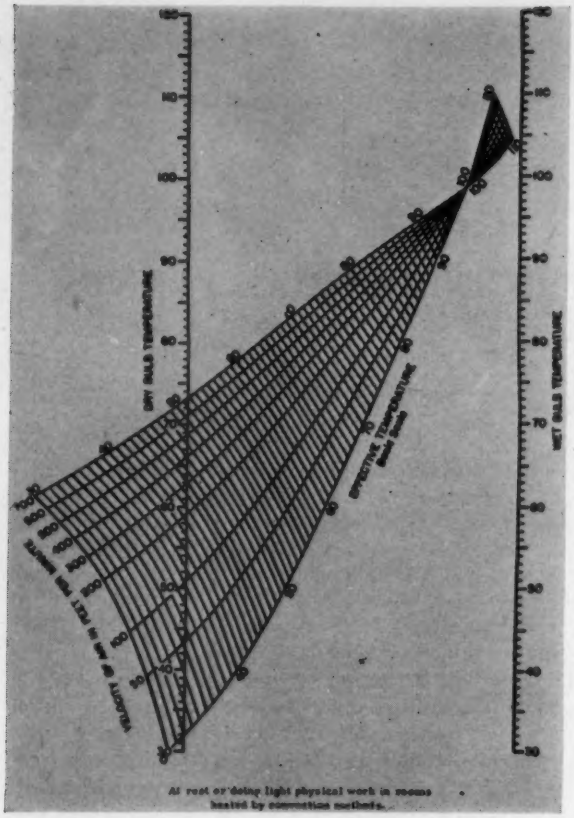


Fig. 2 — Another form of E. T. Chart.

a more detailed breakdown which include sex, age and geographical location.

It is not the author's intention to suggest that industrial plants provide so called comfort conditions, however, it does serve as a benchmark, and if we add upper and lower limits, we will have an indication of the wellbeing of our employees.

If we are interested in our people maintaining efficiency, then we will provide working areas that approach the comfort zone.

Figure 2 is another form of the E. T. chart in which air motion is taken into account. The wet bulb temperatures are given on the right-hand side and the dry bulb temperatures on the left. The diagonal lines running toward the upper right corner are for different air velocities, with the lower one reading 30 feet per minute. Anything less is considered still air. On this chart velocities from 30 feet through 700 feet per minute are given. The E. T. lines are similar to those on Fig. 1 running at a diagonal toward the lower right-hand corner. The dotted line represents a condition created by 65-degree wet bulb and 80-degree dry bulb. The E. T. created by this condition varies from 72-degrees to 60-degrees depending on what velocity the man is exposed to. At 40 feet per minute the E. T. would be 71-degrees which is ideal for summer. At 100 feet per minute the E. T. is reduced to 68-degrees which is most desirable during the winter months.

As stated before, this chart (Fig. 2) plots the effect of air movement up to 700 feet per minute. We are often asked what velocity is best for summer cooling and what is the maximum that can be used. Although it has been stated that there is no additional value of air moving at more than 800 feet per minute, this is not agreed to by workers in some hot industries. Air movement exceeding 3000 feet per minute is often insisted upon. In some plants, small high velocity jets are used so that the employees walk in and out of the air pattern. The claim is made that the cooling effect is more noticeable in this way.

In most of our discussions with plant personnel we have no difficulty explaining the effects of wet and dry bulb temperatures or air motion, but radiation and the strain it places on those exposed seemingly is not well understood.

There are several reasons for this misunderstanding. Radiant heat travels in a straight line like light and can be reflected or absorbed. It cannot be controlled by ventilation unless it is converted to some other form of heat. It cannot be measured by an ordinary thermometer without attachments.

The common method of measuring radiant heat is by the use of a thermometer encased in a thin wall black globe. An ordinary copper toilet float which has been painted will serve. This unit is placed where the reading is desired and left until equilibrium is reached. This will result when the loss by convection just balances that received by radiation. According to Bedford of England this reading can then be used in place of the dry bulb reading taken at the same location providing that the moisture content of the air is maintained.

The psycrometric chart (Fig. 3) will allow us to do that. The dry bulb temperatures are plotted across the bottom of the chart, and the wet bulb along the diagonal and project diagonally downward on the curved lines.

The lines running from lower left to upper right are for relative humidity and can be ignored.

The horizontal lines are for grains of moisture and must be considered if we are to use our globe

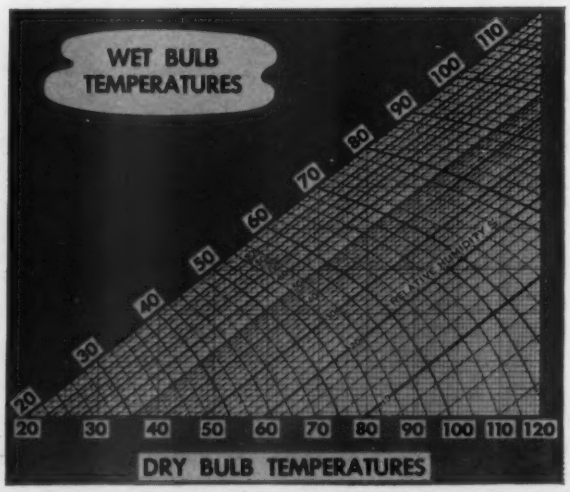


Fig. 3 — Psycrometric Chart.

thermometer readings.

In a plant where the 65-degree wet bulb and 80-degree dry bulb readings were taken that were shown on Fig. 2, a globe thermometer reading was found to be 100-degrees. To substitute this, we must follow the 80-degree dry bulb reading upward until we intersect the curved 65-degree wet bulb line. To maintain the same moisture content, we move to the right on the gr. of moisture line until we intersect the 100-degree dry bulb line, we then follow the curved line upward and to the left and read our new wet bulb as 71-degrees.

By substituting these new temperatures on the E. T. chart (Fig. 2) we can determine if we are in the comfort range. We find by the heavy line that the E. T. will be between 77-degrees and 81-degrees depending on the air movement. A few people might be comfortable at 77-degrees but 81-degrees E. T. is too warm for comfort even for a person at rest. This is not an exaggerated condition. For example in one plant globe temperatures were found to be 40-degrees and 50-degrees above dry bulb, and in another temperatures taken at the pouring station were more than 60-degrees above dry bulb temperatures.

In fact, the globe temperatures were still going up. Surely you have noticed the effect of radiation when standing in front of a fireplace or in front of a large picture window. The reverse of what has been shown here can also happen and the body will then lose heat to the cold surface.

When dry bulb temperatures drop much below 65-degrees accidents tend to increase. When temperatures are above 80-degrees, speed usually decreases and errors increase. When E. T. exceeds 90-degrees work can usually be carried on only one-third to one-half the time and the balance of the time is spent in resting.

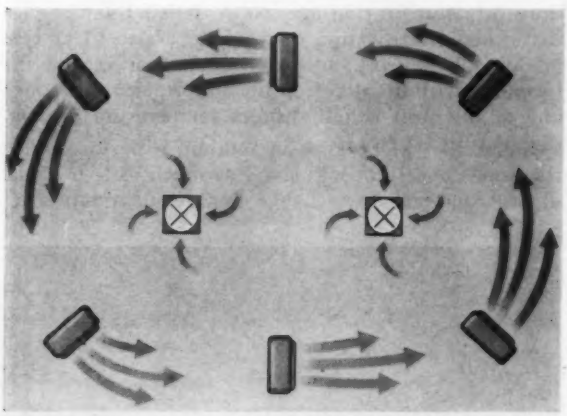


Fig. 4 — Plant layout showing location of heaters and flow of air.

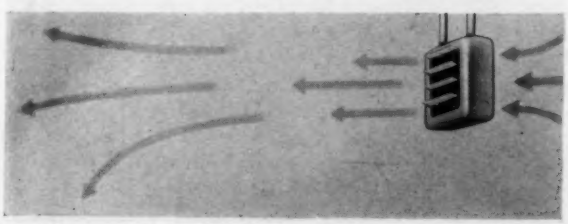


Figure 5

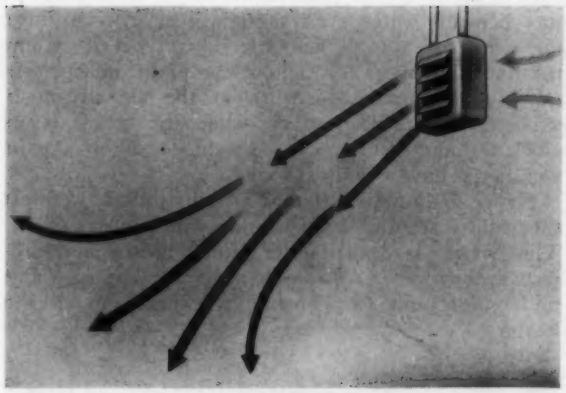


Figure 6

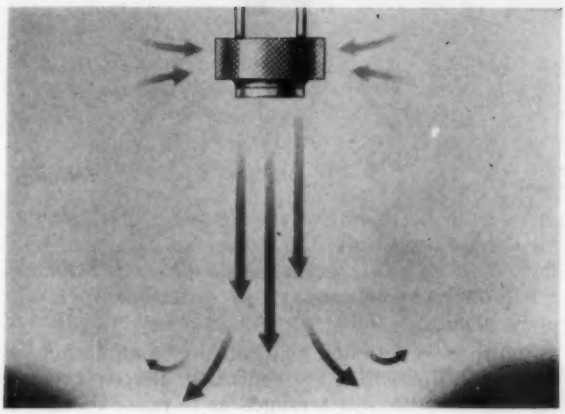


Figure 7

With this prelog of heating and comfort let us now see where makeup air fits into the picture.

In Fig. 4 we are looking down on the plan of a typical plant as laid out by many designers. The unit heaters are spaced with the thought that each one will project heated air forward to heat the wall between it and the next. Usually the heating and the exhaust fans are placed as though there was to be no manufacturing within the plant or no process heat or exhaust to be considered.

Figure 5 shows an elevation of a typical plant with the unit heater suspended near the ceiling above the building trusses. Ordinarily this type of unit is controlled by a thermostat which turns the fan on and off. If steam is used it is usually uncontrolled and if it is gas fired the burner comes on with the fan. The warmest air in the room is forced through the heater and discharged at an even greater temperature. Although this is a common type of installation, it is inefficient and does little to create a comfortable working zone. (Working zones are occupied by workers.)

Figure 6 represents a heater similar to the one shown in Fig. 5 but with the fins turned downward in an attempt to heat the living area. Due to the difference in density of the air this warm air will not stay down any more than a ping pong ball will

stay beneath water even though it is injected into the water at a high velocity. This type of system results in hot spots throughout the plant and anyone unfortunate enough to work in the air stream is exposed to hot blasts of air when the heater is on and resultant chilling when the heater is off. Again you do not have a comfort zone.

Figure 7 shows a different type of unit heater but the results are the same as those previously shown. The hot air is drawn in the sides, heated and projected downward. Again we have the hot spot but no comfort zone for anyone working beneath the heater.

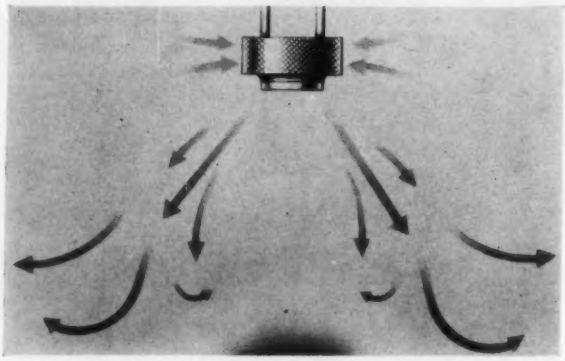


Figure 8

Figure 8 is another view of the same heater in which the fins which are concealed directly beneath the fan have been so arranged as to cause the air to be projected downward at an angle rather than straight down. If the air has sufficient velocity, it will affect more people, but will not provide comfort.

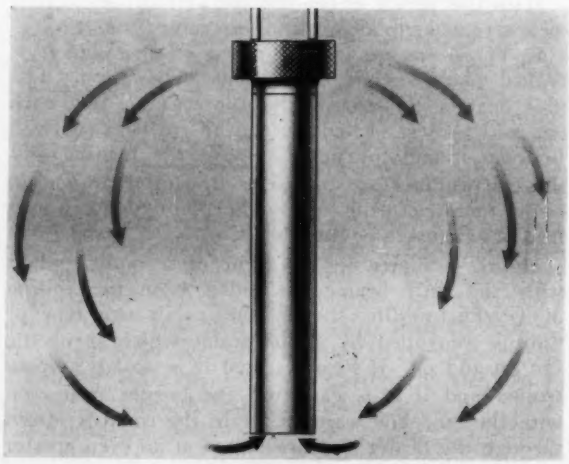


Figure 9

Figure 9 shows one method of improving such an installation. In this case the motor has been reversed and a duct added from the heater to within approximately one foot of the floor. Now the cold air is lifted from the floor, taken up and heated by passing through the heater which results in the warm air being drawn to the floor to fill the low pressure area caused by the removal of the cold air. Actual smoke tests prove that a heater arranged such as this will provide comfort in the working zone under many conditions. This is of value in small buildings where light work is being carried out and little or no heat is generated by the operation.

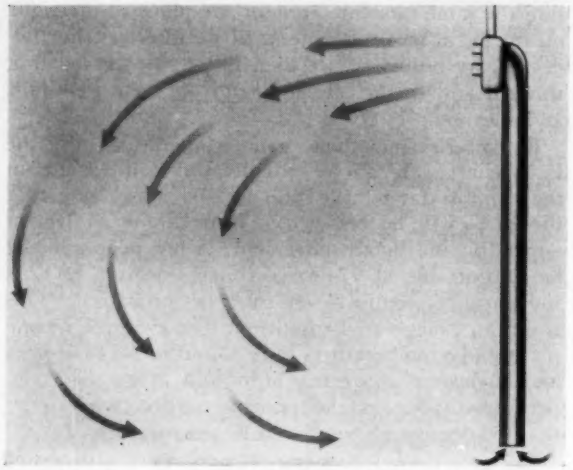


Figure 10

In Fig. 10 we have a heater such as the first one shown except that here too we have added a return air duct. Notice the fact that the louvers direct the air straight out in order that a greater area will be covered. We know that this air will return along the floor and will bring heated air down in to the working zone. Previously, the author mentioned the type of controls usually found on unit heaters. Of course, when the thermostat is satisfied and the fan shuts off, stratification may occur. It has been proven that when stratification does occur, a person at rest or doing light work, such as woman working on inspection, will usually feel uncomfortable with only one or one and a half degrees temperature change.

If the air can be kept in motion, a person under the same conditions will stand as much as five or six degrees temperature change before discomfort re-

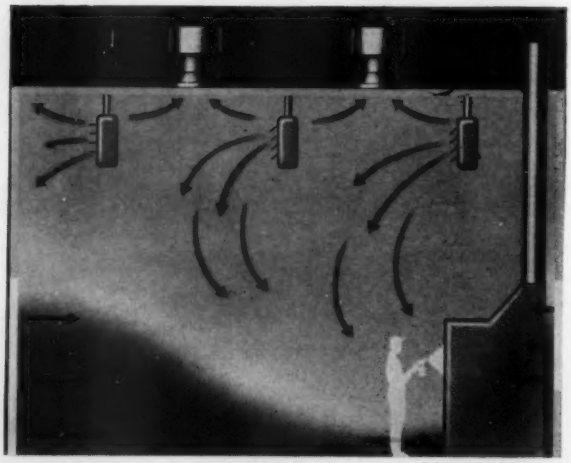


Figure 11

sults. Undoubtedly, many persons have had the same experience in their own home; although the thermostat operates on a one and a half to two degree differential, it is cool before the burner comes back on. In many cases this same problem in industry can be eliminated by varying the heat input to the unit heaters according to outside temperature and allowing the fan or blower to operate constantly during the working hours.

Figure 11 shows arrangement of the unit heaters and the ventilators in a typical plant. A comfort zone has not been provided. Exhausters are removing air from the building which results in the infiltration of cold air in the working zone. Even though we are talking primarily of winter operations, the exhaust ventilator on the right may cause difficulty in the summer months. If exhausts similar to these are placed over the heat treat area the combination of heat beneath the roof from the heat treat furnaces, the heat absorbed by the roof from the sun, and the discharge of hot air downward from the exhaust fans to the roof will result in temperatures above that which the roof will stand. Temperatures on a mild day exceeded 165 degrees on the roof and in order to eliminate the melting of the roofing material, baffles have had to be installed to prevent this hot air from striking the deck.

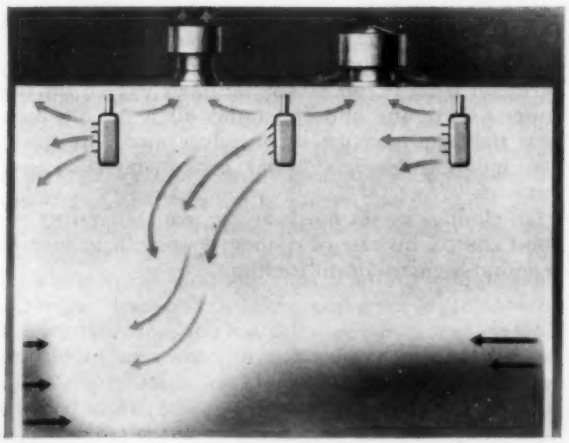


Figure 12

Figure 12 also shows a section through the plant. We have added a typical plant operation which requires the exhausting of additional quantities of air. Infiltration increases which results in a very uncomfortable condition around the edge of the building. It is also an added load on the heating system as this cool air mixing with the room air absorbs heat. By the time this cold air is heated, it has been drawn across the room and is exhausted by the ventilation required by the operation. The air bound condition caused by this high infiltration rate may also cause down draft through stacks and flues, carrying with it such toxic gases as carbon monoxide which are frequently present in the flues from direct fired heating units.

Figure 13 shows a plan view of this same plant showing not only the spray booth but two operations

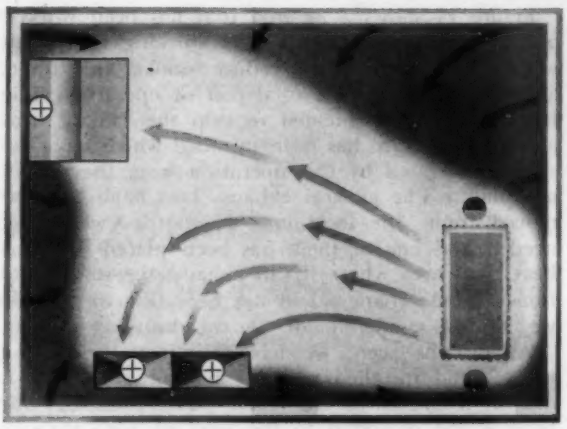


Figure 13

that are exhausted as well as a heat treat furnace. The two dots represent men working near this furnace; they are exposed to high temperatures on one side and cold drafts on the other side. We have checked similar conditions and found radiant temperatures ranging from 130 degrees to about 170 degrees. The thing that is most typical of all plants needing makeup air is the cold area completely around the exterior of the plant with the excessively warm area through the center. As previously mentioned, the cold area near the edge is brought about by the high infiltration rate. To overcome this condition the tendency is to continually increase the thermostat setting with the result that the central part of the plan receives excessive heat.

For all practical purposes, it is impossible to heat air which is drawn in through the doors and windows by ordinary heating systems. A number of plants have been studied and air velocities of 2,000 feet per minute were found entering shipping room doors. With doors having from 100 to 200 square feet that means 200 to 400 cubic feet of air per minute at perhaps zero degrees entering the working area. It would take 15 to 35 million Btu's per hour to temper this amount of air which shows that unit heaters near the doorway is not the answer.

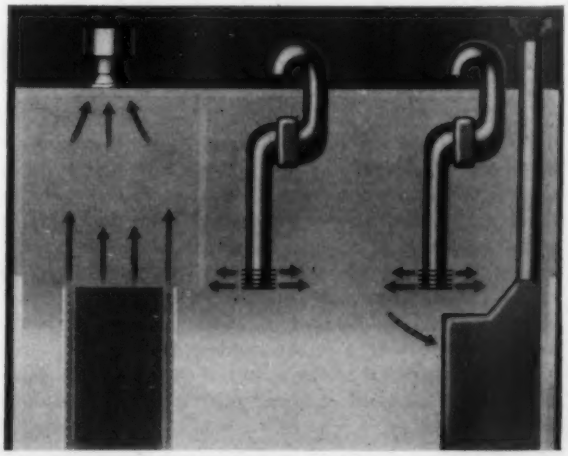


Figure 14

Figure 14 shows a method that has been worked out to increase comfort in the working zone both winter and summer, to reduce smoke and fume, and also to reduce the total cost of operation. The furnace has been shielded to stop the radiant heat and a fire curtain has been installed which separates smoke generated by this operation from the rest of the plant. The general exhaust fans have been removed from the locations where they were not needed and one of them has been placed over the heat treat area where it can exhaust the smoke and fumes. These changes will not be satisfactory unless we supply tempered air to the building proper.

As can be seen, we have eliminated our unit heaters and installed makeup air units. The air is drawn from above the roof, passed through the heaters and delivered to the working zone through the ducts and grilles shown. By the use of proper controls the air is delivered at the most desirable temperatures. Many times the air will be supplied below room temperature in order to absorb the heat generated by the machinery. Only in this way can a comfortable working zone be furnished for the employees.

Contrary to a misconception, it requires good engineering to determine the size, location and type of makeup air units which are to be used. This is even more important with the flat wide buildings that are so popular today. As an example, one plant has spent many thousands of dollars in revamping the controls even though the architect had a mechanical engineer supposedly design a proper system. In their case the controls were situated in the outside air stream so that whenever the temperature dropped below 40 degrees all outside air was closed off and, of course, this was when they needed the outside air the most.

Another thing, the air was discharged beneath the roof line which gave little or no cooling effect to the employees during the warm weather. In the case of another plant, the units were sized to allow 25 per cent of the air to be brought in from the outside. Unfortunately the dampers were of such a size that the control motors could not operate them so that outside air was not being handled at any time. They have reworked these units, added new coils and changed the controls in order to make them suitable to handle all outside air. At the same time, duct work was added so that this air can be delivered to the working zone throughout the year.

As mentioned before, this air is very desirable during the summer months as you cannot draw air through the windows of our modern wide plants and have it satisfy the people in the center of the plant. At the same time air supplied in this location will allow heat, smoke and fumes to rise to the roof where it may be easily exhausted. Another plant has an installation similar to this in their core room and they open the skylights and allow the smoke to escape by gravity. Before the makeup air system was installed the building was under such negative pressure that whenever the skylights were open cold air was drawn in and the weight of this air held the smoke from the core ovens and the hot cores in the working zone.

In order to give such an installation flexibility, the registers are arranged so they may be adjusted by the employees to project the air horizontally or 45 degrees downward as may be preferred.

It is time for engineers to design and sell equipment on what it can do for the man located on the lower 6 ft of the building rather than on the basis that their equipment will introduce more Btu's into the building for less initial cost than their competitors.

In closing, we do not heat the man but rather we must control his rate of cooling if we wish to furnish a desirable environment for him.

ANIONIC OR CATIONIC AGENTS — A SOLUTION TO SAND PROBLEMS

By

G. J. Vingas* and A. H. Lewis*

ABSTRACT

Variations in the rate of surface drying, while in the green state, of molds made of bentonite-bonded sands were found to be due to variations in the rate of moisture migration in the sand. Unfavorable effects were corrected by the addition of suitable surface active agents with a resulting overall improvement in the sand. Variations in the surface drying tendencies of different base sands are attributed to the presence of hydrogen ions and an experiment is presented supporting this hypothesis.

Introduction

Premature surface drying of green molds has been a problem in the foundry for many years. This condition leads to difficulties in finishing the molds and is especially noticeable in the case of large molds which of necessity must remain open for some time.

Some sands seem to dry out more readily than others, for example synthetic iron molding sands dry out faster than naturally-bonded sands. The usual method used to overcome this tendency is to add fuel oil or kerosene with the object of trapping the moisture in the sand and thus retarding evaporation.

Studies described in this paper were initiated with the introduction of sand reclaimed by a wet method instead of new sand in the green sand mixture in the authors' steel foundry. Introduction of the reclaimed sand caused immediate complaints from the molders that the sand was drying out too rapidly and was difficult to tool and finish. The addition of fuel oil or kerosene was tried in the conventional manner without success. The problem was eventually solved by the addition of suitable surface active agents to the sand mixture.

Materials

The reclaimed sand is obtained from a wet reclaimer and in consequence the sand grains are not completely freed from adhering carbonaceous material thus giving the sand an overall gray appearance. The sand reclaimed consists of a mixture of steel foundry molding and core sand and grade 5 S Dorchester sand base from the iron foundry. The proportions of each vary widely in accordance with the foundry operations. The Dorchester sand contains subangular high purity silica sand grains as compared to the rounded sand grains used initially in the steel foundry. The reclaimed sand is split on a 40-mesh screen and the minus 40 grade is used in the steel facing mixtures.

Table 1 lists typical sieve analyses of the minus 40 reclaimed sand, the new Ottawa washed and dried sand and the Dorchester sand.

TABLE 1 — AVERAGE SIEVE ANALYSIS OF SANDS

	AFS-GFN	30	40	50	70	100	140	200	270	Pan
Minus 40 reclaimed	51	.05	.31	38.6	40.2	14.5	2.9	0.6	0.2	0.15
Washed and Dried Ottawa No. 60	58	—	2.0	26.0	40.0	21.0	7.0		3.5	
5 S Dorchester	50-60	10	13	20	18	11	7	2.5	0.8	1.5

Table 2 gives a typical green sand mixture in the steel foundry, in use at the time of the change to reclaimed sand.

TABLE 2 — STEEL FOUNDRY GREEN SAND FACING

Reclaimed sand	Remainder
Silica flour	5 per cent
Western Bentonite	6 per cent
Corn flour	1 per cent
Soda ash	8 oz/ton
Water	4 per cent

Outline

In considering the problem of surface drying it is evident that the surface layers of a body of sand will dry out when the moisture evaporates faster than it is replaced from the interior of the sand. It is of course, simply a matter of time for the surface layers of any body of sand to dry out on standing, assuming normal atmospheric conditions, since the rate of moisture replenishment at the surface decreased with time due to the greater distance that the moisture must travel through the sand as it dries, and since the rate of evaporation tends to increase as the sand becomes dryer.

*Sand Technician and Chief Metallurgist, respectively of Dominion Engineering Works Ltd., Montreal, Canada.

Drying out is thus caused by the rate of evaporation vs the rate of moisture replenishment. Variations in the rate of water migration in different sands could thus readily explain variations in the rate of surface drying.

A simple test was run to determine if new sand absorbed water at a different rate to reclaimed sand. The effect of an anionic surface agent on this process was noted.

The results of this test were significant and a drying out test was run on a typical bentonite-bonded steel sand mixture using new sand and reclaimed sand as well as anionic surface agent, fuel oil and soluble cutting oil. This test confirmed the value of the surface active agent and showed that the fuel oil and cutting oil tended to accentuate the difference in drying out pattern between the new sand and reclaimed sand.

Information obtained at this stage was sufficiently significant to justify the use of surface active agent in the steel foundry green sand facing mixture with immediately beneficial results.

Further investigational work was undertaken to determine the effect of surface active agents on the drying out of damp sharp sands in order to eliminate any possible effects of the bonding ingredients used in the previous tests. Two different types of surface active agent were used in this test.

The aforementioned test showed a drying out pattern similar to the drying out behavior of the bonded sands, thus indicating that the effects observed were due primarily to the properties of the sharp sand rather than the bonding ingredients.

A theory was devised as to why the sharp sands should lose moisture at varying rates. It was found possible, in line with this theory, to affect at will the rate of loss of moisture, by suitable prior treatment of the sharp sand.

Experimental Work

Water Absorption of Sands

The water absorbing qualities of the new and reclaimed sand were first compared by using the same apparatus (Fig. 1) which was used to evaluate comparative water absorbing properties of bentonite. The results of these tests were inconsistent probably because of inadequate sensitivity of this test.

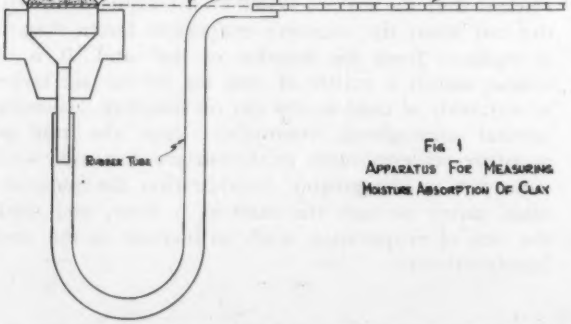


FIG. 1
APPARATUS FOR MEASURING MOISTURE ABSORPTION OF CLAY

After several unsuccessful trials the following apparatus was adopted and found to give clear and reproducible results. A glass tube 1/8-in. I.D. x 8 in. long was filled with the sand to be tested and one end of it dipped into a beaker full of distilled water; the rate of rise of water into the tube was recorded. The test was repeated with a few drops of surface active agent added to the distilled water. The results are shown in Table 3.

TABLE 3 — RATE OF MOISTURE ABSORPTION IN SHARP SAND

Sand	Rise of pure distilled water in 60 sec	Rise of distilled water plus anionic surface agent in 60 sec
New Silica	4 in.	6 in.
Reclaimed	2¾ in.	3¾ in.

NOTE — Values given are the average of six determinations. Maximum variation of any reading from the average is plus or minus 1/8 in.

The reclaimed sand absorbed water slower than the new silica sand. When the surface active agent was added the rate of absorption was increased, probably due to the lower surface tension of the water. It is interesting to note that the addition of surface active agent caused the reclaimed sand to absorb water at almost the rate as the new sand absorbed untreated water.

Loss of Water by Evaporation

A typical bentonite-bonded sand mixture was prepared with both new sand and reclaimed sand as a basis. The reclaimed sand mixture was repeated with various modifications by adding small amounts of anionic surface active agent, soda ash, number two fuel oil and water-soluble cutting oil. A description of the various mixtures is given in Table 4.

TABLE 4 — EXPERIMENTAL SAND MIXTURES FOR EVAPORATION TEST

Ingredients	Mixture No. 1	2	3	4	5	6
Silica sand	Rem.	—	—	—	—	—
Reclaimed sand	—	Rem.	Rem.	Rem.	Rem.	Rem.
Silica flour, %	5	5	5	5	5	5
Western bentonite, %	6	6	6	6	6	6
Corn flour, %	1	1	1	1	1	1
Soda ash, oz/ton	—	8	—	—	—	—
Soluable cutting oil, oz/ton	—	—	112	—	—	—
No. 2 fuel oil, oz/ton	—	—	—	240	—	—
Anionic surface agent, cc/ton	—	—	—	—	100	—
Water, %	4.5	4.5	4.5	4.5	4.5	4.5

The comparative rates of water evaporation were checked by preparing 2-in. x 2-in. standard specimens from each batch of sand, exposing them together to the atmosphere and weighing them at intervals. The losses in weight due to evaporation for various periods up to 22 hours are plotted (Fig. 2). All determinations were run in duplicate and the maximum variations from the values plotted did not exceed plus or minus 0.1 gram.

It is evident that the rate of water evaporation varied in much the same degree as the rate of water absorption. Thus the reclaimed sand mixture lost water more slowly than the new sand mixture, just as it absorbed water more slowly. Paradoxically then the slower the rate of water loss by evaporation the faster the surface dries out.

The addition of anionic surface agent caused the rate of moisture loss of the reclaimed sand mixture to exceed that of the new sand mixture. The constant high rate of evaporation showed by this curve indicates the continued efficient replenishment of the surface moisture of the sand specimen.

The mixtures containing fuel oil and cutting oil exhibit a very low rate of evaporation. If the rates of evaporation of these mixtures were relatively constant during the testing period, we might conclude that the oils were simply retarding evaporation. The rates of loss, however, show a decided decrease during the test thus indicating that surface drying occurred. Evidently the oils adversely affected the mobility of the moisture in the sand, possibly due to lack of adhesion between the oily water and the sand grains.

The beneficial effect of soda ash on the evaporation rate is quite interesting and a possible explanation will be advanced in a later portion of this paper. It might be mentioned here that the addition of soda ash alone to the reclaimed sand mixture did not produce a sand that was acceptable to the molders from the viewpoint of drying out.

In order to determine the evaporation rates of damp sharp sand it was necessary to enclose the sand in a suitable container. The procedure eventually developed involved the use of ¾-in. I.D. x 6¾-in. long test tubes. The tubes were filled by first adding 17 grams of distilled water with or without surface active agents and then 80 grams of the sand to be tested was placed on top of the water. When thus prepared the sand came to about ½-in. below the lip of the test tube.

The evaporating area was much smaller in this test than in the previous series thus necessitating a much longer time to run the tests. The migration of the water through the sand could be clearly observed and the rate of migration varied in accordance with the rate of evaporation.

The test tubes were exposed to the atmosphere

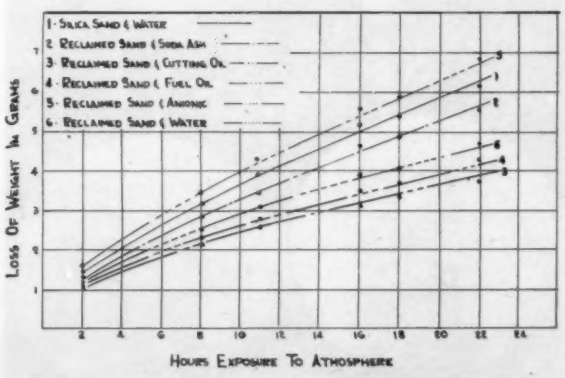


Fig. 2 — Evaporation rates of bentonite-bonded sands.

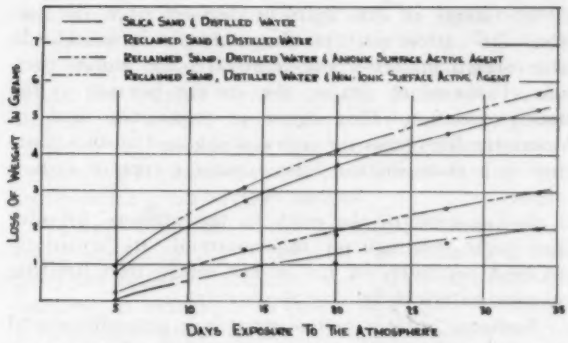


Fig. 3 — Evaporation rates of new and reclaimed sand.

and were weighed at intervals to check the loss of weight due to evaporation.

Four samples were selected for this test; new silica sand plus distilled water, reclaimed sand plus distilled water and the latter mixture to which anionic and nonionic surface active agents were added. The cumulative water loss in grams of these four mixtures over a period of 35 days is plotted (Fig. 3). All determinations were run in triplicate and the maximum variations from the values plotted did not exceed plus or minus 0.1 gram.

In comparing Fig. 3 with Fig. 2 the same pattern of evaporation rate is noted for reclaimed sand and new sand thus indicating that the mechanism of evaporation is not changed by the presence or absence of the bonding ingredients used. In both tests the new sand lost moisture much faster than the reclaimed sand with distilled water. Again in Fig. 3 the profound effect is noted of anionic surface active agent in increasing the evaporation rate of the reclaimed sand. The nonionic surface active agent is not nearly as effective as the anionic.

In order to determine if these surface active agents affect the evaporation rate of distilled water, several evaporating dishes were filled with distilled water and to some, a few drops of the different surface active agents were added. The losses in weight on exposure to the atmosphere did not disclose any significant difference.

Surface Active Agents

Surface active agent is the correct term for materials known more commonly as detergents or wetting agents. Technically it is defined as a solute having such properties as to alter the surface energy of the solvent.

In general, surface active agents have the property of partial solubility in water and in the foreign material requiring emulsification. The agent thus forms a link between the water and the foreign material thus promoting emulsification and removal of the foreign matter. Details concerning three types of surface active agent are presented hereunder.

Anionic surface active agent is a sulphonated aliphatic polyester. The pH of a 1 per cent solution is 6.5-7.5. The agent is stable in moderately acid or alkaline solution but breaks down when heated moderately. Its flash and fire points respectively are 94 F and 104 F.

The name of this agent is derived from the fact that the cation part of the molecule is soluble in the foreign matter where it attracts the anions present. It therefore acts on the anionic portion of the foreign matter. This agent is commonly used in domestic detergents for dishwashing and clothes-washing. It is generally the least expensive type of surface agent.

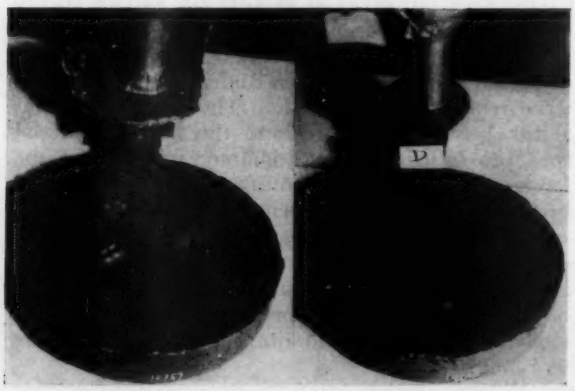
So far, most of the work in the authors' foundry has been centered on this material. Its sensitivity to heat precludes its use in hot sands, thus limiting to some extent its field of application.

Nonionic surface active agent is a polyethoxylated nonylphenol, which is not affected by hard water, metallic salts or alkalies. It is stable at elevated temperatures. The flash and fire points are respectively, 535-555 F and 570-630 F. The name of this agent is derived from the fact that both anion and cation in its molecule are soluble in the foreign matter. It therefore does not act specifically on either cation or anion in the foreign matter. This agent is commonly used in heavy duty detergents for industrial floor cleaning and the like where high alkalinity or acidity may be encountered. It tends to be somewhat more expensive than the anionic agent.

The tests reported here indicated that the nonionic is not as effective as the anionic agent in promoting evaporation and this was confirmed by actual trials in the foundry.

Cationic surface active agent consists basically of alkyl dimethyl benzyl ammonium chlorides in which the alkyl group ranges from C_8 to C_{18}. It is compatible with most alkalies, phosphates and nonionic and cationic surface active agents. However, like other high molecular weight quaternary ammonium compounds, this agent is not compatible with soap or anionic synthetic surface active agents. It is stable in boiling water, but it decomposes at elevated temperatures in the presence of excessive alkalies. The name is derived from the fact that the anions in its molecules are soluble in the foreign matter thus acting on the cations present. This agent is commonly used in sprays and insecticides. It is a newer development than the other two types and is the most expensive of the three.

Fig. 4 (Left) — Plate cast in sand containing fuel oil (Table 5), cope side; Fig. 5 (Center) — Plate cast in sand containing anionic agent (Table 5), cope side; and Fig. 6 (Right) — Drag side of plates shown Figs. 4 and 5.



Preliminary work with this material has indicated favorable results but so far there have been insufficient data available to justify its use in the foundry.

Practical Applications

The effects observed in the usual sand tests of surface active agent addition to a high silica flour mixture are illustrated in Table 5.

TABLE 5 — EFFECT OF SURFACE ACTIVE AGENT ON A HIGH SILICA FLOUR MIXTURE

	Old	New
New Silica sand, %	remainder	remainder
Silica flour, %	22.5	22.5
Western Bentonite, %	6.5	6.5
Corn flour, %	1.0	1.0
Iron oxide, %	1.5	1.5
No. 2 fuel oil, oz/ton	96	—
Anionic agent, cc/ton	—	30
Water, %	7.5	7.0
Green compression, psi	10 - 12	9 - 11
Permeability	60 - 80	50 - 70
Dry shear, psi	35 - 65	85 - 115

It will be noted that the replacement in this mixture of anionic agent for fuel oil caused a slight drop in green strength and permeability and a large increase in dry strength. The reduced green strength and permeability are probably a reflection of the improved flowability that was noted. The higher dry strength may be due to improved clay distribution and efficiency. It was observed that the substitution of the anionic agent for the fuel oil produced a more consistent sand from batch to batch.

The old sand mixture shown in Table 5 was used in a large mold requiring considerable finishing. In spite of the fact that new silica sand, rather than reclaimed sand, was used, difficulty was encountered due to drying out and crumbling of the mold surface. Replacing the fuel oil with surface active agent minimized the drying out problem and improved the moldability of the sand.

The casting finishes produced by the two sands were compared by casting two plates in each sand, 8½ in. in diameter and 1½ in. thick. The fuel oil mixture showed dirt inclusions on the cope side of each plate while the surface active agent mixture gave plates that were completely clean on the cope faces (Figs. 4, 5, 6).



TABLE 6 — SAND MIXTURES WITH SURFACE ACTIVE AGENTS, IN USE IN THE STEEL FOUNDRY

Green molding sand.			
Reclaimed sand	Remainder	Green compression, psi	8 to 10
Silica flour, %	5	Permeability	100 to 150
Western bentonite, %	6	Deformation, in.	0.030
Corn flour, %	1		
Soda ash, oz/ton	8		
Anionic surface agent, cc/ton	15		
Water, %	4		
Roll molding sand, dry.			
New silica sand	Remainder	Green compression, psi	10
Silica flour, %	20	Permeability	40
Fire clay, 20 mesh, %	10	Deformation, in	0.030
Wood flour, %	0.5	Dry shear, psi	100
Corn flour, %	1		
Anionic surface agent, cc/ton	15		
Water, %	6½		
Heavy duty molding sand, dry.			
New silica sand	Remainder	Green compression, psi	12
Silica flour, %	20	Permeability	40
Fire clay, %	15	Deformation, in	0.040
Western Bentonite, %	5	Dry shear, psi	90
Wood flour, %	0.5		
Corn flour, %	1		
Anionic surface agent, cc/ton	15		
Water, %	6½		
Standard core sand.			
New silica or reclaimed sand	Remainder	Green compression, psi	7
Silica flour, %	14	Tensile, psi	110
Western bentonite, %	3.5	Permeability	50
Natural resin, %	1.4	Deformation, in.	0.024
Wood flour, %	0.7	Baked hardness	80
Anionic surface agent, cc/ton	15	(reclaimed)	
cc/ton	7	(new silica)	
Water, %	6.75		
Heavy duty core sand.			
New silica sand	Remainder	Green compression, psi	5½
Silica flour, %	30	Tensile, psi	140
Natural resin, %	1	Permeability	15
Western bentonite, %	1.2	Deformation, in.	0.025
Anionic surface agent, cc/ton	10	Baked hardness	85
Water, %	8		

Table 6 shows the various sand mixtures in which anionic surface active agent is presently in use. The average properties of these mixtures are also shown.

In the case of the sands listed in Table 6, it is not possible to furnish the average properties obtained without the addition of surface active agent, to compare with the properties given, since other changes were made at the same time that the surface active agent was added. In general, the usual laboratory test properties seemed to be most noticeably affected in the mixtures containing a high percentage of silica flour. Similar advantages in the foundry were noted in these mixtures, after the addition of surface active agent, to those noted previously for the mixture given in Table 5.

Theory

Booth[1] has shown that increasing pH decreases green strength, and increases dry strengths in foundry molding sands. The experiment reported here showed that similar effects are observed in certain cases when surface active agents are added to the sands.

Ludwig[2] states that wetting agents although they lessen the helpful surface tension forces, are useful in that they spread the liquid activity to more points of action. He also shows that air could be entrapped or attached by adhesion to the surface of the capillary valley of the grain to grain contact.

Oulton[3] states that acid treatment of montmorillonite results in the removal of the aluminum ions from the lattice of the clay, and that after thorough washing to remove the excess acid, there remains a substantial quantity of hydrogen ions attached by adhesion to the surface of the clay.

Davies[4] states that clays and sands are derived, principally from ancient rocks by disintegration, both chemical and mechanical, and by pneumatolysis, that is to say that clays and sands have been at sometime exposed to acid actions.

It is therefore possible that clays and/or sands as mined or as washed or as exposed to the actions of the organic acids, resulting from the chemical decomposition of the organic binders added to the sand mixtures, could acquire adhesion of hydrogen ions on their surfaces. Oulton[4] proved that clays could have such an adhesion that could affect the performance of the clay. This could explain Zang's[5] theory and findings on pH of sands, i.e. the additions of sodium carbonate into the sand mixture neutralizes the free hydrogen ions, thus permitting better flocculation and distribution of the clay into sand mixture.

An effort was made to prove that the same mechanism could occur in the sand grains themselves, i.e. hydrogen ion could adhere to the surface of each individual sand grain, affecting the capillary forces that drive the water in the mass of sand toward the surface, to replace the water that has evaporated. This should be detected by measuring the rate of water evaporation in a sand that was loaded with hydrogen ions.

A few pounds of new silica sand were boiled in a 10 per cent solution of sulphuric acid for an hour and the sand was then washed with distilled water until the water showed a neutral color with pH paper. A sample was kept and the same sand was boiled in a 5 per cent solution of sodium hydroxide and washed with distilled water until a neutral solution was obtained. This accentuated the conditions;

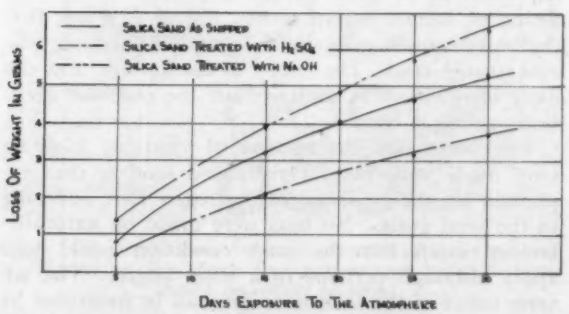


Fig. 7 — Effect of prior treatment on evaporation rate of new sand.

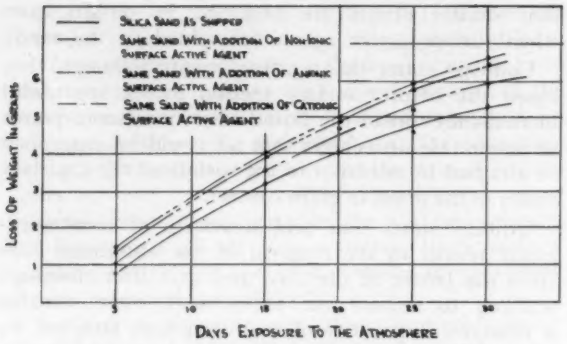


Fig. 8 — Effect of surface active agents on evaporation rate of untreated sand.

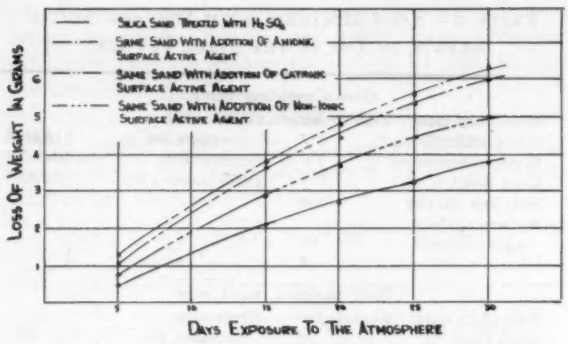


Fig. 9 — Effect of surface active agents on evaporation rate of acid-treated sand.

if any hydrogen ions could adhere to the surface of the sand they would have done so when boiled in the acid solution; the hydrogen ions would be completely removed when the sand was boiled in the hydroxide solution.

The sands were dried at 212 F for two hours and the evaporation test described previously was repeated. As before, 17 grams of distilled water with or without additive were placed in a ¾-in. I.D. x 6¾-in. long test tube and 80 grams of the sand to be tested placed on top. The rate of evaporation was measured by weighing the tubes at intervals.

Evaporation rates of the three sands with distilled water are plotted in Fig. 7. The effect on the evaporation rates of the three types of surface active agent are shown, when added to the untreated sand (Fig. 8) and when added to the acid treated sand (Fig. 9). The surface active agents did not affect the evaporation rate of the caustic-treated sand. All tests were run in triplicate and the maximum variation of any reading from the average did not exceed 0.1 gram.

Reference to Fig. 7 shows that the evaporation rate of the untreated silica sand is reduced when the sand is treated with acid and increased when it is treated with caustic. This behavior is readily explained by the presence or absence of hydrogen ions thus serving to confirm the theory that hydrogen ions affect the migration rate of moisture in the sand. The previously noted beneficial effect of soda ash on the evaporation rate of a molding sand is readily explainable on this hypothesis since the soda ash being alkaline, will tend to neutralize any hydrogen ions present.

Curves in Fig. 8 show that the evaporation rate of untreated silica sand can be improved by the addition of surface active agents. Figure 9 shows that the same agents can almost completely restore the acid-treated sand. The effects of the anionic and cationic agents are very similar while the nonionic agent is not nearly as effective.

The premature drying out of synthetic molding sand made with new or reclaimed sand is thus explained by the presence of hydrogen ions adhering to the sand grains. No tests were made on naturally-bonded sands but the same condition could well apply although perhaps to a lesser degree. The adverse effect of the hydrogen ions can be overcome by the addition of small quantities of surface active agent, either anionic or cationic. It has been noted also that continued treatment of the sands with soda ash to maintain a pH value of eight to nine, has reduced the drying out tendency. For example, the difference in drying out rates of the reclaimed sand and the new sand is not nearly as great now as it was when the tests reported here were run, since soda ash additions were begun at about the same time as this work commenced.

The work reported here is concerned with the effect of surface active agents on sharp sand. It is quite probable that these agents have important effects on clay matter as suggested by Bradley's[6] statement, that montmorillonite clays have been found to form complexes with polyamines and polyglycoles. Surface active agents could form similar complexes with clays thus affecting their plasticity. Further investigation is necessary to determine these effects.

Summary

1. The relative rates at which the surface layers of bodies of different sands dry out are a function of relative rates of migration of moisture in sands rather than inherent differences in evaporation rates.

2. Unfavorable drying out effects, notably in synthetic sands particularly when made with reclaimed sand, can be overcome by the addition of anionic or cationic surface active agents with a resulting overall improvement to the sand.

3. Variations in the tendency to dry out of various sands are attributed to hydrogen ions which may be present to a greater or lesser degree. Maintenance of a high pH content in the sands will minimize the drying out tendency.

4. More investigation is needed to determine the effects of surface active agents on clay.

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ENGINEERING, COST, AND QUALITY

By

James M. Barrabee*

ABSTRACT

Under the stress of today's intense competition, the foundry industry finds itself faced with the problem of engineering its castings for cost and quality. From the initial drawing, through the foundry processing, to its ultimate delivery to the customer, the casting must be contemplated from all angles. Tolerances and finishes, foundry methods, quality control, inspection, and the necessary communication between engineering and production are all phases to be considered. The effect of these various phases are discussed — the ultimate goal being higher quality castings at competitive cost.

The need for improved engineering, costs, and quality in the casting field should come as no surprise to any foundry organization in view of today's highly competitive market. Cost and quality must be engineered into a casting from its very inception on the designer's board to its ultimate delivery to the customer. New molding processes, sand developments, metallurgical advances, as well as materials handling and supervisory techniques make it necessary for each phase of foundry operation to be studied with an eye to better castings at competitive costs.

Quality actually begins with a blueprint—a casting print is preferable. Too often the foundry is expected to use divine wisdom in executing their equipment design from a finished piece part print. All basic dimensions should be given from a fixed location point, continuous dimensioning from one point to another can often lead to a build-up of errors. Tolerances should be consistent with the process being used—green sand molding, shell molding, investment casting, diaphragm molding, etc. For example, such as-cast tolerances as 128ths or 0.0075 in green sand are often not only inconsistent with the process, but are often used with no inherent knowledge of the working function of the individual part. Haphazard dimensioning can also be one of the main reasons for losing a job due to inability to hold specified dimensions. The customer should be consulted, and all critical dimensions and tolerances should be reviewed with an eye toward the best functioning part at a reasonable cost.

The pattern shop holds a commanding position in regard to cost, quality, and engineering. Sound patternmaking practice as well as that of molding, melting, coremaking, and chipping is essential to a quality engineered casting. Many of the operational difficulties occurring in both the foundry and the shop can often be traced back to the pattern design and gating system as developed from the engineer's drawings. Do we have engineers and foundry personnel who are

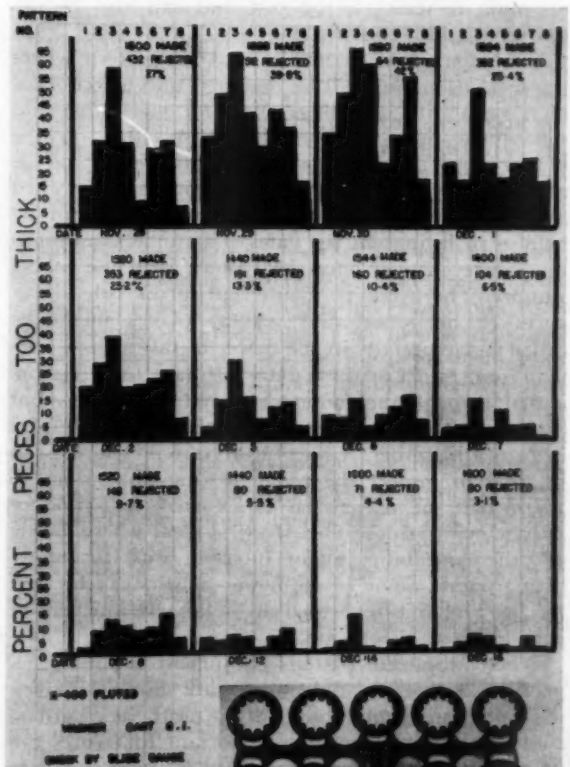


Fig. 1 — Complete history of scrap reduction of fluted washers is depicted on this chart. One pattern was found to have caused two-thirds of scrap.

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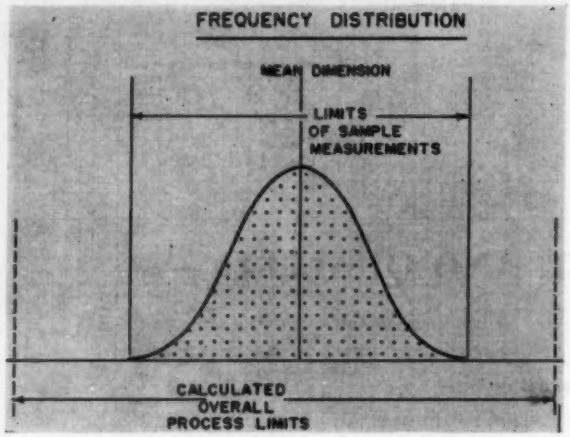


Fig. 2 — Frequency Distribution Curve.

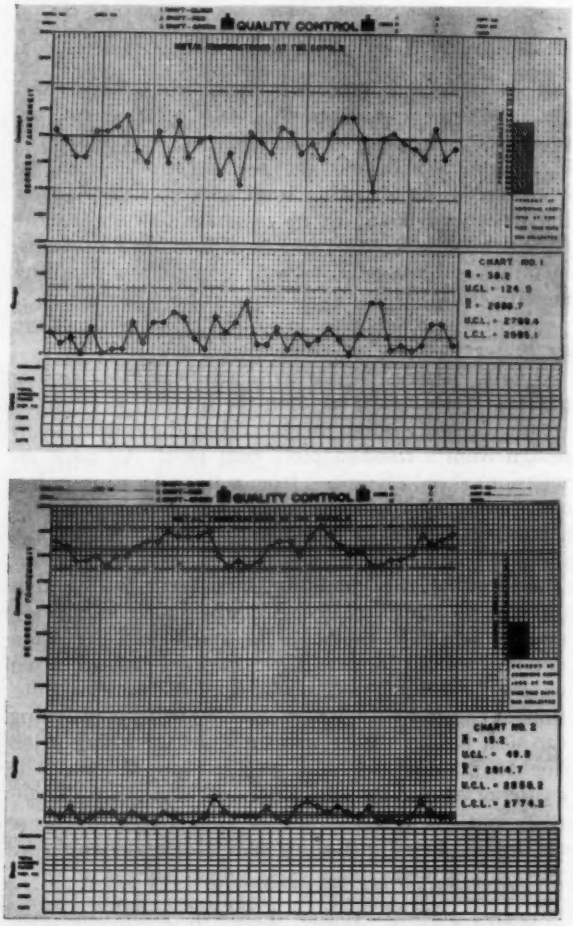
fully experienced in all phases of foundry operation, so that they are able to design equipment which will give satisfactory service with little or no rework? The price of quality can be very high and a good designer can often make the difference between success and failure on a job bid. Consumers cannot be expected to go along with numerous pattern changes which not only increase casting cost, but can also be the source of costly time delays.

No patternmaker wants to make a pattern where the locating points will be designed into a section of high variability. Locating points for machining and subsequent operations should be specified on the blueprint so that any source of future error can be eliminated. Gating practice based on tables worked up from volume, weight, and pouring head can be helpful in laying out the gates and runners, minimizing future changes once the pattern is in use in the foundry. The necessity of different shrink rule allowances in horizontal and vertical planes might be called for—especially in the newer metals such as nodular iron. Much of these data will have to be accumulated through the foundry's actual experience, but explicit records of such details can be most helpful in future planning. A flagging fixture or templet can be used to inspect the rough casting dimensions to assure that the pattern, as constructed, will produce satisfactory castings. This setup is ideal where gating or other variability might interfere with future locating points. Some of the newer plastics can be used for jigs and fixtures which are relatively inexpensive and do a satisfactory job.

Where duplicate pattern equipment is in use, especially in cope and drag work, care should be taken to assure interchangeability of patterns. In regard to core boxes and dryers it is important that cores can be used with all the combinations of cope and drag patterns. Wherever possible, critical cored locating pads, core prints, or sections should lie on the side away from the drier affording the least opportunity for crushing in handling. Where cored surfaces are used for locating points in the finished casting, the necessity for a clean core is easily understood.

Many foundries today have a regular maintenance schedule for production patterns and core boxes. At the end of each run, the pattern or core box is first returned to the pattern shop for a dimensional check before being stored in the pattern vault or sent to the molding floor. Such checks eliminate the need for hasty repairs which were forgotten once the end of the previous run was reached. Similar arrangements can be made with the maintenance department on molding, coremaking, and mill room equipment as a preventive maintenance measure.

A characteristic card on each casting can be drawn up listing such items as pattern equipment, driers, chaplets, number of pieces on a plate, flash size, etc. This is of importance to both production and maintenance personnel. All repair work and engineering changes should be noted on this card to keep the foundry abreast of any changes which might vitally affect the scrap produced during the run of any particular casting. For example, on an eight-part pattern, it takes only one bad section to produce over 12 per cent scrap on that run. A break-down by individual piece numbers on that plate (Fig. 1) in histogram form can be very informative in isolating the single section or gating portion at fault.



Figs. 3 and 4 — Metal temperature and scrap record was improved (50 per cent less loss due to cold metal) after better melting and handling techniques were introduced. Fig. 3 (Top) was made before improvement, Fig. 4 (Bottom), after.

Inspection and quality control cannot be overlooked in their relation to engineering, costs, and quality. The new advances in quality control, statistical or otherwise, have already proven themselves beyond a doubt. Simple statistical tools such as the previously-mentioned histogram should be employed wherever necessary, for such a picture can tell a story far better than a jumbled mass of unrelated data. Quality control is ideal for use in the processing departments of the foundry, and it can effectively be coupled with good inspection practices on finished castings.

Too often, however, the inspection department finds itself saddled with the same problems as the pattern shop and the foundry. How often do we find that the requirements set on a casting necessitate a large amount of excess grinding and polishing in the mill room to meet specifications? These operations are not inexpensive and in most cases could have been avoided if proper liaison was carried out before production was begun. The foundry will naturally do everything in their power to meet specifications, but it must also be acquainted with the processing of the casting once it leaves their doors. Heavy production runs which must be reworked due to inadequate lead time in relation to machining should be avoided at all costs. Too often a large run is made based on a casting layout which meets specifications, but is later found inadequate when the machining is completed.

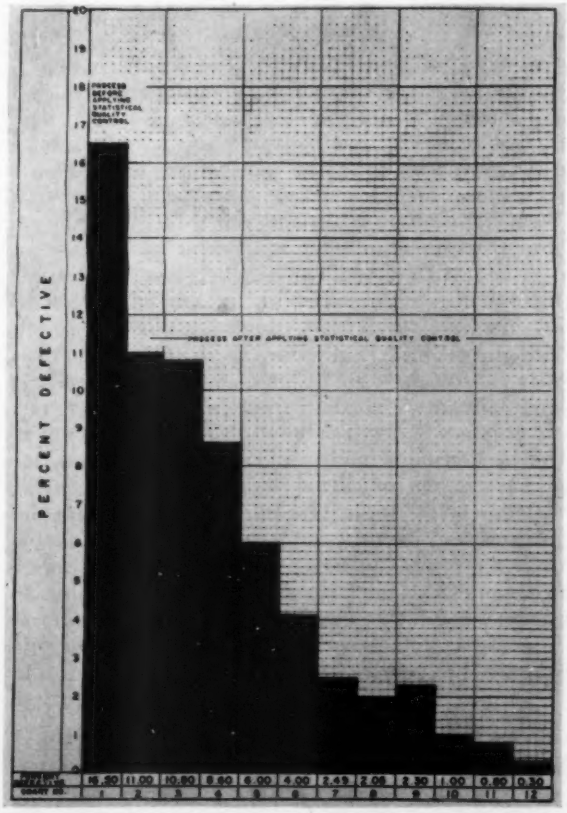


Fig. 5 — Per cent defective cores dropped steadily after statistical quality control was applied.

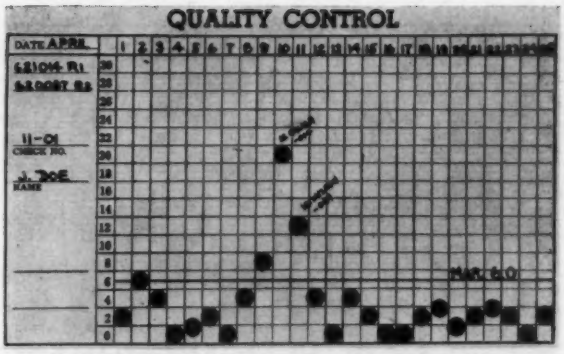


Fig. 6 — Simple chart for use in recording individual molder's performance in a floor molding shop. Control in this case must be based on overall charting picture of molder performance for effective results.

Extra time may be spent in the mill room on sections which are irrelevant while a critical area may pass unnoticed. This is especially important if any incentive plan is used to set production rates which have a vital effect on the casting cost. Any deviation from established standards cannot only create new problems from the cost angle, but can also result in strained labor relations.

Quality control fits very neatly into almost every operation in the foundry, and can also be used effectively in conjunction with engineering. The frequency distribution, for example, based on a sampling procedure can give an accurate prediction of tolerance spreads. Measurements can be taken on a random group of castings and a reliable answer can be formulated to the question, "Can we hold this specified tolerance and within what limits?" (Fig. 2). It can be easily seen how effective such a presentation can be in determining whether the requirements as set by engineering can be met with the processing procedures set up in the foundry. As a result, it may be necessary for the foundry to modify its current practice, or have the engineering department review its blueprints to see if such requirements can be altered to relieve the problem. The cost of such changes early in the planning stage far outweighs such changes made after production has commenced.

Control of raw materials can be studied, for without consistency or definite knowledge of lack of consistency, little can be done to isolate or make adjustments for some of the basic causes of casting defects. Items such as sand, pig iron, scrap, chaplets, lumber, etc., should all be sampled to determine conformity to desired specifications.

The use of simple control charts has proven invaluable in the various departments in providing a visual means of measuring the variation in men, machines, and material. The "$\bar{x}$R" or average and range chart shows not only the level of a certain operation, but also measures the consistency with which the level is held. In sand preparation, such a chart can be applied to the various constituents which make up the mixture. Similarly it can apply to cupola charging where variation can have an adverse effect on the melt. How often has a charging operator var-

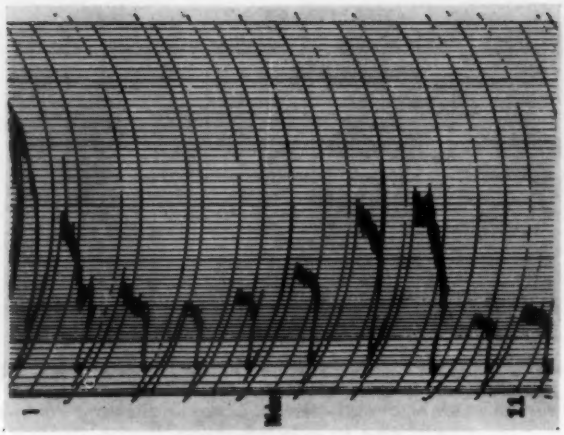


Fig. 7 — Wattmeter chart for facing sand.

ied the mixture to correct for a previous overweight charge, thereby raising havoc with analysis and fluctuation in temperature? Cupola temperatures plotted in this manner (Fig. 3 and 4) give a clear indication of variations in practice and the results after their correction.

The "P" or per cent defective chart is handy in presenting a graphic picture of such items as per cent of scrap castings, core defects (Fig. 5), unit performance, and individual molders scrap (Fig. 6). Such charts can show trends in production and like the xR chart often can be used to isolate the reasons for such fluctuations.

Another control not quite as statistical is the wattmeter chart (Fig. 7) which automatically measures the load on a motor. Attached to a muller, for example, it can record not only the cycle time but also the load carried by the motor during the preparation of the mixture. By correlating the amounts of the various constituents added during the cycle to the peaks on the chart, a consistent mixture can be produced.

The tangible results of a quality control program cannot be easily measured. It can be studied only on its effect on the operating and engineering departments. Costwise it is often condemned because additional manpower is generally required to initially staff such an organization. However, if the performance of the foundry departments can be shown in its relation to an improved quality picture, there should be no doubt as to its effectiveness and worth.

Thus we have seen that it is just about impossible to talk individually about engineering, costs, and quality without discussing their interrelationship. Any job shop, captive shop, floor molding shop, or highly automated foundry, therefore, in order to meet the keen competition in today's casting market must be aware of all three. Careful planning in the initial stages, with consideration given to the proper function of each foundry department, can result in a well engineered casting, of high quality, competitively priced.

A TENSILE BAR SHELL MOLD FOR LIGHT ALLOYS

By

Lt. K. L. Herrick* and R. C. Harris*

ABSTRACT

A tensile bar mold design for the shell molding process has been developed using the principles of fluid flow. This design causes the metal stream entering from opposite ends of the tensile bar mold cavity to join at the grip end of the bar. Tensile properties of 356, 220, 40E, and 195 alloy were evaluated at 1400°, 1325°, and 1250° F pouring temperature and with four different riser sizes. Properties of 356, 220, and 40E cast in shell molds were found to be equal to or slightly better than the tensile properties of the alloys when cast in green sand molds. Properties of 195 cast in shell molds were significantly lower than the tensile properties of 195 when cast in green sand molds.

Thermal gradients along the axis of the tensile bar were essentially symmetrical from the center of the gage length. Riser size did not effect thermal gradients in the gage length portion of the shell cast tensile bar but did effect solidification time.

Introduction

The separately-cast tensile-test bar has long been the principal means for determining and controlling melt quality in making castings for structural applications. Each casting method has associated with it some test-bar mold design which has evolved through the years and has become the "standard" design for the method. With the emergence of the shell molding process as a practical foundry method, a need exists for a well designed test bar mold.

Due to the unique properties of the shell mold, the use of sand-cast tensile bars is not justified. For example, the shell material has different thermal properties than green sand. This leads to low thermal gradients and associated feeding difficulties within the solidifying shell mold casting. Thermal gradients and solidification rate, while important in any casting operation, are of particular importance in the casting of aluminum-base alloys. The inherent physical characteristics of aluminum (low melting point, high thermal conductivity, and high heat of fusion) are not conducive to the development of favorable temperature gradients or solidification rates but, instead, favor the attainment of thermal equilibrium throughout the casting before solidification is completed.

It can be expected that mechanical properties of shell-mold test bars would be different from those obtained in green sand. The value of these properties obtained with careful foundry practice, will serve as a criterion of melt quality when the shell-cast test bar is employed. Similar data for copper-base alloys have already been published by Ames, Kahn, and Donner.†

Object

The object of this study was to develop a tensile-bar mold design suitable for use in casting aluminum by the shell-mold process.

The work described in this report is divided, essentially into two parts. The first is the development of a satisfactory tensile bar mold. This mold incorporates a horizontal gating system designed in accordance with fluid flow principles developed under Ordnance Corps sponsorship at Battelle Memorial Institute. The mold was studied with respect to the following process variables: gating arrangement, risering practice, and pouring temperature. The second part of the work consists of determination of the level of properties which could be expected from several aluminum-base casting alloys utilizing what was considered to be the most favorable shell mold design.

Mold Design Considerations

A test-bar pattern plate employing a horizontal gating system was selected for study. While early shell-mold practice tended toward vertical gating systems, the current trend is in the direction of horizontal systems. This is due largely to the fact that horizontal gating results in lower hydrostatic pressures, thus, eliminating the need for back-up material.

One objection to many tensile bar mold designs is that the metal stream entering the mold cavity meets within the gage length of the tensile bar. This condition can often lead to entrapment of dross and oxides at that point which is the most critical portion of the bar. One method which has been used

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†Ames, Donner, Kahn, "Metallurgy of Shell Molding," AMERICAN FOUNDRYMAN, January 1952.

in the past to overcome this objection is to allow all the metal going into the risers farthest from the sprue to flow through the tensile bar cavity. This arrangement, increases the chances of mold washing. In addition, the mold material in the gage length area of the mold is overheated leading to the formation of unfavorable temperature gradients within the casting. Several gating arrangements have been reported which overcome both these difficulties by the use of inclined tensile bar cavities. This solution is applicable to sand molds but would make the machining of a shell molding pattern plate overly complicated.

Fluid flow principles offer a less complicated means of overcoming both of these objections. By controlling the cross-sectional area of the various parts of the runner system, it is possible to control the amount of metal entering each end of the tensile bar cavity without the use of chokes or similar techniques.

Fluid flow principles as applied to horizontal gating systems have been carefully studied.* A good gating design should make it possible to fill the mold cavity completely with the least amount of turbulence and to maintain proper directional solidification. To accomplish this, individual portions of the system require special attention. Good gating and pouring practice require the use of pouring basins, tapered sprues, properly designed sprue bases, and streamlined gates utilizing proper gating ratios.

Methods and Materials

The Pattern Plate

The patterns and pattern plate were machined from aluminum with the exception of the tensile bar halves, which were machined from steel. The plate was designed so that with minor changes both the cope and drag shells could be made from the same plate. This accomplished two purposes: only one plate had to be fabricated utilizing removable parts, and positive alignment was assured because the tensile bar halves and the male and female alignment pins were never removed from the plate. A separate sprue and pouring basin assembly was attached to the cope portion of the pattern plate. Small bolts were used to fasten pattern parts to the plate and also to fill in unused holes. Various size patterns were also machined from aluminum. The riser patterns were mounted in this way. The riser patterns were also machined from aluminum. The four sizes used were 2½ in. in diameter by 3 in. high; 2, 1½ and 1⅛ in. in diameter by 2 in. high. This procedure made a study of different gating arrangements much more practical. In order to compensate for ovality of the cast tensile bar as a result of parting line effects, each tensile bar half had 0.007 in. machined off the flat side. The pattern plate is shown in Figs. 1 and 2.

*SYMPOSIUM ON PRINCIPLES OF GATING, AFS (1951).

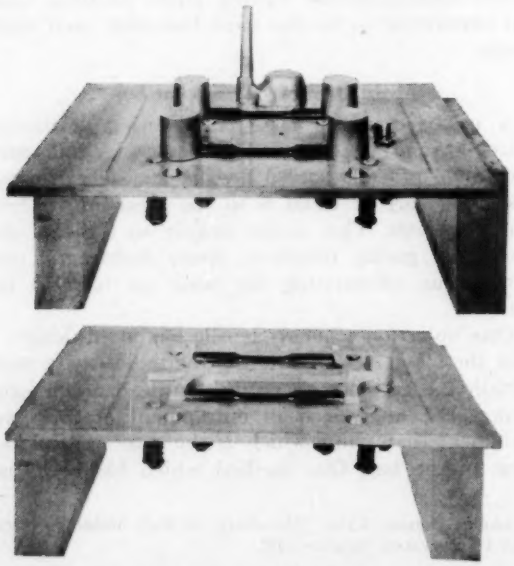


Fig. 1 (Top) — Shell-mold pattern plate for making cope portion of mold and Fig. 2 (Below) drag portion.

Molding Procedure and Shell Assembly

Tensile bar molds were made from a shell molding mixture having a composition as shown in Table 1.

TABLE 1. — SHELL MOLDING SAND MIXTURES

Component	Weight, %
120-150 AFS fineness sand	94.8
Proprietary resin	5.0
Kerosene	0.2

One hundred-pound batches of sand were mulled dry for a short period in a portable sand muller and the kerosene wetting agent added. After two minutes of mixing, the dry resin was then progressively added and the entire batch was mulled for a period of ten to fifteen minutes.

After the pattern plate was cleaned, it was coated with silicone spray mold release and put in an electric furnace which was maintained at 600 F. When the plate temperature reached 425 to 450 F the plate was removed from the furnace. A transite gasket the size of the dump box was then clamped in place and the plate clamped to the dump box. The purpose of the transite gasket was to insulate the lip of the dump box from the heat of the pattern plate. This prevented the resin from adhering to the dump box and thereby eliminated "drop oils." The dump box used contained 20 to 25 pounds of sand. New sand was added as it was used to keep the sand depth about eight inches. The dump box was inverted and held for 15 to 20 seconds. The dump box was then returned to its original position and the pattern plate put back in the 600 F furnace for an additional curing time of 2 to 2½ minutes before the shell was stripped from the pattern plate. A shell thickness of 3/16 to ¼ in. was held constant and all molds were poured without any back-up material.

The shell halves were held together by eight ¼ in. diameter bolts. The sprue was cut off and assembled to the shell mold as a separate piece using

Table 2 — Composition of Aluminum Alloys, Per Cent

Alloy	Al	Cu	Si	Mg	Zn	Fe	Mn	Ti	Cr	Others
40E	Remainder	0.3	0.25	0.5 to 0.65	5.2 to 6.0	1.0	0.3	0.1 to 0.25	0.4 to 0.6	0.05 ea.
220	Remainder	0.2	0.2	9.5 to 10.6	0.1	0.3	0.1	0.2		0.05 ea.
195	Remainder	4.0 to 5.0	1.5	0.03	0.3	1.0	0.3	0.2		0.05 ea.
356	Remainder	0.8	6.76	0.34	0.2	0.09	0.01	0.11		0.01 ea.

heated shell molding mixture as a cement for holding the sprue in place.

Melting Stock

Three commercial aluminum alloys used in this investigation were 220, 195, and 40E. Spectrographic analysis of the alloys showed that they were within the allowable compositional ranges. The fourth alloy tested was a recently introduced variation of 356 alloy having a low iron content. The nominal compositions of 220, 195, and 40E are shown in Table 2 along with the composition of the special 356 alloy.

Melting Procedures

A 3000-cycle induction lift coil furnace was used for melting purposes. Degassing was accomplished by passing dry nitrogen through the 220 and 356 alloys and chlorine through the 195 and 40E alloys for 15 minutes with the melt temperature approximately 1300 F. Chlorine is considered the best degasser but nitrogen is satisfactory if it is absolutely dry. Chlorine was not used on the 356 and 220 alloys because of the loss of magnesium resulting from the chemical reaction between the two elements. This loss of magnesium is not too critical in 220 but is very important in 356. The efficiency of degassing was determined by means of the reduced pressure solidification test. Density measurements of the samples solidified at reduced pressure were used as a basis for calculating the gas content. Density samples were taken immediately after degassing at 1300 F. When pouring temperatures higher than 1300 F were desired, the metal temperature was raised. Temperature measurements were made using a portable potentiometer and a chromel-alumel thermocouple encased in a stainless steel tube.

Heat Treatment

The heat treatments for the alloys are shown in Table 3.

Table 3 — Heat Treatment Cycles

Alloy	Solution Treatment	Quenching	Precipitation	Cooling
195	16 hr at 960 F	Water at 150 to 212 F	4 hr at 310 F	Air
356	16 hr at 1000 F	Water at 150 to 212 F	4 hr at 310 F	Air
220	20 hr at 810 F	10 min in oil at 250 to 300 F; water quench at 70 F		
40E			10 hr at 355 F	Air

All solution and precipitation temperatures were controlled to ±5 F. All heat treating was done in a circulating air furnace. The heat treatment of 195 and 356 is the standard T6 condition. An optional 21-day precipitation at room temperature treatment can be substituted for 40E.

The heat treatment used for the 220 alloy is outlined in Table 3. This alloy is used in the solution-treated condition since artificial aging impairs its resistance to corrosion and imparts mechanical properties that are less interesting from an engineering viewpoint. The castings are quenched in hot oil (250 to 300 F) and cooled to the oil temperature. The charge is then transferred into water at room temperature to complete the solution treatment.

Results

Mold Design

The original calculations of the dimensions of the sprue, runner, and gate system of the mold were based on a flow rate of metal entering the mold of 1.25 lb/sec. Subsequent measurement of this flow rate showed that the actual flow rate was very close to the intended rate.

One of the important considerations in the design of the pattern equipment was to avoid having the metal stream meet at the center of the tensile bar. To determine where the metal streams actually met in the two tensile bars, vermiculite balls were placed in the mold cavity at the time of assembly. When the metal was poured into the mold the vermiculite balls originally held in position by fine iron wires were pushed ahead of the metal stream until stopped by the stream coming from the other direction. The castings were then radiographed to determine the location of the vermiculite balls and to show where the metal streams met.

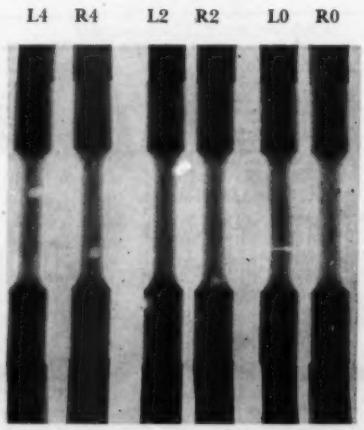


Fig. 3 — Radiograph of tensile bars showing location of vermiculite balls in molds poured at different angles of incline. R4 was broken when cut from casting. Black lines are wires used to hold balls in position before pouring. L0 and R0, 0° incline; L2 and R2, 2° incline; L4 and R4, 4° incline.

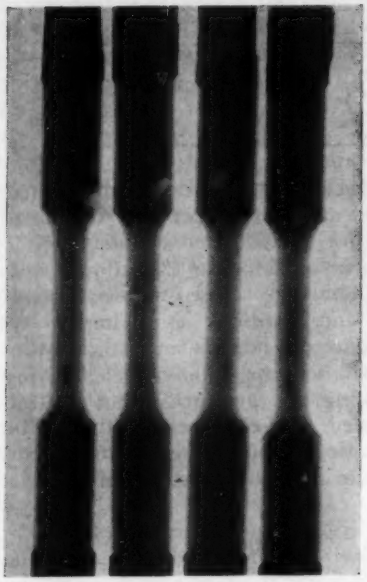


Fig. 4 — Radiograph of tensile bars showing location of vermiculite balls in molds poured at 0° incline. Black lines are wires used to hold balls in position before pouring. Light portion in gage length of R1 is a fleck of vermiculite. L1 and R1, main runner decreased 1/8 in.; L2 and R2, main runner and far ingates decreased 1/8 in.

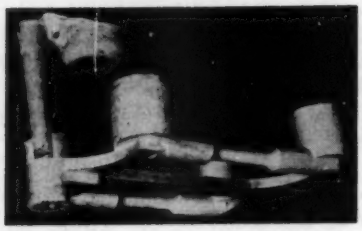


Fig. 6—Casting showing gating arrangement of tensile-bar shell mold. The risers are 2 in. in diameter and 2 in. high.

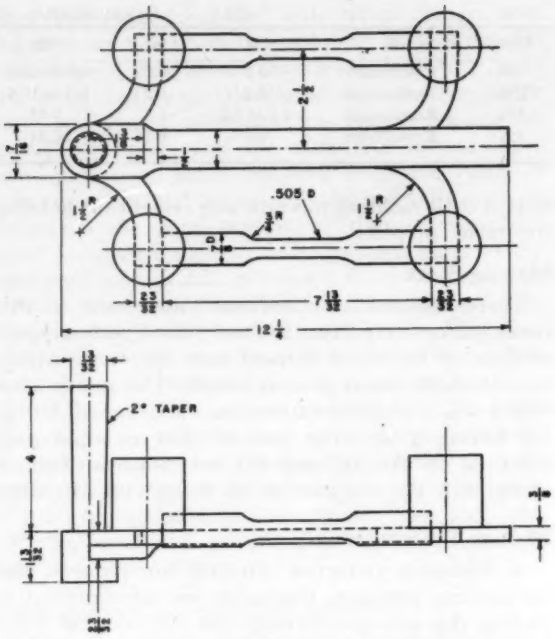


Fig. 5 — Dimensions of tensile-bar shell mold.

This technique was applied to molds poured horizontally and also inclined 2 and 4 degrees from the horizontal. Radiographs of the tensile bars are shown in Fig. 3. A mold inclined 2 degrees or more caused the metal streams to meet outside the gage length.

Adjustments were made in the gating arrangements to make the metal stream meet outside the gage length when the mold is poured level. In one case the main runner was decreased from the original depth of 1/2 in. to a depth of 3/8 in. In another case the main runner was decreased from the original depth of 1/2 in. to a depth of 3/8 in. and the two far ingates were decreased from the original depth of 3/8 in. to a depth of 1/4 in. Radiographs of the tensile bars are shown in Fig. 4. The final dimensions decided upon are shown in Fig. 5 and Fig. 6 shows a casting produced using these dimensions. The only change in the original dimensions developed in the Appendix is the decrease in the main runner depth of 1/8 in. This design allows the molds to be poured in a level position with the metal streams meeting outside the gage length.

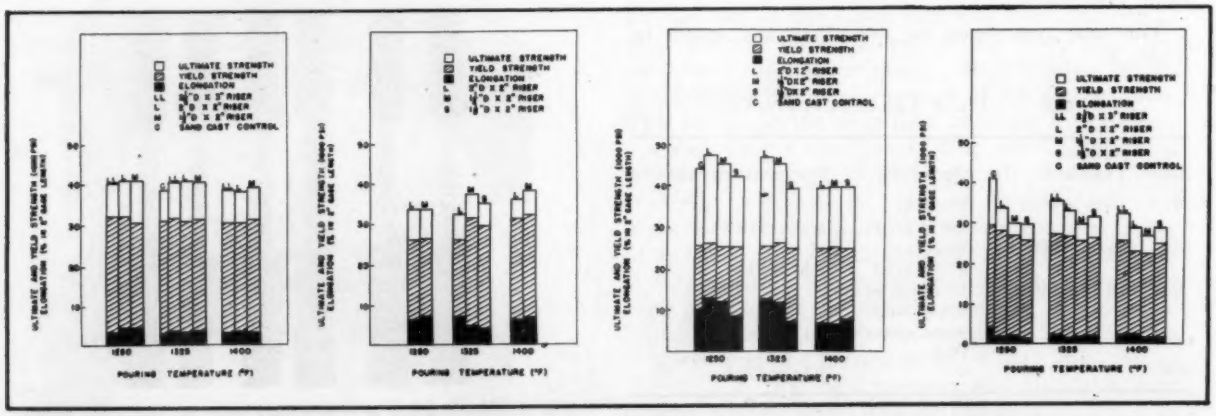


Fig. 7 — Effect of riser size and pouring temperature on properties of shell-mold cast-to-size 356 alloy tensile bars.

Fig. 8 — Effect of riser size and pouring temperature on mechanical properties of shell-mold cast-to-size 40E alloy.

Fig. 9 — Effect of riser size and pouring temperature on mechanical properties of shell-mold cast-to-size 220 alloy.

Fig. 10 —Effect of riser size and pouring temperature on mechanical properties of shell-mold cast-to-size 195 alloy.





Fig. 11 (Left) — Green-sand mold used as a control.
Fig. 12 (Above) — Tensile-bar molds showing 1⅛, 1½, and 2-in. diameter risers.

Mechanical Properties

The mechanical properties of 356, 40E, 220, and 195 are plotted in Fig. 7 through 10. The bar graphs show the effect of riser size and pouring temperature. A green-sand control mold was poured with each alloy except 40E. These results are also included. Each value on the graphs represents the average of two tensile bars from one mold; Figure 11 shows the green sand mold gating arrangement and Fig. 12 shows three of the four riser sizes used in the shell mold. The green-sand mold had risers which were 2 in. in diameter by approximately 2½ in. high. The heat treatments used for each alloy have previously been listed in Table 3. Results of the reduced pressure solidification tests indicated that the measured gas content was less than 0.05 cc/100 gms for each heat. This is generally regarded as a thoroughly degassed metal.

In general, the properties of bars cast in the shell mold were as high or slightly higher than the sand mold control for 356, 40E, and 220 alloys. The tensile properties of shell-molded 195 alloy, however, were lower than the sand cast controls. In order to determine if the low values obtained with shell-cast 195 alloy were due to the gating, a sand mold was prepared using the shell-mold gating system. Castings produced in this mold developed mechanical properties (T-6) of 39,600 psi tensile strength, 30,500 psi yield strength, and 2.7 per cent elongation. These values are of the same order of magnitude as was obtained with the sand-cast control bars.

The photomicrograph shown in Fig. 13 illustrates a typical defect found in the shell-cast tensile bars of 195 alloy. It is believed that the appearance of these intergranular cracks may be associated with the hot tearing tendencies of this alloy discussed later in this report.



Fig. 13 — Photomicrograph of shell-mold 195 alloy tensile bar showing typical cracks. Unetched. Mag. — 100X.

Temperature Gradient Measurements

Cooling curves were obtained by inserting thermocouples in the shell mold at positions along the tensile bar and within the risers at each end. After pouring, the temperature at each location within the mold was recorded on a multipoint recorder. From the cooling curves the temperature gradients within the mold were obtained. All temperature gradient studies were made using 195 alloy and a pouring temperature of 1325 F.

These results did now show a pronounced effect of riser size on the temperature gradients within the shell mold. Increasing riser size, however, did produce a significant increase in the time required for solidification of the casting. The time required for the shoulder of the tensile specimen to reach the solidus temperature (1020 F) was 5½ minutes for the large riser and only 2 minutes for the small riser. By comparison, the time required for the shoulder of the tensile bar in the sand mold to reach the solidus temperature was 2¼ minutes. Temperature gradients in the sand mold were considerably higher than in the shell mold having the same riser size. In addition to producing higher temperature gradients, the sand mold also develops the gradients much

Table 4 — Thermal Conditions Within Tensile Bar Mold After Various Locations Have Reached the Solidus Temperature

Type of Mold	Temperature at Locations Along the Tensile Bar (°F) Center	Shoulder	Far Riser	Near Riser	Time Required to Reach Solidus (1020 F) (min)
Sand, standard gating	1020	1150	1180	1180	½
Sand, shell gating	1020	1100	1160	1170	1
Shell, large riser	1020	1070	1130	1160	1½
Shell, small riser	1020	1050	1100	1130	1½
Sand, standard gating	—	1020	1140	1160	2¼
Sand, shell gating	—	1020	1130	1140	2¼
Shell, large riser	—	1020	1050	1080	5½
Shell, small riser	—	1020	1060	1100	2
Sand, standard gating	—	—	1020	1050	5¾
Sand, shell gating	—	—	1020	1050	4¾
Shell, large riser	—	—	1020	1060	6¼
Shell, small riser	—	—	1020	1080	2¼
Sand, standard gating	—	—	—	1020	6¾
Sand, shell gating	—	—	—	1020	5¼
Shell, large riser	—	—	—	1020	8
Shell, small riser	—	—	—	1020	3½

sooner after pouring presumably as a result of the higher initial chilling properties of the sand mold. Table 4 shows the thermal conditions existing within the tensile bar after various locations have reached the solidus temperature.

Discussion

The data obtained in this study show that by proper proportioning of all the elements of the gating system it is possible to control the location of the junction of the two streams of metal entering the test-bar mold cavity. This has been done without resorting to choking, tilting, or directing the whole metal stream through the test bar. Further work will naturally be required to make a statistical comparison with results obtained from standard bars which are in use today.

While the thermal gradient studies have shown that more favorable feeding conditions exist in the sand mold of standard design, the mechanical properties of the shell-mold castings compare favorably with those of the sand-mold castings. This may be due in part to the superior surface obtainable in the shell mold and to the fact that these molds are free from moisture. Elimination of the metal junction from the gage length section of the test bar, however, should be an important factor promoting uniformity in the tensile data. The fact that the proposed gating system does not adversely affect the thermal gradients developed in the casting is demonstrated by the measurements made in a sand mold using the new design. These data, shown in Table 4, reveal that substantially the same thermal gradients are developed in metal poured into molds of both new and standard design.

The effect that riser diameter has on tensile properties cannot be definitely established from this work. The optimum riser diameter appears to depend on the alloy being used and the pouring temperature.

The lower properties obtained in the shell mold with 195 alloy may be due to the hot tearing tendency of the alloy. When aluminum is cast into a shell mold, insufficient heat is available to properly decompose the resin binder. The strength of the shell therefore may introduce sufficient constraint to degrade the mechanical properties of an alloy which is susceptible to hot tearing. The defect shown in the photomicrograph (Fig. 13) may be a result of stresses developed in the casting while a liquid phase was still present. The fact that satisfactory properties were obtained when this alloy was cast into a sand mold with the identical gating design supports this theory. On the basis of these results it would indicate that due caution should be exercised in selecting an aluminum alloy for a shell mold casting. Those alloys which are prone to hot tearing should be avoided wherever possible in choosing a composition for a shell-mold casting.

Experience with 356 alloy, both in shell molds and sand molds, has indicated that while tensile and yield strengths exceeded typical values it was difficult to obtain elongation values of over 2 per cent. By controlling the iron and calcium contents at a low level, elongation values of 3 to 4 per cent were obtained while yield and tensile strength could be maintained at 30,000 and 40,000 psi, respectively.

Conclusions

It is concluded that:

1. Pattern equipment can be designed for a shell mold to accomplish the following:
   a. Cause the metal stream entering from opposite ends of the tensile-bar mold cavity to join at the grip end.
   b. Create thermal gradients along the axis of the tensile bar which are essentially symmetrical from the center of the gage length (thermal gradients in green-sand molds using this mold design are about the same as those obtained when using the "standard" design).
2. Riser size does not significantly affect thermal gradients developed in the gage length area of the shell-cast tensile bar. However, riser size does have a pronounced effect on solidification time.
3. The tensile properties of shell-molded 195 alloy tensile bars are significantly lower than green-sand properties. This decrease in properties may be due to the hot tearing tendencies of the alloy.
4. The tensile properties of shell-molded 356, 220, and 40E alloys are equal to or slightly better than the properties obtained in green sand molds.

Acknowledgment

The authors wish to thank the personnel of Pitman-Dunn Laboratories for the various services they performed in connection with this work.

APPENDIX

Calculation of Tensile Bar Mold Dimensions

The dimensions are calculated to give a controlled velocity of the metal in the system, using a ratio of 1:4:4 for the relationship of sprue, runner, and ingate cross-sectional areas.

The velocity of a fluid at the base of the sprue is expressed:

$$V = \sqrt{2gh/\beta} \qquad (1)$$

where:

V = Velocity
g = acceleration of gravity
h = total height of sprue (sprue + pouring basin)
β = friction factor = 1.3 for circular sprue

Therefore:

$$V = \sqrt{\frac{2 \times 384 \times 6}{1.3}} = 59.6 \text{ in./sec}$$

The area at the base of the sprue (A_B) is expressed:

$$A_B = \frac{W}{\omega V} \qquad (2)$$

where:

W = flow rate = 1.25 lb/sec from previous experience
ω = sp wt of aluminum
A_B = area at base of sprue

A flow rate of 1.25 lb/sec is about as fast as the metal can be poured conveniently.

Therefore:

$$A_B = \frac{1.25}{0.086 \times 59.6} = 0.244 \text{ in.}^2$$

The diameter at the base of the sprue is $^9/_{16}$ in.

For this mold design the area of the well was made equal to five times the area at the base of the sprue.

Therefore:

$$A_W = 5 \times A_B \quad (3)$$
$$A_W = 5 \times 0.244$$

The diameter of the well is $1^7/_{16}$ in.

The area of the first runner is expressed:

$$R_1 = 4 \times A_B \quad (4)$$

The runner is twice as wide as it is deep. Therefore the dimensions of the first runner are $1^{13}/_{32}$ in. and $^{23}/_{32}$ in.

The depth of the well is twice the depth of the first runner. The depth of the well is $1^{13}/_{32}$ in.

The area at the top of the sprue is expressed:

$$A_T = A_B \sqrt{\frac{h}{\text{well depth}}} \quad (5)$$

Therefore, the diameter of the top of the sprue is $1^3/_{16}$ in.

$$A_G = 4 A_B \quad (6)$$

Each ingate is twice as wide as it is deep and 5 per cent is added to each gate area to allow for friction. The dimensions of the ingates are $^3/_8$ and $^{23}/_{32}$ in.

The runner between the set of ingates is decreased in area by an amount equal to the first two ingates. Then 5 per cent is added to the first two ingates. Then 5 per cent is added to the area for friction. The runners dimensions are then 1 and $^1/_2$ in. The runner beyond the second set of ingates has the same area.

DISCUSSION

Chairman: E. V. Blackmun, Aluminum Company of America, Pittsburgh, Pa.

Co-Chairman: F. L. Turk, Joy Mfg. Co., New Philadelphia, Ohio.

D. L. Colwell[1]: What was the thickness of the shell and what was the backup material? The backup material could affect the cooling rate in comparison with green sand mold.

Mr. Harris: The thickness of the shell used throughout this work was about $^1/_4$ in. It was recognized that shell thickness could affect cooling rates and for that reason the "dwell-time" during the molding cycle was carefully controlled to yield uniform shell thickness from mold to mold. No backup material was used for the molds.

1. Apex Smelting Co., Cleveland.

TENSILE PROPERTIES OF ALUMINUM-SILICON-MAGNESIUM ALLOYS AND THE EFFECTS OF SODIUM MODIFICATION

By

R. C. Harris, S. Lipson and H. Rosenthal*

ABSTRACT

A systematic study has been made of some mechanical properties of aluminum-silicon-magnesium alloys centering around the composition range of the 356 alloy. Particular attention was directed to the influence of sodium modification on those properties. The effects of the alloying elements and the modification process were evaluated by means of age hardening characteristics and tensile properties. Results indicate that, within the range of silicon contents studied, there is a pronounced sensitivity to magnesium content. Modification effects were not observed to any extent in alloys containing magnesium in excess of 0.2 per cent. Below this amount modification appears to be beneficial to the tensile properties, particularly the elongation. Studies of an exploratory nature indicate that if room temperature aging occurs prior to the elevated temperature aging treatment, lower tensile and yield strengths may result.

Introduction

This investigation was originally initiated for the purpose of studying the effects of sodium modification on aluminum-silicon magnesium alloys of approximately the composition of commercial 356 alloy. Since the modification process is known to be beneficial to binary alloys of aluminum and silicon it was anticipated that these same beneficial effects might be realized in the commercial 356 alloy which contains magnesium as an additional alloying element. Of particular importance would be any improvement in elongation values since this alloy, like many of the high-strength aluminum casting alloys, exhibits inherently low ductility.

As a preliminary to the experimental work, a literature survey was made to obtain information pertaining to sodium modification of aluminum-silicon alloys, and to obtain data on the properties of aluminum-silicon-magnesium-alloys. It was intended that the published data concerning the ternary alloys would serve as a basis for evaluating the effects of modification. No detailed information was found in the literature on the ternary alloys, however, and it was necessary to include in the investigation, a study of the variables associated not only with the modification process, but also those associated with the alloys.

In view of the fact that 356 alloy is used to a considerable extent at the present time, it would appear that a study of the variables concerned with the alloy would contribute to a better understanding of the alloy and thus permit better control of mechanical properties.

The scope of this project, therefore, has not been limited to the effects of modification alone, but has been expanded to include a study of aluminum-silicon-magnesium alloys centering around the 356 alloy composition.

Results of Literature Survey

Pacz,[1] in 1921, learned that the addition of a small quantity of alkaline fluoride to the surface of molten aluminum-silicon alloys would promote substantial improvement in the mechanical properties. The cause was ultimately narrowed down to the presence of sodium in the melt. Since this time, investigations have been carried out to determine what other elements might produce modification effects. The use of cadmium, bismuth,[2] and antimony[3] have been reported to produce modification to some degree but there is considerable disagreement as to the effectiveness of these elements. Otani[4] used sodium hydroxide, sodium and potassium carbonate, sodium and potassium chloride and borax but found that with each of these, modification was difficult to achieve. More recently, Mascre[15] has reported that phosphorus will produce modification in high silicon alloys.

The two most effective means of producing modification in aluminum-silicon alloys appear to be through the use of metallic sodium or a sodium rich flux. Several methods of introducing each of these into the melt have been suggested.[5,6,7] There are two fundamental differences between using a flux and using metallic sodium. When a flux is employed the sodium content of the melt increases as the holding time between the addition and pouring increases. On the other hand, increased holding times reduce the sodium content of the melt when metallic sodium is used.[7]

Due to the extremely small amounts of sodium involved, an analysis of any type is subject to consider-

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able uncertainty. Fairlie and Brook[8] report a wet method of chemical analysis which was used with some success. Thall and Chalmers[9] reported a semi-quantitive method in which samples are subjected to neutron bombardment. By measuring the rate of radioactive decay the sodium content can be approximated.

Sodium modification of aluminum-silicon alloys produces a marked change in the microstructure. Under equilibrium conditions the microstructure of a hypoeutectic alloy would consist, at room temperature, of primary aluminum dentrites plus the eutectic. The silicon in the eutectic normally appears in the form of needles or platelets. Similarly, a hypereutectic alloy would consist of primary silicon plus the eutectic. When sodium is added to these alloys there is a change in the appearance of the silicon. The microstructure of the hypoeutectic alloy again contains the primary aluminum dendrites but the eutectic silicon is in the form of globules. Slightly hypereutectic alloys, up to about 14 per cent silicon, also exhibit a similar structure indicating a shift in the eutectic composition. Observations by Archer and Kempf,[5,10] Gaylor,[11] and Gwyer,[12] have shown that additions of sodium to these alloys produce the same changes in microstructure that undercooling does.

As a result of the considerable amount of experimental work done on the subject of modified aluminum-silicon alloys, several theories to explain the mechanism of modification have been developed.[4,5,13,14,17,18] Thall and Chalmers suggest that sodium additions to aluminum-silicon alloys change the relative interfacial energies between the alpha aluminum, silicon and the melt. Mascre[15] suggested that sodium affects the viscosity of the liquid and, thus, the rate of deposition of silicon atoms. Along similar lines Kondic and Kozlunski[16] reported that sodium additions decrease the casting fluidity of an aluminum-silicon eutectic alloy.

Methods and Materials

Melting Stock

Alloys investigated in this program fall into two general classifications. These two classifications are: (a) alloys within the compositional limits of 356 alloy, (b) experimental alloys of the Al-Mg-Si type.

The melting stock used in the first classification consisted of commercial 356 alloy. Pre-alloyed ingots obtained from commercial sources were first melted in a gas-fired tilting furnace and poured off into small ingots weighing 2-3 pounds each. Each heat prepared in this manner weighed approximately 100-120 pounds.

Melting stock used in the second classification of alloys consisted of commercial aluminum-50 per cent silicon alloy, aluminum-65 per cent magnesium alloy and aluminum ingot of commercial purity. All heats prepared using these materials weighed approximately 30 pounds and were melted using induction equipment; 2-3 pound ingots were made from each heat.

Chemical analyses were made on two ingots from each heat. Table 1 shows a typical analysis obtained from alloys within the compositional limits of 356 alloy. A range of values represents the maximum and minimum values obtained for that element throughout this work. Synthetic alloys produced in the laboratory which contained silicon and magnesium in quantities which were outside specification limits of 356 alloy had impurities whch were not greater than those shown in Table 1.

Melting Procedures

A 3,000-cycle induction lift coil was used for melting purposes in the actual experimental work. Degassing procedures, where employed, were accomplished by bubbling dry nitrogen or chlorine through the melt. The efficiency of the degassing treatment was determined by means of a reduced pressure solidification test. A small test specimen weighing approximately 50 grams was taken from the melt using a thin steel sampling cup. This was placed in a vacuum chamber and the chamber evacuated to approximately 100 mm Hg. Three minutes were allowed for solidification of each sample. Density measurements were made on each sample using a gravitometer.[19]

In most cases, a density sample was obtained immediately before pouring in order to determine the gas content of the melt at the time of pouring.

Temperature measurements were made using a portable Leeds and Northrup potentiometer and a chromel-alumel thermocouple encased in a stainless steel tube.

Molding Practice

Tensile bar molds were made using a synthetic green sand having a composition as shown in Table 2. The gating and risering arrangement is shown in Fig. 1. This arrangement conforms essentially to the standard test bar design given in Fig. 11a of the Federal Specification QQ-A-151a, with the exception of the horn gate, which was used to produce a bottom gating effect.

Table 1 — Chemical Analysis of 356 Alloy Used

	Element, per cent									
	Cu	Si	Fe	Mn	Zn	Mg	Ti	Al	Pb	Sn
Fed. Spec. QQ-A-601a Class 3	0.2 max.	6.5-7.5	0.5 max.	0.3 max.	0.3 max.	0.2-0.4	0.2 max.	R	0.05 max.	0.05 max.
Experimental Stock	0.05-0.10	6.5-7.2	.23-.33	<0.01	<0.10	.28-.35	<0.14	R	<0.01	<0.01

< — less than R — remainder

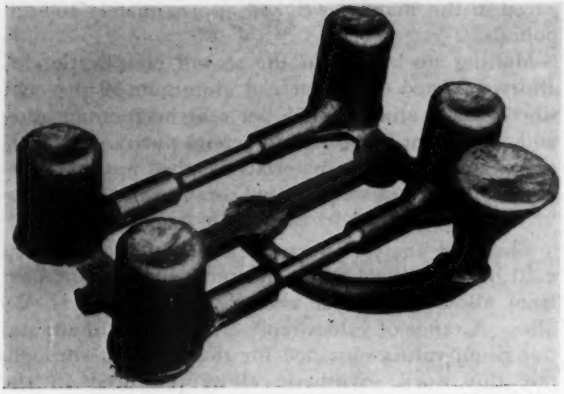


Fig. 1 — Gating arrangement of 0.505-in. tensile specimen.

TABLE 2 — COMPOSITION OF MOLDING SAND

Component	Per Cent
Silica Sand (AFS Grain Fineness No. 80)	94.25
Southern Bentonite	1.5
Western Bentonite	1.5
Cereal	0.25
Water	2.5

Modification Procedures

Since metallic sodium will oxidize rapidly upon exposure to air, it was necessary to adopt a method of sodium addition which would minimize the time of contact of the sodium with air. This was accomplished by the use of a steel rod drilled as shown in Fig. 2.

The sodium used in the experimental work was stored under kerosene. To obtain a sodium sample the steel rod was pushed through a block of sodium while still under the kerosene. Sodium was forced into the large hole at the end of the steel rod and at the same time the kerosene escaped through the small hole on the side. In this manner it was possible to determine when the tube was completely filled by the slight extrusion of sodium through the small hole. The rod was then wiped free of kerosene and the end wrapped in aluminum foil. The sample was used as soon as possible, usually within 30 minutes.

To introduce the sodium, the tube was pushed below the surface of the melt and held several inches from the bottom of the crucible. After a period of 5 to 10 seconds, burning sodium would appear on the surface. At this time, the sodium on the surface was gently stirred into the melt using the sample tube. The temperature of the melt for all sodium additions was maintained between 1250-1300 F.

Several experiments were conducted using a sodium-rich flux in place of metallic sodium. Although these fluxes produced a modified structure equally as well as metallic sodium, some difficulties were encountered in preventing the molten flux from entering the mold and producing inclusions. The direct addition of metallic sodium was finally adopted because it permitted easy variation of sodium content while avoiding flux inclusion difficulties.

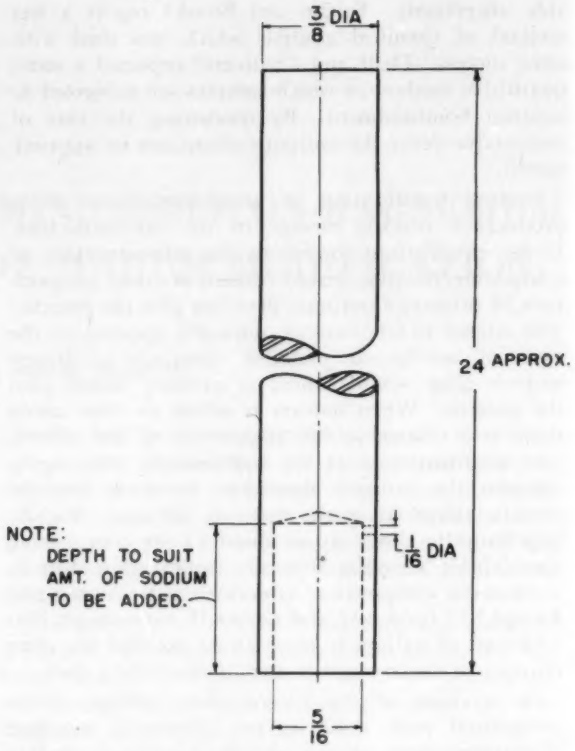


Fig. 2 — Sketch of steel tube used for sodium additions.

Results

Modification

Amount of Sodium and Holding Time — An investigation was made of the interrelated effects of the quantity of sodium added to the melt and the holding time after the addition and prior to pouring.

Sand molds were prepared which provided for pouring cylindrical bars ½-inch in diameter by approximately 7 inches long. Three melts of 356 alloy containing 0.28-0.30 per cent magnesium and weighing approximately 22 lb each were prepared, to which were added 0.02, 0.03 and 0.05 per cent sodium, respectively. The charge was first melted and heated to 1250-1300 F, degassed with chlorine for five minutes and a control sample poured. The sodium was introduced and a second sample poured immediately. Eight additional samples were subsequently poured at 5-minute intervals. The melt temperature was maintained between 1250-1300 F throughout the run.

Samples were cut from each of the bars for metallographic examination. The test bar results from this series are shown in Table 3 and photomicrographs of several samples are shown in Figs. 3 and 4.

Figure 3a is a typical photomicrograph of the control sample and shows the characteristic normal structure with the silicon constituent in the form of needles or platelets.

The microstructure shown in Fig. 3b is that of a sample poured 5 minutes after the addition of 0.02 per cent sodium, and shows the silicon constituent in the form of needles or platelets which are smaller

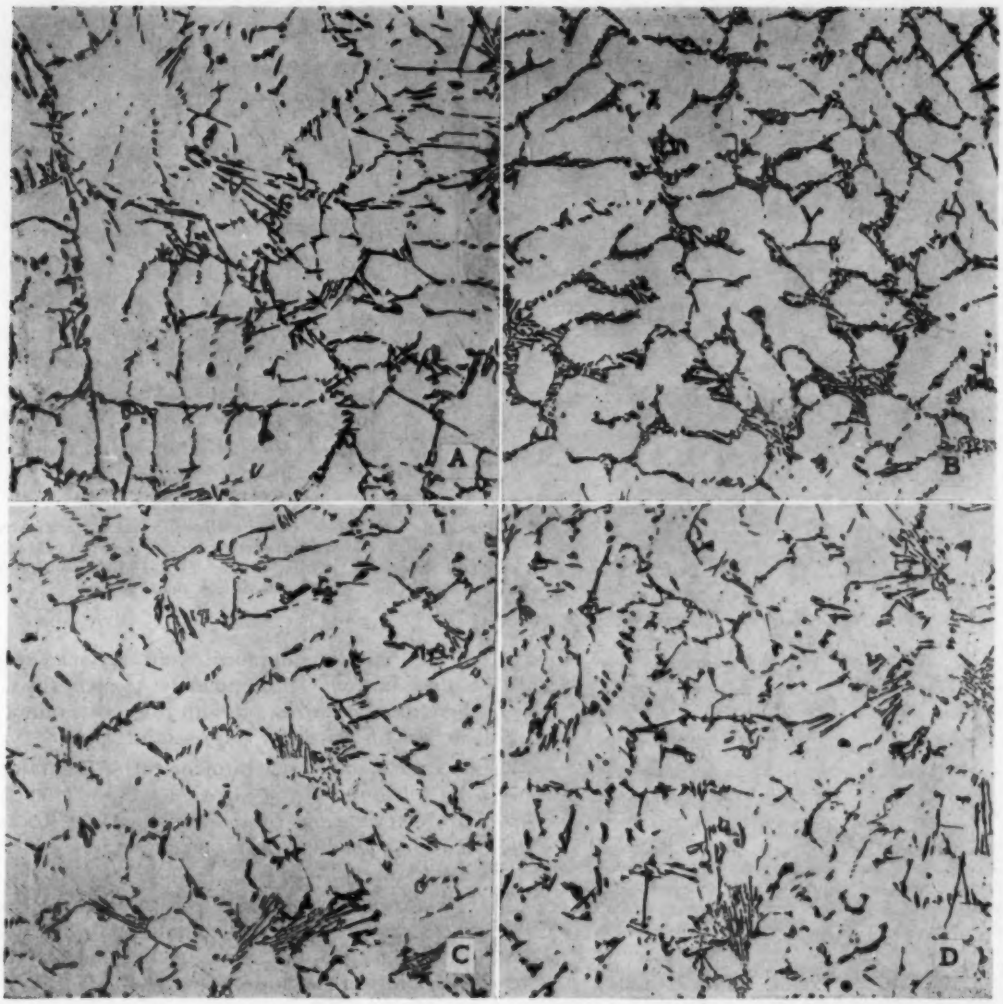


Fig. 3 — Effect of holding time on as-cast microstructure of 356 alloy. Modified with 0.02% sodium. Mag. — 100X. (A) control sample, modified; (B) 5-minute holding time; (C) 10-minute holding time; (D) 15-minute holding time.

than those in the control sample but are not quite globular. This type of structure is arbitrarily referred to as a semi-modified structure. Subsequent samples in this series show a completely normal structure similar to the control sample. The transition from the modified to the normal structure appears to be gradual and is not a sudden change from one form to the other. When 0.05 per cent sodium is added modification effects last for approximately 15 minutes as shown in Fig. 4.

TABLE 3 — DESCRIPTION OF MICROSTRUCTURE FOR AMOUNTS OF SODIUM AND HOLDING TIMES AS SHOWN (356 ALLOY CONTAINING 0.28-0.30% MG)

Holding Time, (Minutes)	Type of microstructure for percentages of sodium additions			
	0	0.02	0.03	0.05
0	normal	modified	modified	modified
5	normal	semi-modified	modified	modified
10	normal	normal	modified	modified
15	normal	normal	semi-modified	modified
20	normal	normal	normal	semi-modified
25	normal	normal	normal	normal

Relationship of Tensile Properties to Microstructure — During the holding time employed after the sodium addition, there is a gradual transition of the form of the silicon constituent from the globular to the plate-like structure. If sodium modification is beneficial to the tensile properties of 356 alloy it is presumably as a result of the formation of globular silicon particles. Therefore, the degree of improvement should correlate to the extent that the microstructure shows modification effects.

Microstructures with varying degrees of modification were obtained by adding 0.01 per cent sodium to a melt and pouring tensile bar molds after holding periods of 0, 5, 10, 15 and 20 minutes. The tensile properties were determined and compared to the microstructures of the individual bars.

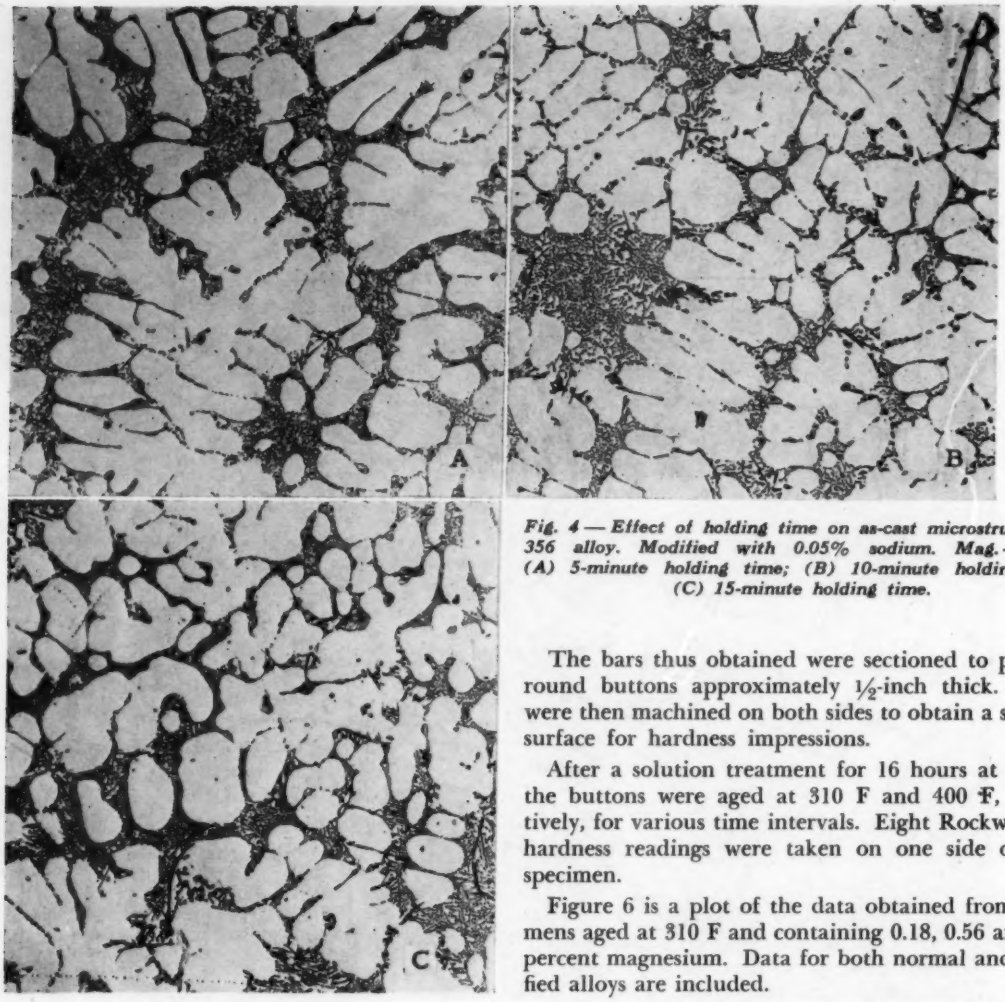


Fig. 4 — Effect of holding time on as-cast microstructure of 356 alloy. Modified with 0.05% sodium. Mag. — 100X. (A) 5-minute holding time; (B) 10-minute holding time; (C) 15-minute holding time.

From the photomicrographs in Fig. 5 it is again apparent that there is a gradual reversion of the modified structure to the normal or platelike structure. This change begins after 10 minutes and is complete at the end of 15 minutes. The tensile properties of 356 alloy containing 0.30 per cent magnesium, which are associated with each microstructure are given in Table 4. These results indicate that the form of the silicon has no significant effect on the tensile properties of this alloy. The tensile properties remain substantially unchanged throughout the entire holding time.

Effect of Magnesium

Age Hardening Studies — To study the effect of magnesium on the age-hardening characteristics of aluminum-silicon-magnesium alloys and to determine whether modification alters the characteristics, normal and modified melts were prepared containing from 0.1 to 1.0 per cent magnesium and approximately 7 per cent silicon. Cylindrical bars ¾-inch in diameter and 7 inches long were cast in green sand molds from these melts. The melts were modified by adding 0.05 per cent sodium and holding 5 minutes before pouring.

The bars thus obtained were sectioned to produce round buttons approximately ½-inch thick. These were then machined on both sides to obtain a suitable surface for hardness impressions.

After a solution treatment for 16 hours at 1000 F the buttons were aged at 310 F and 400 F, respectively, for various time intervals. Eight Rockwell "F" hardness readings were taken on one side of each specimen.

Figure 6 is a plot of the data obtained from specimens aged at 310 F and containing 0.18, 0.56 and 1.07 percent magnesium. Data for both normal and modified alloys are included.

A similar plot is shown in Fig. 7 for an aging temperature of 400 F and magnesium contents varying from 0.18 to 1.07 per cent as shown. For all magnesium contents the maximum hardness occurs between one and two hours at this aging temperature.

For both aging temperatures investigated, the modified alloys age to a lower hardness level than the normal alloys. The difference in hardness level decreases with increasing magnesium content. How-

TABLE 4 — EFFECT OF HOLDING TIME OF THE TENSILE PROPERTIES OF 356 ALLOY (0.30 PER CENT MG) IN THE T6 CONDITION.* (0.01 PER CENT SODIUM ADDED)

Holding Time, (minutes)	Yield Strength 0.2% Offset (psi)	Tensile Strength (psi)	Elongation %
0	30,400	37,100	2.4
5	29,900	39,500	5.1
10	30,500	38,900	2.9
15	30,000	38,200	3.5
20	30,100	39,100	3.8

*All tensile bars solution treated for 16 hours at 1000 F and aged for 4 hours at 310 F. Each value is an average of three bars.

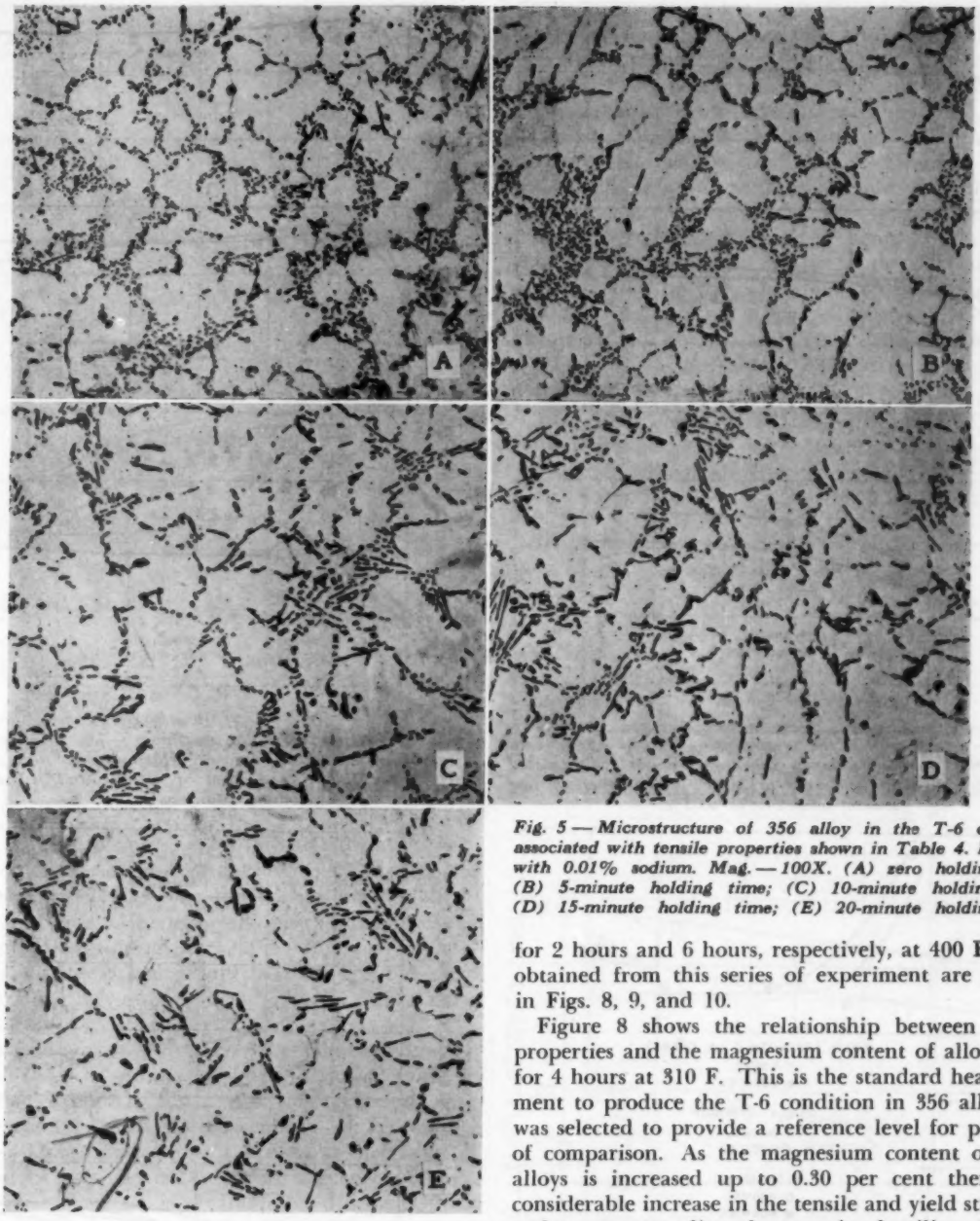


Fig. 5 — Microstructure of 356 alloy in the T-6 condition associated with tensile properties shown in Table 4. Modified with 0.01% sodium. Mag. — 100X. (A) zero holding time; (B) 5-minute holding time; (C) 10-minute holding time; (D) 15-minute holding time; (E) 20-minute holding time.

ever, both the normal and modified alloys reach a maximum hardness after the same time interval.

Effect of Magnesium on Tensile Properties — Additional work was done to determine the effect of magnesium content on the tensile properties of normal and modified melts containing approximately 7 per cent silicon and from 0.1 to 0.6 per cent magnesium. Modified melts were prepared by the addition of 0.03 per cent sodium and a 5-minute holding time. Prior to pouring and modification all melts were degassed for 15 minutes using dry nitrogen.

The tensile bars thus obtained were then solution treated for 16 hours at 1000 F and subsequently divided into three groups. One group was aged for 4 hours at 310 F and the remaining two groups aged for 2 hours and 6 hours, respectively, at 400 F. Data obtained from this series of experiment are plotted in Figs. 8, 9, and 10.

Figure 8 shows the relationship between tensile properties and the magnesium content of alloys aged for 4 hours at 310 F. This is the standard heat treatment to produce the T-6 condition in 356 alloy and was selected to provide a reference level for purposes of comparison. As the magnesium content of these alloys is increased up to 0.30 per cent there is a considerable increase in the tensile and yield strengths and a corresponding decrease in ductility. Within this range the yield strength of normal alloys increases from 17,000 psi at 0.13 per cent magnesium to 29,000 psi at 0.30 per cent. The tensile strength shows a similar increase from 29,000 to 37,000 psi within the same range of magnesium content. Elongation values show a decrease from 7 to 3 per cent.

By increasing the aging temperature to 400 F the tensile and yield strengths can be significantly improved but at the expense of the ductility, as shown in Fig. 9. When aged for 2 hours at 400 F, the tensile strengths of normal alloys were 32,000 psi at 0.15 per cent magnesium and 42,500 at a magnesium level of 0.3 per cent. There was a similar trend for the yield strength. Elongation values are considerably lower after aging at this temperature.

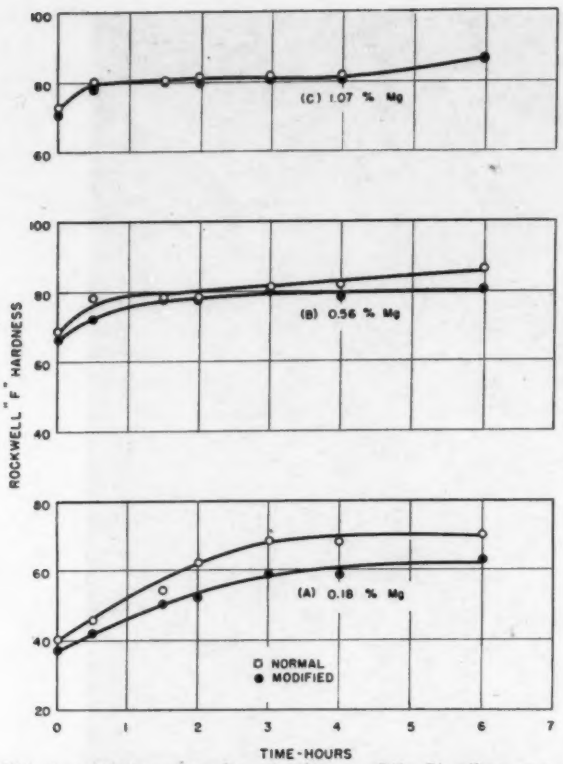


Fig. 6 — Aging curves for aluminum — 7% Si alloys containing Mg in amounts shown. All alloys were solution treated for 16 hr at 1000 F and aged at 310 F. Modified alloys had sodium additions of 0.05%.

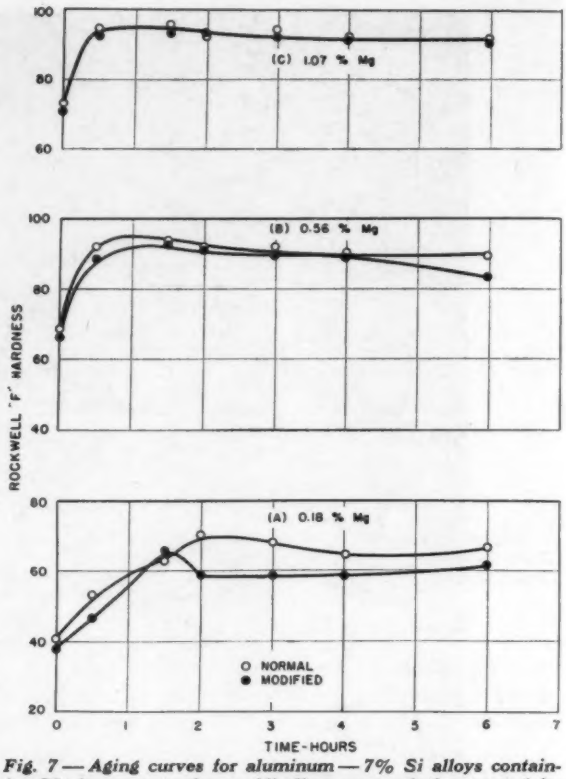


Fig. 7 — Aging curves for aluminum — 7% Si alloys containing Mg in amounts shown. All alloys were solution treated for 16 hr at 1000 F and aged at 400 F. Modified alloys had sodium additions of 0.05%.

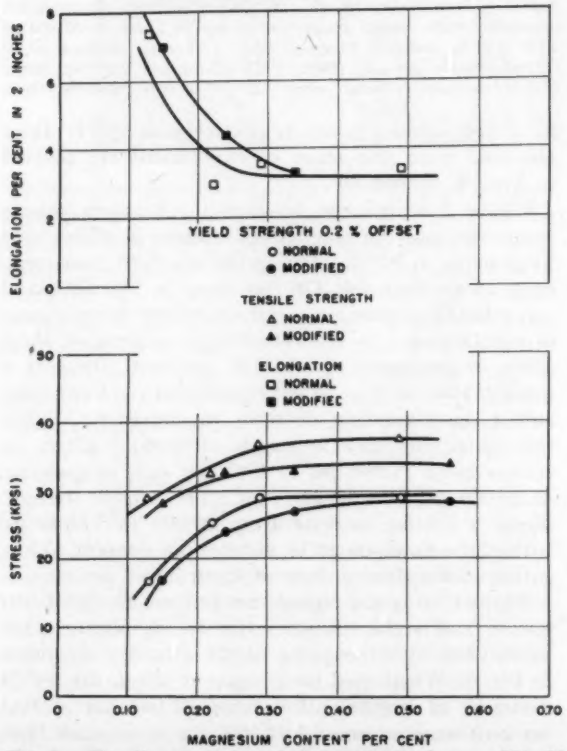


Fig. 8 — Tensile properties of degassed aluminum — 7% Si alloys containing various amounts of magnesium. All alloys were solution treated for 16 hr at 1000 F and aged for 4 hr at 310 F. Modified alloys had a sodium addition of 0.03%.

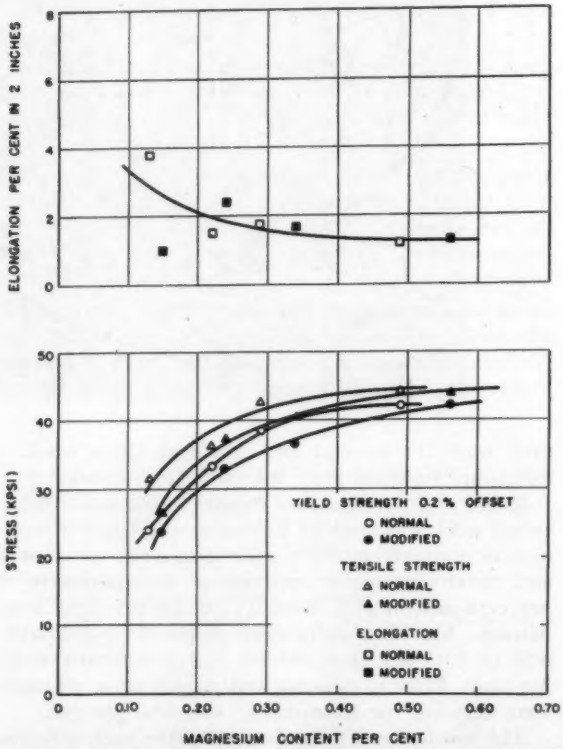


Fig. 9 — Tensile properties of degassed aluminum — 7% Si alloys containing various amounts of magnesium. All alloys were solution treated for 16 hr at 1000 F and aged for 2 hr at 400 F. Modified alloys had a sodium addition of 0.03%.

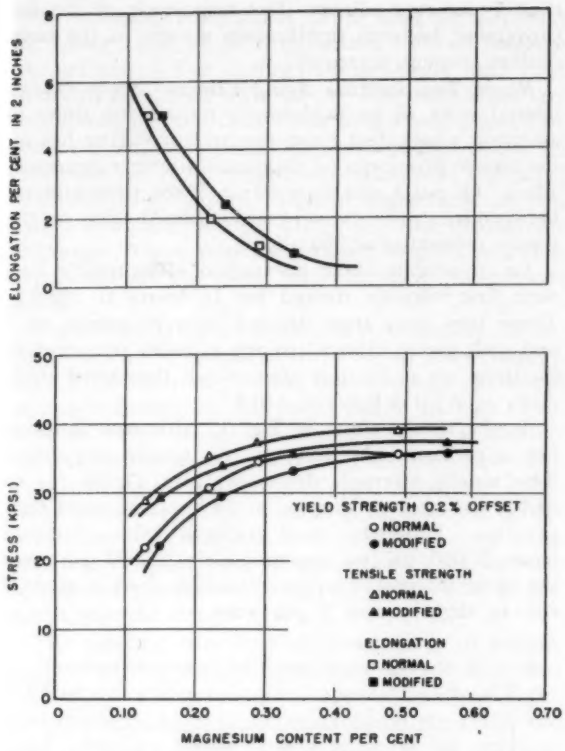


Fig. 10 — Tensile properties of degassed aluminum — 7% Si alloys containing various amounts of magnesium. All alloys were solution treated for 16 hr at 1000 F and aged for 6 hr at 400 F. Modified alloys had a sodium addition of 0.03%.

When aged for 6 hours at 400 F, the elongation is slightly improved and the tensile and yield strengths are lowered, particularly at the higher magnesium contents. Above 0.30 per cent magnesium these alloys, while exhibiting reasonably high strength properties, are quite brittle and have elongations of the order of 1 per cent or less. This is shown in Fig. 10.

For the entire range of magnesium contents studied, the modified alloys produced generally lower properties than the normal alloys with the exception of the elongation. Density measurements of reduced pressure solidification samples taken from each melt at the time of pouring, indicate that the gas content of the modified melts was greater than the melts to which no sodium had been added. The difference in the level of properties between normal and modified melts is not, therefore, directly attributable to microstructural changes. The fact that the modified melts have a higher gas content would indicate that as a result of the sodium addition, not only is the modified structure produced but additional hydrogen gas is introduced into the melt. This may be due to the introduction of kerosene along with the sodium even though all precautions were taken to remove the kerosene.

In order to critically evaluate the effects of modification, the tensile properties of specimens poured from melts having similar gas contents should be compared. In an effort to obtain data from which this comparison could be made several melts were intentionally gassed by "poling" with pieces of wood. Degassing procedures were then employed and tensile specimens poured after intervals of the degassing treatment. At the time of pouring the tensile specimens, samples were taken from the melt for the determination of the gas content by the reduced pressure solidification test. Figure 11 shows the results of these runs, with the tensile strength, yield strength and elongation plotted against the gas content of the melt at the time of pouring. There is a considerable amount of scatter in the data and it is apparent that no trend is established with respect to modified and normal alloys. There is, however, a general trend toward lowering the tensile and yield strengths as the gas content is increased, as would be expected.

Effect of Silicon — To study the effect of silicon content on the tensile properties of aluminum-silicon-magnesium alloys, alloys were prepared which contained from 5 to 13 per cent silicon and from 0.1 to

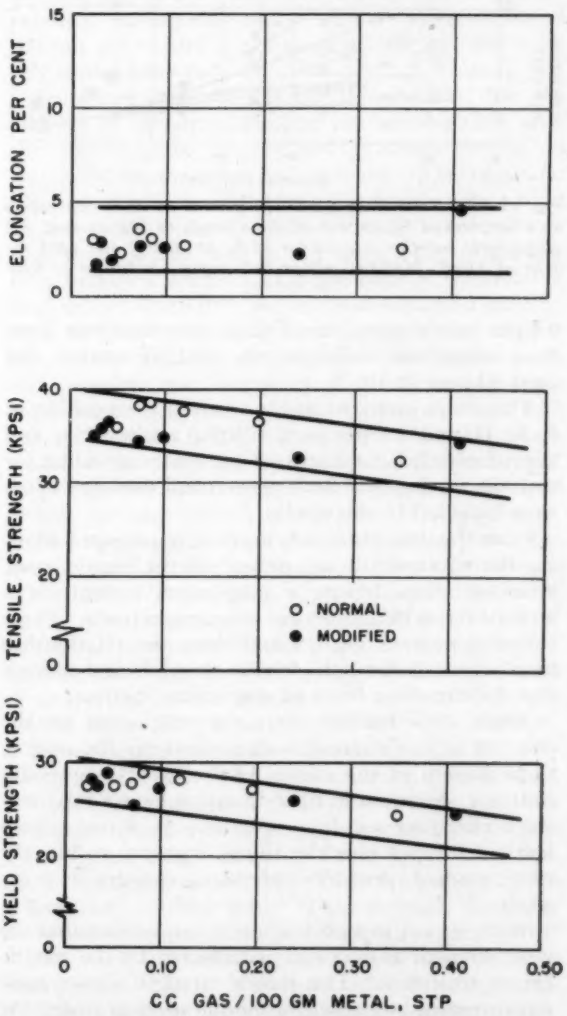


Fig. 11 — Comparison of tensile properties of 356 alloy in the normal and modified conditions at various levels of gas content. All tensile bars were solution treated for 16 hr at 1000 F and aged for 4 hr at 310 F. Modified alloys had sodium additions of 0.01%.

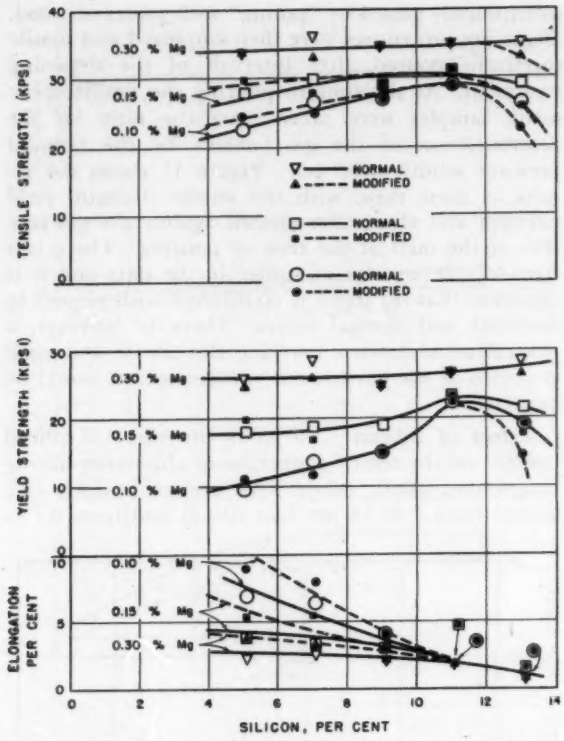


Fig. 12 — Variations of tensile properties of Al-Si-Mg alloys as a function of Si content at three levels of Mg content. All alloys were solution treated for 16 hr at 1000 F and aged for 4 hr at 310 F. Modified alloys had sodium additions of 0.01 to 0.05%.

0.3 per cent magnesium. Tensile bars were cast from these alloys and subsequently solution treated and aged 4 hours at 310 F.

The alloys prepared had a nominal composition of 5, 9, 11 and 13 per cent silicon, respectively, and approximately 0.1, 0.2 and 0.3 per cent magnesium for each silicon content. Both normal and modified alloys were included in this study.

From the data obtained curves were prepared showing the effect of silicon content on the tensile properties of alloys having a magnesium content at a level of 0.1, 0.15 and 0.3 per cent, respectively. These curves appear in Fig. 12 and show the relationship between yield strength, tensile strength and elongation for the three levels of magnesium content.

These data indicate that the magnitude of the effect of silicon content is dependent on the magnesium content of the alloy. At the lower magnesium contents, variations in silicon content have a relatively large effect on tensile properties. At a magnesium level of 0.3 per cent the silicon content, within the range studied, produces only small changes in properties.

With respect to modification, it can be seen that the yield strength is essentially unaffected by the modification treatment. The tensile strength shows some improvement as a result of modification provided the silicon content is below 7 per cent and the magnesium below approximately 0.15-0.20 per cent.

Elongation values show some improvement as a result of modification in all alloys containing less than 7 per cent silicon. The magnitude of the improvement becomes significantly greater as the magnesium content decreases.

Room Temperature Aging Effects – Some experimental work of an exploratory nature was done to ascertain what effect room temperature aging has on the tensile properties of aluminum-silicon magnesium alloys. Of particular importance is the time interval between solution treatment and artificial aging on the tensile properties of 356 alloy.

To investigate these effects 356 alloy tensile bars were first solution treated for 16 hours at 1000 F. These bars were then divided into 10 groups of 3 and each group allowed to age at room temperature for times up to 10 days after which they were artificially aged for 4 hours at 310 F.

As the results show in Fig. 13, this time interval has a pronounced effect on the tensile properties. The tensile strength decreases from 37,000 psi to 34,000 psi after an interval of 24 hours at room temperature. Similarly, yield strength values decrease from 31,000 psi to approximately 26,000 psi over the same interval. Elongation values show a gradual rise to slightly over 2 per cent.

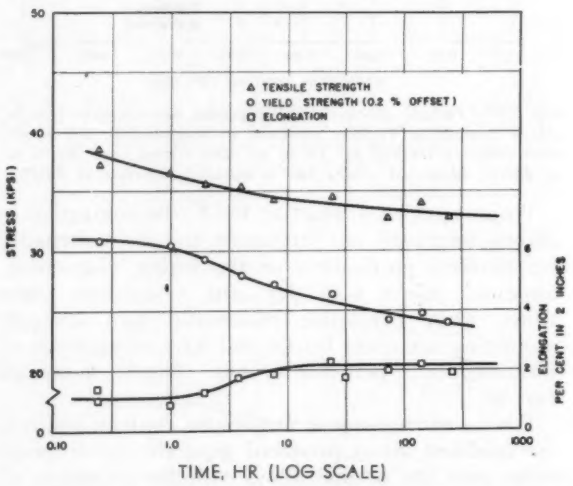


Fig. 13 — Variation of tensile properties of 356 alloy as a function of room temperature aging time. All alloys solution treated for 16 hr at 1000 F, aged at room temperature for the times shown and subsequently aged for 4 hr at 310 F.

Discussion

Effects of Magnesium

It is apparent from this work that as the magnesium content is varied in the 356 type alloy containing 7 per cent silicon, there is a wide range of tensile and yield strengths and a comparatively narrow range of elongation values which are attainable. In order to retain some ductility in the alloy it is necessary to confine the upper limit of the magnesium content to approximately 0.4 per cent. It would be desirable from the standpoint of the tensile and yield strengths not to have the magnesium vary much from that amount. However, in order to make processing in the foundry feasible a compositional range must be specified within practical limits.

By allowing a variation of at least 0.2 per cent the lower limit of the magnesium content becomes 0.2 per cent. The corresponding tensile and yield strengths associated with 0.2 per cent magnesium are considerably lower than those at 0.4 per cent. The results of this work have demonstrated that in the T6 condition the yield strength decreases from 29,000 psi to 23,000 psi within this range of magnesium content, with a minimum of 3 per cent elongation throughout the range. These values are subject to some variation depending on the time interval prior to artificial aging.

If artificial aging is carried out immediately following the solution treatment then, for an alloy containing 0.35 per cent magnesium, the yield strength would be approximately 31,000 psi as shown in Fig. 13 instead of 29,000 psi as shown in Fig. 8. It is obvious that, assuming the ductility of the alloy meets the requirements for a particular application, the maximum allowable design stresses must conform to the minimum properties obtained at 0.2 per cent magnesium even though the alloys containing 0.4 per cent magnesium possess yield and tensile strengths 25 per cent above the design figure.

The extreme sensitivity of these alloys to magnesium content could be attributed to the fact that magnesium additions up to approximately 0.3 per cent produce Mg_2Si in amounts which are below the solid solubility limits of this phase at the solution treating temperature. Within this range any small increase in magnesium content would tend to produce a correspondingly large increase in tensile and yield strength and decrease in ductility. Additions above 0.3 per cent magnesium produce Mg_2Si in amounts which approach the solid solubility limit and a small increase in magnesium would, in this case, have a relatively small effect on properties.

This severe sensitivity of 356 alloy to magnesium content tends to make processing in the foundry more difficult. Magnesium, having the higher vapor pressure and affinity for oxygen of the alloying constituents, would suffer the greatest loss as a result of any extended holding time in the melting furnace. Similarly, this element would be depleted most readily by any chlorine degassing treatment. Chlorine, therefore, is objectionable as a degassing agent, even though it is considered more effective than nitrogen unless provision is made to restore the magnesium to the original level.

Effect of Silicon

From the results obtained it is apparent that, in comparison to magnesium, variations in silicon content, produce relatively small changes in tensile properties. The magnitude of the effect is dependent on the amount of magnesium present in the alloy. As the magnesium content is increased up to 0.3 per cent the effect of the silicon content within the range 5-13 per cent becomes less pronounced, for this particular aging treatment studied. At a magnesium level of 0.3 per cent variations in silicon content within the limits specified for 356 alloy (6.5-7.5 per cent) have very little effect on the tensile and yield strengths and a slight effect on the elongation. For silicon contents below these limits there is still little change in the tensile and yield strengths but some increase in elongation.

An alloy containing 5 per cent silicon and 0.3 per cent magnesium has a yield and tensile strength equivalent to that of 356 alloy but elongation values are slightly improved, particularly when the alloy is in the modified condition. However, this fact may not justify the use of such an alloy for commercial purposes since the problem of sensitivity to magnesium content would become even more pronounced.

Modification

One of the experimental difficulties encountered in the work on modification of aluminum-silicon-magnesium alloys was a means of determining the amount of sodium actually in a given alloy. Chemical analysis, whether by wet methods or by spectrographic means, is tedious and costly. With the number of analyses required for this work separate analyses were deemed impractical, therefore, the amount of sodium reported throughout this work are the amounts of sodium added and not necessarily the amounts actually in the alloy.

From the results obtained it is evident that the amount of sodium added and the holding time after the addition are inter-related variables insofar as the modified structure is concerned. The optimum amount of sodium added to a particular alloy would be the smallest amount which would allow a practical holding time and at the same time, produce a fully modified structure. On the basis of the holding time series 0.01 to 0.03 per cent sodium and a holding time of 5 to 10 minutes would meet these requirements. In deciding what addition should be used, consideration must be given to the degassing treatment used. For melts degassed with chlorine 0.03 per cent sodium is the least amount that will retain modification effects for 5-10 minutes (Table 3). For melts degassed with nitrogen, as in Fig. 5, an addition of 0.01 per cent sodium is sufficient.

This approach assumes that the primary cause for the increased tensile properties in binary aluminum-silicon alloys is the form of the silicon constituent in the eutectic areas. If this were the case, it would be logical to assume that as a result of an extended holding time, there would be a gradual degradation of properties as the microstructure reverted to the normal form. This is not true for 356 alloy, however, as shown by the correlation of tensile properties with the microstructure. The microstructure goes through the complete transition from the modified to the normal structure with no change in tensile properties. The original data, shown in Table 4, indicates that the elongation increases from 2.4 to 5.1 per cent after a 5-minute holding time. This increase, however, could not be reproduced in subsequent experiments.

The work described in this report indicates that modification effects are detectable in aluminum-silicon-magnesium alloys containing 5 per cent silicon and magnesium in amounts less than approximately 0.2 per cent. The age hardening studies have shown that the hardness level attained by the modified alloys is slightly lower than that attained by the un-modified

alloys and the extent of this difference becomes less as the magnesium content is increased.

Room Temperature Aging Effects

From the results obtained on this phase of the work it is evident that the interval between the solution and aging treatments is an important factor insofar as the final properties which are attained are concerned. By allowing the tensile specimens to remain at room temperature for an interval of 24 hours following the solution treatment, it is possible to reduce the tensile and yield strengths by as much as 4000-5000 psi. Elongations, on the other hand, are slightly improved.

Quadt[20] investigated this effect and obtained similar results, and also observed that the effect appears to be confined to alloys in which Mg_2Si is the precipitating phase. This investigator suggested a 24-hour interval prior to aging as a means of meeting minimum elongation requirements in borderline cases. While it is apparent that the minimum elongation requirement of 3 per cent has not been met in this phase of the work, it should be further noted that the magnesium content of alloys used was at the upper limit of the range given in Table 1 (the actual analysis was 0.32-0.35 per cent magnesium). The difficulty encountered in meeting elongation requirements in this case lends further support to the belief that the magnesium content of 356 alloy should be maintained towards the lower limit of the specified range of 0.20-0.40 per cent.

Although it is not within the scope of this work to determine the fundamentals of this phenomenon, it seems obvious that the age-hardening characteristics at elevated temperatures are altered by the prior precipitation. In all probability the effects observed here are reversion or retrogression phenomenon described in the literature on precipitation hardening. The net result of the two aging processes is a gradual lowering of the strength values and increase in elongation as the first aging time is increased.

Heat Treatment

Throughout this investigation only two aging temperatures have been used to any extent. These temperatures, 310 F and 400 F, were selected as representative of those used to produce 356 alloy in the T6 and T7 conditions, respectively. The time at each temperature was selected on the basis of the maximum hardness attained as determined from the age hardening studies.

It is immediately apparent that, although the tensile and yield strengths are substantially above the values reported in the various handbooks, the elongations are below the typical values and in some cases below the minimum values. In order to produce elongations of the order of 4-5 per cent, it may be necessary to readjust the time at the aging temperature (in addition to closely controlled magnesium contents) to produce a slightly underaged condition. The increase in elongation necessarily occurs at the expense of the tensile and yield strengths.

Conclusions

It is concluded that:

1. Small variations in the magnesium content of aluminum-silicon-magnesium alloys in general, and in 356 alloy specifically, produce a pronounced effect on the tensile properties. The allowable range of magnesium content in 356 alloy is capable of producing a spread of 4,000-7,000 psi in the observed yield strengths.

2. Small variations in silicon content have little effect on the tensile properties of alloys of approximately the 356 alloy composition. The variations become significant when less than approximately 0.2 per cent magnesium is present.

3. Sodium modification does not appear to be beneficial to the heat-treated 356 type alloy. Modification effects that improve mechanical properties are observed to some degree in alloys containing 5 per cent silicon and less than approximately 0.2 per cent magnesium.

4. For aging times up to 6 hours sodium additions do not significantly alter the aging characteristics of the 356 type alloy. The time required to age to maximum hardness is essentially the same for both modified and normal alloys.

5. Aluminum-silicon-magnesium alloys of the 356 composition must be utilized in an underaged condition if satisfactory elongation values are to be obtained. (In this condition, however, the full strength potentialities are not realized.)

6. A significant impairment of the tensile and yield strengths with an improvement in elongation occurs as a result of an extended time interval between the solution treatment and artificial aging treatment. Room temperature aging effects occurring over a period of 24 hours are sufficient to lower the tensile and yield strengths by 4,000-5,000 psi.

Acknowledgments

The authors wish to thank the Department of the Army for permission to publish this paper and to extend their appreciation to the members of Pitman-Dunn Laboratories who contributed to the work.

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DISCUSSION

Chairman: E. V. BLACKMUN, Aluminum Company of America, Pittsburgh, Pa.

Co-Chairman: F. L. TURK, Joy Mfg. Co., New Philadelphia, Ohio.

D. L. COLWELL[1]: The elongation of the regular grade 356T6 (not the special high purity grade) seemed to be under the 3 per cent minimum in most cases. Could this point to something unusual which might throw some doubt on the validity of the conclusions? On relative values the conclusions seemed justified.

But if Mg is kept on the low side to get elongation, the yield strength is not obtained, and if kept on the high side to get yield, elongation is not obtained. Is then, the specification wrong?

MR. HARRIS *(Reply to Mr. Colwell):* The author's recommendation that the magnesium content of 356T6 alloy be controlled at the low side of the range is based upon results which show that all the mechanical property requirements can be most easily satisfied by this practice. This does not imply that every heat containing more than 0.3 per cent magnesium will fail to comply with elongation requirements. The results obtained with a given heat of metal are not solely dependent on the magnesium analysis but upon such additional factors as impurity content, gas content, and foundry practices employed. When these factors tend to be less favorable, a definite advantage is realized by setting up internal controls as recommended to insure compliance with elongation requirements under these conditions.

The dependence of the properties of this alloy on the impurity content is shown in Fig. A. This is a plot of tensile properties vs aging time for two 356-type alloys of different iron and magnesium contents. It will be noted that in spite of the higher magnesium content of the special alloy, the elongations are considerably improved. The difference in tensile and yield strengths probably reflects the difference in magnesium content.

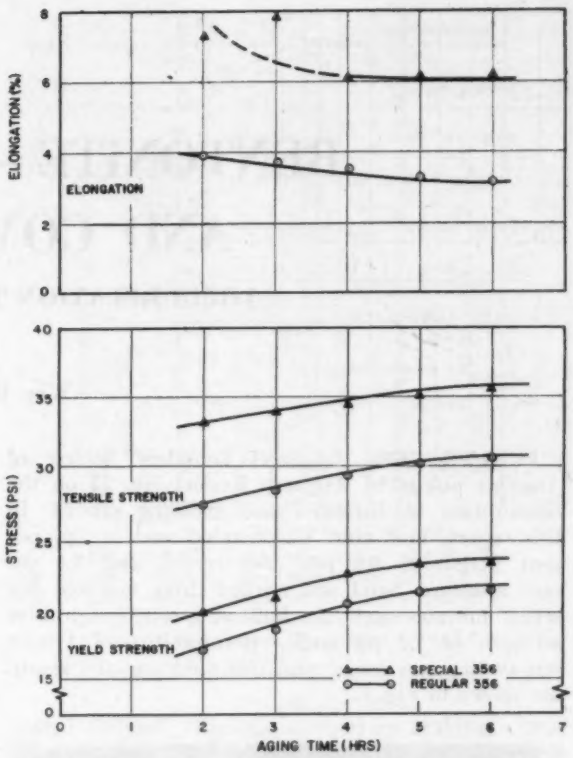


Fig. A — Tensile properties vs aging time for two 356-type alloys. Regular 356 contains 6.98% Si, 0.21% Mg, 0.38% Fe. Special 356 contains 7.35% Si, 0.33% Mg, 0.08% Fe. Solution treated 16 hr at 1000 F for times shown. Each point represents average of three bars.

A. M. MONTGOMERY[2]: What structural changes were brought about with 0.05 per cent sodium in comparison with the changes brought about with only 0.01 per cent sodium?

MR. HARRIS *(In Reply to Mr. Montgomery):* In general, the structural changes are the same for both 0.05 and 0.01 per cent sodium additions. There is, however, a difference in the time interval that modification effects will persist in the melt. When small additions are made, the melt must be poured soon after. For larger additions an appreciable time interval can pass before modification effects disappear.

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1. Apex Smelting Co., Cleveland.
2. Aluminum Company of America, Cleveland.

BENTONITE — PROPERTIES AND COMPOSITION

THEIR RELATION TO CASTING DEFECTS

By

Wm. D. Emmett*

In April, 1950, the Steel Founders' Society of America published Research Report No. 23 on the constitution of discarded steel molding sands.[1] In this report, new sand was bonded with a 2½ per cent bentonite, 0.6 per cent cereal, and 3.1 per cent moisture. Sand was mulled three minutes dry, seven minutes wet. It had a green compressive strength of 5.2 psi and a permeability of 140. A test casting was made with this sand and the results are shown in Fig. 1.

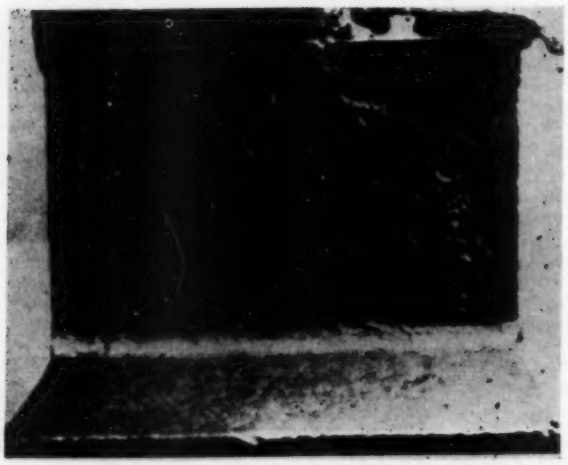


Fig. 1 — Test casting made with new sand.

The casting and molds were shaken-out in wheelbarrows and the collected sand was remixed for the second use by breaking up the lumps in the muller. No screening was used to remove the fines. Bentonite and water were added to obtain the same sand strength and moisture content as in the first mixture. In all, five castings were made.

Figure 2 shows the results of the fifth casting. This casting shows heavy erosion as well as buckling and almost everything that can happen when metal fills a mold.

Figure 3 shows a used sand bonded to the same physical properties as in Fig. 1 and 2, however, the fines and metallic substances were removed. These castings were three out of a series of 50 castings which were made. Removing the fines eliminates buckling and the other defects, however, erosion of the sand is very pronounced. This was characteristic of all 50 castings which were made.

Referring again to the erosion of sand reclaimed by this method, there must be some reason for this. We see the temperature tests, such as green compressive strength, permeability, deformation, and moisture are in line with new sand. What has actually happened to this sand? Are the oxides in the mixture causing the vitrification of the sand? Has the sand itself changed or has the bonding clay changed?



Fig. 2 — Drag surface of plate as produced in fourth re-use of sand.

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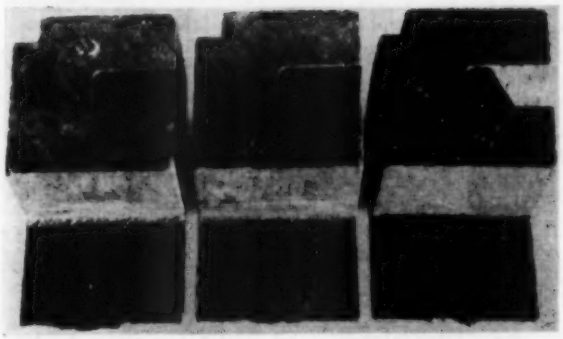


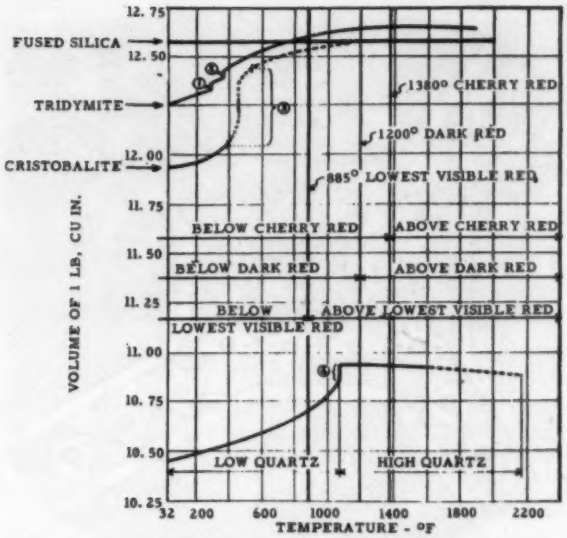
Fig. 3 — Castings made with used sand with fines and metallics removed.

Let us explore the first possibility. New sand containing 6 per cent bentonite and 0.6 per cent cereal was placed in a furnace at 2700 F and allowed to soak for one-half hour. Upon removing, we found that the sand had caked, not actually vitrified. The same mix was again placed in the oven and held at 2500 F. Once again the sand showed caking, approximately the same as before. The same mix was placed in the oven and held at 2300 F, and upon removing no great amount of caking could be observed. The same tests were made with used sand, and as near as could be ascertained, the used sand was comparable to new sand. Apparently, the presence of oxides has not affected the used sand to any appreciable degree.

Let us next consider the sand itself. Does the heat affect the sand grains in such a way that the expansion of used sand is different from that of new sand? If this is true, then no matter how the sand is reclaimed, it will cause trouble.

Figure 4[2] shows the different changes silica goes through upon heating. First we have the low quartz which changes to high quartz at 1063 F. This is the point where we encounter trouble in all silica sands. As can be seen, at this point we have a sudden expansion of 0.86 per cent. This is so abrupt as to be likened to a small explosion. High quartz changes to tridymite at 1598 F. However, this form is not stable at room temperatures and reverts back to low quartz. Consequently, we have nothing to worry about in this stage. At 2678 F, the tridymite changes to cristobalite. This form is very stable at room temperature. Referring to Fig. 4, we see that upon reheating cristobalite at 447 F, we have a volume change of 2.80 per cent which can be likened to a major explosion of the sand grains. If this happens, no bonding material known would hold the sand grains in place. However, refractory men say that cristobalite is formed only by time and temperature—and the small amount which would form against the face of the casting upon cooling to 447 F, would literally blow itself to small pieces, and be removed as fines. This possibly accounts for the eventual break-down of the sand grains. From this, it can be seen that the heating of the sand has no particular detrimental effect upon the grains, from the standpoint of expansion.

In the 1953 AFS TRANSACTIONS, an article was published entitled, "Factors Involved in Making a Sand Mixture."[3] In this article it was stated, "It is thought by many that it is impossible to re-bond sand that has been subjected to a temperature exceeding 2200 F and some claim that the critical temperature is much lower. The sand grain is thought to lose the ability to be re-bonded because heat in some way alters the residual negative charge on the sand grains."



No. 1 — 0.14% volume change (Reversible) at 243 F
No. 2 — 0.20% volume change (Reversible) at 326 F
No. 3 — 2.80% volume change (Reversible) at 392-527 F
(Most of this usually occurs about 447 F)
No. 4 — 0.86% volume change (Reversible) at 1063 F

Fig. 4 — Effect of temperature on silica.

Casting XX (Fig. 5) was made with sand that was taken from a casting which had 6-in. metal sections. The sand used was taken not farther than ¼ in. from the face of the casting. Pouring temperature of the casting was 2820 F. The reading was taken with an optical pyrometer so some error in the temperature reading can be expected. However, we have every reason to believe that the sand ¼ in. from the casting, or some of it, attained at least 2200 F. This sand was bonded with 6 per cent bentonite and 1 per cent cereal. Mulling time was one minute dry and

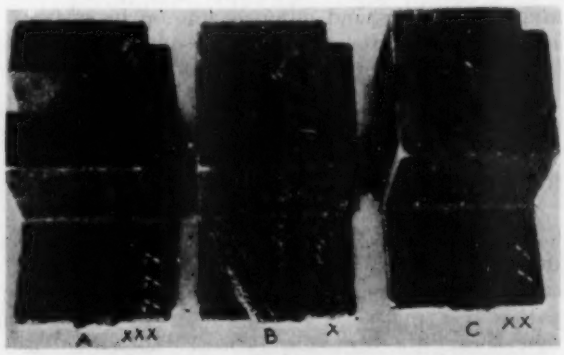


Fig. 5 — Appearance of castings showing effect of heat on bonding ability of used sand mixtures.

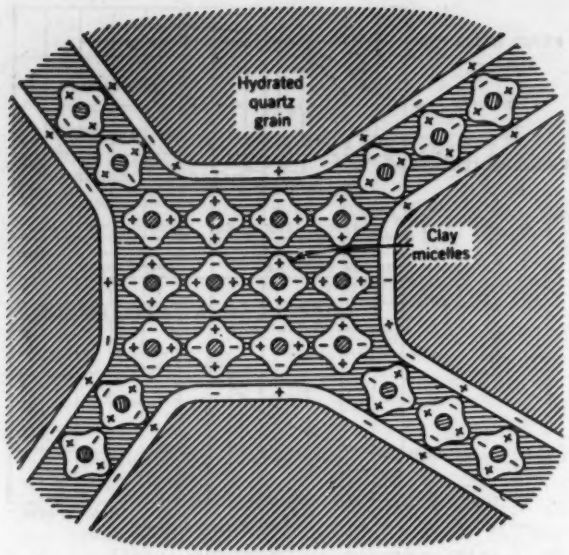


Fig. 6 — Schematic illustration of disposition of hydrated quartz grains and clay micelles in green molding sand.

five minutes wet. As can be seen, there is no erosion.

We will pass over Casting X (Fig. 5) at this time and refer to it later, as it belongs to another series of tests.

Sand used in Casting XXX was placed in an oven and heated to 2500 F for a period of five hours. This sand was also bonded with 6 per cent bentonite and 1 per cent cereal. Mulling time was again one minute dry and five minutes wet. This casting is comparable to Casting XX, as to erosion. Apparently, if the sand does have a negative charge, it is in no way affected by heat.

So far, we have accomplished nothing more than proving to ourselves that oxides and heating the sand have no apparent tie-in with erosion.

We still have the bentonite to explore. In 1952, an article was published in the AFS TRANSACTIONS entitled, "The Nature of Bonding in Clays and Sand Clay Mixtures," by L. M. Doran and H. F. Taylor.[4] In it, they state that most foundrymen, until recently, believed clay to be a simple chewing-gum-like material, which was capable of sticking the sand grains together. However, their findings, shown in Fig. 6, are far more complex. Here we see the sand grain with a plus and minus charge on its surface—and the clay particles also with a plus and minus charge. These sand grains and clay particles, when hydrated, exert a great attraction for each other—and in this way bond the sand together.

However, once the attraction has been established, it still remains if the water of hydration is removed.

Calcium and hydrogen bentonites have very little hot strength and high green strengths. For this reason they are little used in the manufacture of steel castings. By adding sodium carbonate to them, the sodium ion replaces the calcium or hydrogen ion, thus causing the bentonite to become somewhat of a sodium or western bentonite. Unfortunately, these types of bentonite have little of the base exchange

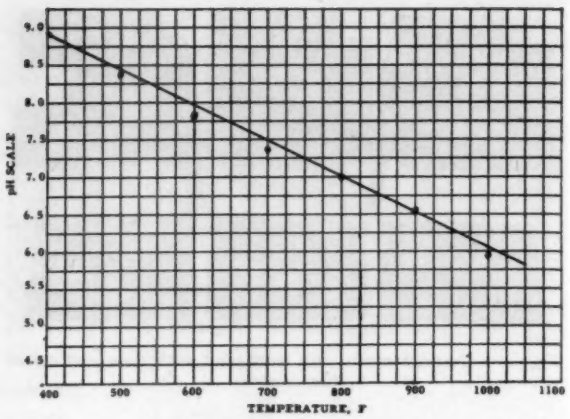


Fig. 7 — Bentonite held at each temperature for 12 hr. Original bentonite, pH 8.9.

material, so not much hot strength can be developed with these bentonites.

Sodium bentonite was placed in an oven at 400 F, 500 F, 600 F, 700 F, 800 F, 900 F, and 1000 F, and held for 12 hours at these different temperatures. It was then removed and allowed to cool to room temperature. The original bentonite, before heating, had a pH of 8.9 and a water of crystallization of 5.5 per cent. See Fig. 7.

Upon testing the 400 F bentonite for pH, it was found to be almost unchanged. The pH was 8.7, a drop of 0.2 of a point. At 500 F, the pH was 8.4, a drop of 0.5 of a point. At 600 F, the pH was 7.8, a drop of 1.1 points. At 700 F, the pH was 7.3, a drop of 1.6 points. At 800 F, the pH was 7.0, a drop of 1.9 points. At 900 F, the pH was 6.7, a drop of 2.2 points. At 1000 F, the pH was 5.9, a drop of 3 points. It will be noticed that at 800 F, the bentonite is neutral; at 900 F, and 1000 F, the bentonite has become acid. What has happened to the sodium ion upon heating is not known by the author. It could be combining with the silica to form sodium silicate, or some more complex compound. However, upon heating, the sodium ion is replaced by the hydrogen ion steadily as the heat increases.

The bentonite was again tested—this time for water

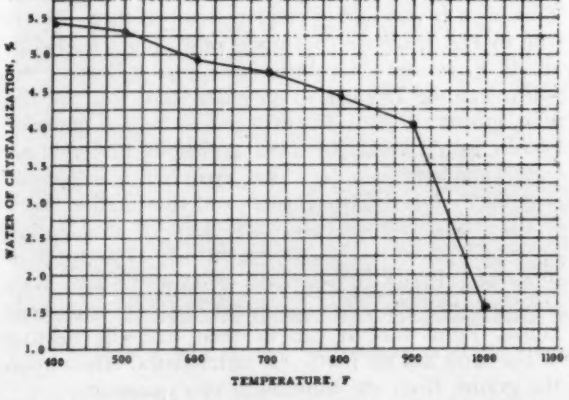


Fig. 8 — Remaining water of crystallization in bentonite after being held at temperature for 12 hr. (Original 5.5%.)

of crystallization (Fig. 8). It was held at 400 F and was found to have lost 1.63 per cent of its water of crystallization. The bentonite heated at 500 F was found to have lost 4.2 per cent of its water of crystallization. Bentonite heated at 600 F lost 10.5 per cent of its water of crystallization; and heated at 700 F it lost 13.1 per cent of its water of crystallization. When heated to 800 F, bentonite lost 18.7 per cent of its water of crystallization; and heated at 900 F, 26.2 per cent of its water of crystallization was lost. Bentonite heated at 1000 F was found to have lost 72 per cent of its water of crystallization.

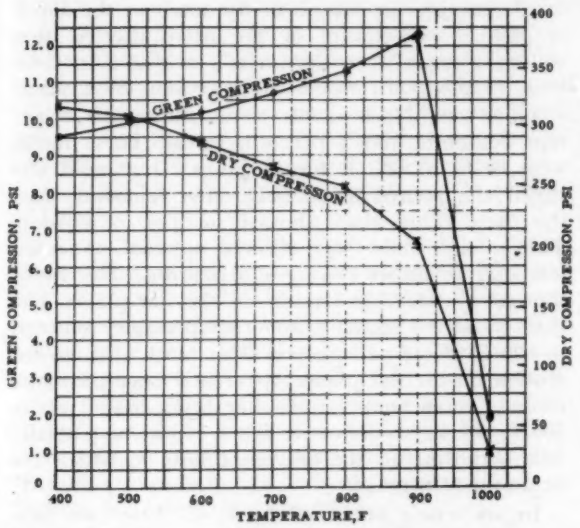


Fig. 9 — Properties of bentonite held at each temperature for 12 hr.

This same bentonite was used to bond sand (Fig. 9). The sand was bonded each time with 6 per cent bentonite and 1 per cent cereal. The sand was mulled one minute dry and six minutes wet. The 400 F bentonite (Fig. 9) had a green compressive strength (gcs) of 9.5 psi and a dry compressive strength (dcs) of 322 psi; 500 F bentonite had a gcs of 9.9 psi and a dcs of 310 psi. Bentonite at 600 F had a gcs of 10.2 psi and a dcs of 294 psi and at 700 F, a gcs of 10.7 psi and a dcs of 277 psi. At 800 F, bentonite had a gcs of 11.4 psi and a dcs of 254 psi and at 900 F, bentonite had a gcs of 12.3 psi and a dcs of 209 psi. Bentonite at 1000 F had a gcs of 1.9 psi and a dcs of 29 psi.

From this work, we can very readily see why our sand erodes as the bentonite is heated. We find we are gaining in green strength, however. As we gain in green strength, we are losing the dry strength, and, likewise, our hot strength. Castings were made from each of the fired bentonites and the erosion became progressively worse.

Referring back to Fig. 5, Casting X was made with the bentonite which had been heated to 1200 F. Six per cent bentonite and 1 per cent cereal were added to new sand. The pH was adjusted to 8.1 by the addition of sodium carbonate to the mix. The sand was mulled one minute dry and six minutes wet;



Fig. 10 — Test castings from bentonite heated at 900 F.

green compressive strength was 1.8 psi and the permeability 105. As can be seen, actually not only the sand on the flat plane eroded badly, but the notch also eroded.

From the test, we see that the bentonite has lost all of its former bonding power. It is completely dead. Any of this which is left in the sand will have very small effect on the erosion problem. Our problem then becomes this—what can we do with the bentonite which is giving us extra green strength? Can this be used to our advantage?

Casting I (Fig. 10) was made from the 900 F bentonite. The sand was bonded with 6 per cent of this bentonite and 1 per cent cereal. Mulling time was one minute dry and six minutes wet. The pH was raised to 8.3 by the addition of sodium carbonate; green compressive strength, 10.1 psi; deformation, 0.021 in. per in., and permeability 140. As can be seen, the flat plane eroded badly.

Casting II (Fig. 10) was bonded in the same way as Casting I. The pH was 8.0, mulling time was one minute dry and seven minutes wet. Green compressive strength was 9.4 psi, deformation, 0.030 in. per in., permeability 127. Erosion was much less than in Casting I. Casting III (Fig. 10) was bonded with the same material and the same quantities as Casting I and II. The pH was 8.1, mulling time one minute dry, nine minutes wet. Green compressive strength was 8.3 psi, deformation, 0.035 in. per in., permeability, 121. This casting shows no sign of erosion whatsoever.

Casting IV (Fig. 11) was bonded the same as Castings I, II and III. The mulling time was the



Fig. 11 — Test castings from heated bentonite.

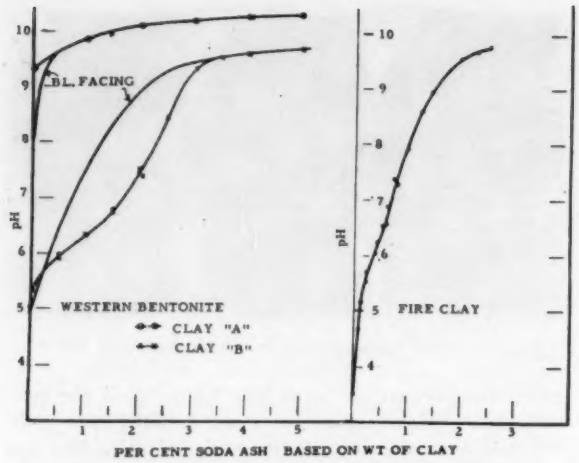


Fig. 12 — Percentage of sodium carbonate needed to raise the pH of bentonite.

same as Casting III, however, the pH was raised to 10.1. This casting shows signs of vitrification on the flat plane. We will refer to Castings V and VI later.

Sodium itself is a very strong flux. Large quantities will lower the vitrification point a great deal. The graph (Fig. 12) by Zang and Grott[5] shows the percentages of sodium carbonate needed to raise the pH of benotonite. It will be noticed that to raise the bentonite from 8 to 10.1, approximately 3¼ per cent soda ash, by weight, must be added. This amount of sodium in the sand would have a great effect on the vitrification point of the sand. hence the vitrification in Casting IV.

Casting V (Fig. 11) shows the same sand as that shown in Casting X, Fig. 5. However, the pH was raised to 8.4 and the sand was mulled one minute dry and 30 minutes wet. The green compressive strength was 6.3 psi, deformation 0.021 in. per in., and permeability, 118. Erosion on the flat plane is very bad, but a great deal less than that of Casting X. Possibly with a long enough mulling time, bentonite, when it is apparently dead, can be revived to make acceptable castings if the pH is properly controlled. This is merely an experiment. Thirty minutes, or more, mulling time would be more than any foundry would care to spend mulling sand.

Casting VI was made with our system sand. We have seen that good castings can be made with 6 per cent bentonite and 1 per cent cereal. In order not to use more bentonite than necessary, and AFS clay test is run daily. In this way, however, our test varies from the AFS procedure. We have a considerable amount of carbonaceous material in our sand which will actually settle about the same as clay. After the test is made and the percentage of loss is obtained, we then take a gram of sand and burn the carbon from it. Subtracting the loss from the original clay loss, we arrive at the true clay content of our sand.

Sand used in Casting VI had retained clay to the extent of 5.1 per cent. Therefore, 0.9 per cent of new clay was added and 0.4 per cent cereal. The pH held at 8.1 and the mulling time was one minute dry and nine minutes wet. Green compressive strength was 10.1 psi, deformation, 0.036 in. per in., and permeability, 124. No erosion has been encountered since we have mixed sand in this way. We use a pH of 7.8 to 8.3 for this reason. Good castings are made with new sand and new bentonite and this sand will test from 7.8 to 8.3.

At a technical meeting, each foundry participating submitted one or more test castings. In each case, the castings which were made in green sand were superior to those that were torch dried or oven dried. Possibly the 1200 F of the torch or the 500 F or 600 F temperature of the oven affected that bentonite exposed on the surface, as shown in these tests, to the point where slight erosion took place.

Three test block molds were made with 6 per cent bentonite and 1 per cent cereal. These molds were dried at 600 F for 12 hours. One mold was closed and poured immediately after removing from the oven. This block showed no sign of erosion. The second mold was allowed to cool to room temperature before closing and pouring. This block showed considerable erosion on the flat plane. The third mold was cooled to room temperature, reheated to 600 F, closed and poured. This block also showed erosion on the flat plane. Apparently once bentonite cools to room temperature after being heated above 300 F, the bond shown in Fig. 6 is lost or partially lost, according to the degree of heat to which the bentonite is exposed.

In an article published by E. C. Troy[6] on bentonite, he noticed a difference in the molding sand with different shipments of bentonite. At times the sand was very brash and sand inclusions were found on the cope side of the castings. He attributed this to the fact that the bentonite contained less than 8 per cent moisture as purchased and found that a small portion sprinkled on distilled water would settle in less than 20 seconds.

We have, from time to time, encountered the same difficulty with different shipments of bentonite, even when the moisture was 8 per cent or better. This led us to believe we should investigate the bentonite further. From the work done by Zang and Grott[5] and also the work by B. H. Booth,[7] we see that the calcium and sodium oxides play a very important part in the behavior of bentonite. During our experiments with bentonite, it was found that when the percentage of sodium oxide did not exceed the percentage of calcium oxide by two to one or more, even if the bentonite contained 8 per cent moisture, the bentonite would settle in less than 20 seconds.

Test blocks were made with this bentonite which had less than a two to one ratio of sodium oxide to calcium oxide and the results shown in Fig. 13 prove there is considerable erosion on the flat planes.

Test blocks were also made with bentonite having a sodium oxide-calcium oxide ratio greater than two to one, but having less than 8 per cent moisture as purchased. The results are shown in Fig. 14.

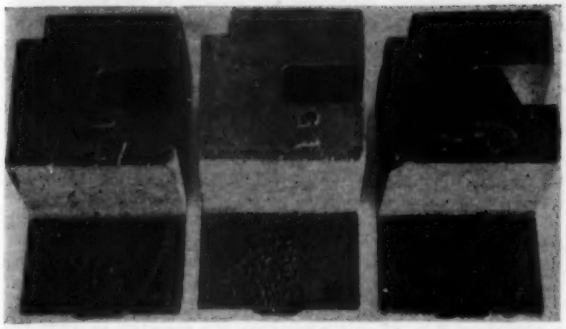


Fig. 13 — Test castings.

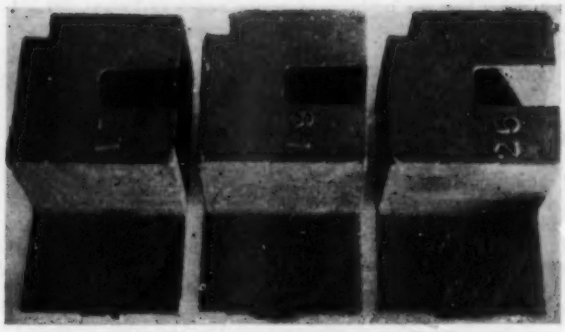


Fig. 14 — Test castings.

These blocks also show erosion on the flat plane. Settling tests were conducted and it was found that if the bentonite had 8 per cent moisture as purchased and the sodium oxide-calcium oxide ratio was two to one or better, with the sodium oxide predominating, the settling time was *always* 20 seconds or longer.

A gram of bentonite was heated to 2400 F for 10 seconds and cooled to room temperature. The pH and water of crystallization were tested. They were found to be unchanged. A gram of bentonite was heated to 2400 F for 25 seconds, cooled to room temperature and again tested for pH and water of crystallization. The pH was found to be 7.9, a loss of 1 point, and the water of crystallization was found to be 4.65, a loss of 15.4 per cent. Apparently, the bentonite loses some of its hot strength even in 25 seconds at elevated temperatures.

Before definite conclusions can be established, more research is required to correlate the loss of hot strength with the lengthening of the pouring period, also to determine whether or not the sand mixture will erode during the pouring cycle when the loss of water of crystallization and pH occur.

Making test castings gives positive results, but this method is too slow, time consuming and expensive. We have developed a test for erosion, which for all appearances, is a very fine substitute for the casting test. These sand mixtures would show very well with the Cain test of continued immersion, alternate immersion, or radiant heat. However, as we have seen, they erode badly. Using a variation of this test, we plunge the specimen into a riser that has been poured. This causes a film of metal to solidify on the specimen. The specimen is then plunged into a ladle of hot metal and held until the film has melted. If the sand will erode, the specimen will show erosion on the surface. This test has been checked many times against the casting and the results are very similar. This simple test allows us in a few minutes to ascertain whether the sand mixture is correct.

In conclusion, the author would like to bring out these points: (1) Sodium carbonate in its concentrated form does not ionize and should be added to the sand mixture in a dissolved state. (2) A laboratory muller should not be used for conducting experiments of this nature, as no correlation is possible between the large muller and the laboratory muller. (3) The pH of molding sand has been quite controversial throughout the United States. We do not, from our experiments, believe that new sand with good bentonite can be improved by altering the pH factor. By good bentonite, we mean bentonite which will test between 8 and 9 as to pH and which has a settling test of not less than 20 seconds. (4) Molds which have been oven dried or skin dried should be closed and poured immediately, before they cool, to avoid erosion of the sand.

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DISCUSSION

Chairman: E. C. TETZLAFF, Pelton Steel Castings Co., Milwaukee, Wis.

Co-Chairman: E. C. TROY, Foundry Engineer, Riverton, N. J.

A. H. ZRIMSEK[1] *(Written Discussion):* The author is to be commended for a fine paper. He has used a basic approach in attacking a basic problem of the steel foundryman.

With one exception, this writer concurs with the general content of the paper. The exception is to his reference to the small laboratory muller as unsuitable for studies of this type. It has been this writer's experience that the mulling conditions of the larger shop mullers can be closely simulated by the laboratory models through proper control of load size, mulling cycle, and order of additions.

Studies on the deterioration of bentonites in the medium temperature range (500-1300 F) have been carried out by this writer with results similar to those of the author. The medium temperature tests should be given consideration when a gage of bentonite durability is desired.

The deterioration of bentonite in this temperature range has two major causes. First, the gelling tendency (viscosity) is decreased by increase in temperature. The decrease is gradual over a range of temperature and is caused by loss of intercrystalline water. The gelling tendency is usually completely lost by extended heating at 900-950 F. However, some Western

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bentonites lose the property at temperatures as low as 700-750 F, and Southern bentonite loses the gelling property at temperatures near 600 F.

The loss in viscosity is not only a function of temperature, but time as well. While some bentonites lose the property after short periods at 900-950 F, others retain it after 2-3 hours at 1200 F.

The second phenomenon observed in the deterioration of bentonite is the reduction in ion exchange capacity with increase in temperature. The author has chosen to measure this property as a function of pH. Although not the most accurate, it is a simple method and must suffice until simpler methods of ion exchange determinations are devised.

The influence of temperature on ion exchange capacity varies from one bentonite to another, and as with gelling is also dependent upon time at temperature. The reduction is dependent upon the crystal structure and the particular ion present. Calcium (Southern) bentonites are subject to exchange reduction at temperatures much lower than Sodium (Western) bentonites. Use of pH as a measure of ion exchange is difficult beyond 1000 F. Above this temperature it is possible for clays to change from basic to acid to basic over a period of time. The return to basic is probably caused by oxidation reactions at the higher temperatures.

While some clays are acid after holding at 900-1100 F, others maintain pH above 8.0 to temperatures near 1300 F. Bonding quality, however, is usually lost at temperatures under 1000 F, held 24 hours. While some clays lose their bonding power after short periods at 1000 F, others retain theirs after holding 2-3 hours at 1200 F. It is also possible to obtain bonding action from clays that have completely lost the gelling tendency.

We would again like to congratulate the author for his fine work. More data of this same type are required to help the foundryman better evaluate one of his basic raw materials.

NOVEL ALUMINUM ENGINE DIE CASTINGS

By

E. N. Jacobi*

Production of aluminum die castings, as an art, has been considerably advanced in recent years by the aluminum engines of the author's company. These engines and their various modified models are an extension of 4-cycle, 1-cylinder engines. They are offered as portable power for home and garden appliances, tools, and for builders' and contractors' auxiliaries.

Engines, the cylinders of which employed aluminum alloys, cast integrally with the crankcase or without, have been produced for many years. These had sleeves of ferrous alloys such as cast iron as bore liners, in most embodiments. These were largely automobile engines and outboard boat engines. A few engines were made without liners, with or without chrome-plated bores.

However, the subject engines are the first air-cooled, 4-cycle engines, die cast with an integral crankcase and with their fins and valve housing parts, that have been manufactured in mass production, and without bore sleeves.

The extension of the die casting process to the production of the subject engine parts was initiated early in the manufacture of these engines. Following the adoption of die-cast governor, carburetor, ignition and small control parts, the larger structural parts were redesigned and adapted to production by the die-casting process and to the alloys suitable to the process.

An early forward step was made by die casting the engine flywheels in zinc alloy by the die-casting process. These flywheels carry the field magnets of the magneto. The field magnets and their pole pieces were fastened to the flywheel simply by imbedding them as inserts in the castings. The magnets are of chromium permanent magnet steel, and necessarily must be heat treated. It was found possible by suitably adjusting the casting conditions and other details of the process, to cast the treated magnets into the bodies of the flywheels without reducing their magnetic hardness and impairing their permanency as magnets; although the molten metal temperature far exceeds the temperature at which chromium magnet steel begins to anneal.

On some models of light-weight high-speed engines later developed, the flywheel magneto structure is die cast of aluminum. The magnets are of alnico. The higher melting temperature of aluminum die casting alloy does not damage these magnets when the casting conditions and the process are properly controlled. While a flywheel normally is designed to provide a large rotating mass, in this design aluminum is used to keep the inertia low in order that the shaft and the other working parts of the engine not be subjected to damaging forces when the blade or other tool driven by the engine, is suddenly stalled by a massive obstruction. The size of the wheel, of course must be such as to produce a proper cooling blower and to be strong enough for its other functions. Therefore the inertia of the wheel could not be brought to the desired value by reducing the sizes of the flywheel, but had to be brought within bounds by the low specific gravity of aluminum.

After the zinc alloy flywheel magnetos were developed, the following parts were redesigned and converted to die castings: pistons; connecting rods, first babbitted and then without bearing liners; cylinder heads; intake pipes and elbows; oil bases, reduction gear housings and cover plates; and unitary cylinders and crankcases.

Each of these advances involved redesign of the part to facilitate die casting execution as well as to function satisfactorily. The redesigning and modifying also affected surrounding parts in many instances. Extensive testing and experimentation was a necessary concomitant of each such advance.

In the development of the aluminum cylinder more than 100,000 hours of engine testing were accumulated prior to tooling and manufacturing. Many hundred engines were made and tested. The engineering work extended over a period of several years. Before starting the construction of die casting molds, engine parts of pressure-cast metals were produced by machining to simulate the production die-cast parts and were subjected to these extensive laboratory and field tests.

Intensive effort also was applied to the designing

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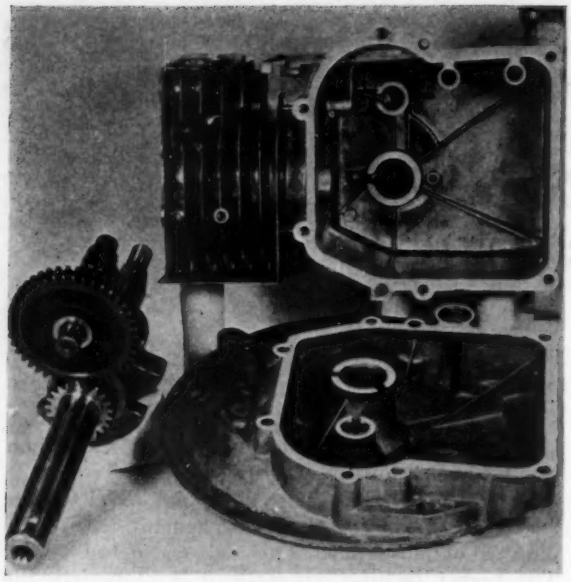


Fig. 1 — Cylinder and crankcase (upper), and oil sump and base plate die castings (lower). Crankshaft and camshaft are shown at the left.

of tools and manufacturing equipment in preparation for successful production. The mold designs, particularly, required much careful planning and experimentation to achieve the required accuracy, soundness, strength and finish. Problems of machining, treatments, finishing and quality required similar engineering and manufacturing process development, as did the process of chrome-plating the piston.

The cylinder and crankcase die casting is shown in Fig. 1. The crankshaft and the camshaft, with their gears in mesh, are shown beside it. The oil sump and mounting base plate casting lies in that position adjacent to the cylinder, that opens the gasket surfaces of the two castings as though these surfaces were the pages of a book. Thus, Fig. 1 shows a horizontal cylinder, vertical shaft with its principal parts laid juxtaposed.

The crankshaft and the camshaft are shown as lifted out of their bearings, which extend upward from the oil sump in the base plate casting. The ends of those shafts that extend forward in Fig. 1 are the ends that are to be assembled downward into the bearing hubs in the oil sump, the long shaft extension on the crankshaft being extended through the bottom wall of the oil sump as a power take off. It is commonly used to drive the blade of a rotary blade lawnmower.

The remote ends of the shafts are carried in bearings in the wall of the cylinder crankcase. These are shown in Fig. 1 as are the details of interior ribbing, the bosses and the baffles for the built-in crankcase venting breather, the valve stem, the ports, and the cam follower tappet guides.

A sectional drawing of the assembled engine (Fig. 2) brings out the aforementioned details. The relative wall thicknesses at various points are indicated in true proportion.

One of the notable design achievements that may be perceived in these illustrations is the use that was made of necessary structural elements needed for functioning of the engine, to facilitate the die casting process. A conspicuous example is the plate that is cast across the outermost parts of the cooling fins. This plate forms a wall to direct the cooling air from the fan at the closed side of the engine. But, it also forms a strong tie for the fins and a feeding channel to promote efficient mold filling as well as soundness in the casting. Other fins and bosses may be seen to combine functions similarly, such as, to strengthen the places requiring reinforcing while supplying mountings or auxiliary locating surfaces for the smaller engine parts, or to cooperate in the manufacturing operations.

The engine design and the mold designs have been coordinated to produce a variety of engines from the same major parts. The same mold may be used to produce horizontal shaft or vertical shaft engines, the side cover plate for the crankcase being different in the two engines. In the vertical shaft engine, the cover plate has a circular mounting flange and the cover plate forms an oil sump. The horizontal shaft engine casting has the mounting lugs on the crank-

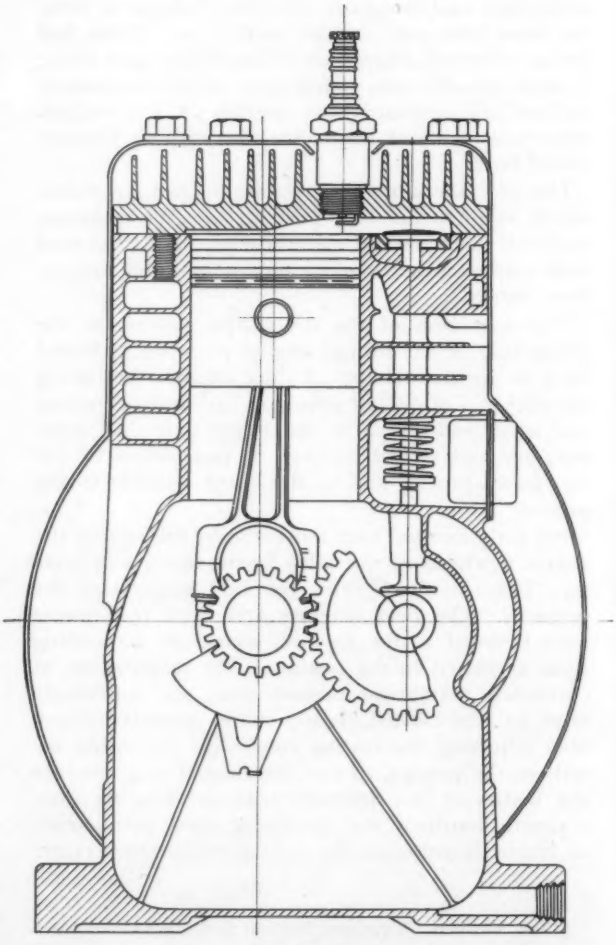


Fig. 2 — Cross section of engine showing relative wall thicknesses.

case. These lugs may be produced or not, at will, by interchangeable insert blocks in the mold.

The design of the engine has been so developed that the internal lubrication is equally effective in the horizontal engines and in the vertical engines. The parts for the lubricating function are different, though, but provision is made by bosses for either set of parts in the crankcases for the two styles of engines and in the crankcase.

The crankcase breather, likewise, is equally effective under all conditions of position and of engine speeds and duty requirements. The baffles, ducts and ports that perform these crankcase venting functions in this versatile manner are incorporated in the coring of the main casting for the crankcase, valve and spring housings, and cylinder.

Several patents were issued by the U.S. Patent Office on these novel concepts in engine construction.

Referring again to Fig. 1, the cylinder and crankcase casting weighs 3 pounds, 9 ounces. After machining, it weighs 3 pounds, 3 ounces.

The fins are .05 inch thick at their tips. The deepest fins are the top and bottom fins at their junctions with the air guide wall across the cylinder. Those drafts are 2.1 inches deep. The draft tapers are 2¼ degrees on a side. At some walls of the castings the draft is 1 degree. The cored crankcase walls have a draft depth of 3.45 inches. The greatest draft is inside the main bearing hub, where there is a 5-degree draft.

The cylinder bore finish is produced by boring and honing. No further surface treatment is applied. The piston is chrome plated.

The chromium plating on the pistons is approximately 0.0001 inches thick. It is applied by the techniques now commonly used for plating on aluminum. These methods consist of the well known cleaning and zinc surface treatments that are in general use for preparing aluminum for plating, followed by 0.00003-inch nickel strike and the chromium plating. The procedures are well known in the aluminum plating art. The chromium plate is of the dense kind, such as is used for mechanical chromium plating rather than decorative plating. It is not porous but is applied in an unbroken film. Every step in the process requires careful supervision and strict attention to details.

The composition of the alloy used for all aluminum alloy die castings, other than pistons, is:

Composition, %							
Si	Cu	Fe	Mg (Max)	Mn (Max)	Zn (Max)	Sn (Max)	Al
11-13	2-3	0.5-0.7	0.10	0.5	1.0	0.15	Bal.

The alloy used for pistons has the composition below:

Composition, %							
Si	Cu	Fe	Mg	Mn	Zn	Sn	Al
5-7½	2-3	0.5-0.7	0.10	0.5	1.0	0.15	Bal.

These compositions are based on a long background of experience in this particular art. Differences from the standards of other die casters were found necessary in order to fill the thin wall sections of engine castings which usually are intricate. This need for fluidity as well as for optimum strength and toughness found the best response in the high silicon range. However, a lower silicon content showed undeniable advantages for piston castings, that made it imperative to use the second mixture tabulated above despite the manufacturing convenience and economy that might have resulted from using the same composition for all aluminum die castings. The advantages of the lower silicon content stated above for pistons are:

1. Smoother surface finish produced in the machining and grinding operations.
2. Better dimensional stability (absence of warping) at the operating temperatures of the pistons. These temperatures range from 500 F. at the center of the head to 350 F. on the skirt.

The allowable iron content for these intricate and thin-walled engine castings is lower than can be allowed for the great majority of casting designs found in the commonly used aluminum die castings. A somewhat wider margin of safety against excessive iron pick-up was found necessary to avoid losses and failures due to brittleness. Some supplies of alloy to be handled in electric furnace equipment instead of in iron pots, may be accepted up to 0.8 per cent iron.

The magnesium, zinc and tin limits likewise were found imperative by experience in producing engine castings. In the general die casting trade, the simpler and thicker die castings commonly produced do frequently permit of using a much higher zinc limit. Most specifications for the tin limit are higher than the limit stated above, 0.30 per cent being a common specification. Nevertheless, it was found necessary to observe the stated limits of 1.0 per cent maximum for zinc and 0.15 per cent maximum for tin in the manufacture of these engine castings, else cracking trouble was encountered.

Melting is performed in electric furnaces. Some gas-fired equipment is being employed in the manufacture of the smaller castings. The molten metal is charged into electric induction holding pots at the cold chamber die casting machines, excepting that the gas-fired furnaces for the casting machines producing the smaller castings, function both as remelting and as holding pots.

These aluminum engine die castings are used as shaft bearings at a number of points, notably in the connecting rod, and at the main and cam gear bearings. The same alloys are used as tabulated above. These are not special bearing alloys, the tin content being regarded as an undesired incidental ingredient. They are not given any special treatment after the ordinary machining operations.

These machining operations consist of rough and finish boring, followed by diamond boring or by roller burnishing. The diamond boring and the roller burnishing produce smooth surfaces and exceptional accuracy, comparable with the finish and accuracy used in automotive engine manufacturing practice. The total tolerances range from 0.00025 inches in the wrist pin bearings in the piston and

connecting rod, to 0.0005 inches in the crankpin bearings, the main and other bearings.

Steel forged crankpins and the bearing journals are induction hardened to Rockwell C56. Pearlitic malleable iron crankshafts are not specifically hardened. The wrist pins are carburized and hardened.

These bearing combinations and the chrome-plated pistons bearing directly on the aluminum bores have been used from the beginning of the work on these aluminum engines. There are several years of successful experience on more than a million engines in the hands of users with these bearing combinations. Moreover, the diamond bored aluminum die castings on steel shafts have been used for many years earlier in cast iron cylinder engines. Engines also have been taken from production periodically and tested to destruction for durability. Their life is comparable to that of cast iron engines of the same size and performance.

Summary

This paper describes the four-cycle, one-cylinder engine of Briggs & Stratton from the standpoint of the problems of production by the aluminum die casting process. Redesigning of parts formerly produced in other metals necessitated the solution of a variety of problems of performance as well as of processing.

DISCUSSION

Chairman: B. L. MEREDITH, Federated Metals Div., American Smelting & Refining Co., South Plainfield, N.J.

Co-Chairman: W. BABINGTON, Bell Telephone Laboratory, Murray Hill, N.J.

CHAIRMAN MEREDITH: In the development of this casting it is assumed that the tolerance for tin and magnesium was established individually. The presence of minor amounts of tin in some aluminum alloys containing copper as a principal element is known to contribute to "hot-shortness". However, it has also been determined that additions of magnesium will suppress this effect by the formation of a magnesium-tin compound which has a higher melting point than tin. Did you have an opportunity to make this casting when both tin and magnesium were present, and if so, were the results different than when only tin was present?

MR. JACOBI: The tin and magnesium limits were fixed individually. No attempt was made to determine their joint effects on "hot-shortness." The magnesium was quite low in relation to limit of 0.10 per cent when difficulty was traced to tin.

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1. Apex Smelting Co., Cleveland.

EXPERIENCES WITH BASIC CUPOLA REFRACTORIES AND MELTING

By

E. Boywid* and A. P. Alexander**

H. Morrogh's classic paper[1] on the production of spherulitic or nodular graphite irons, presented in 1948, was the forerunner of many excellent works on this subject. Since higher casting quality was indicated, our management showed great interest. An investigation was encouraged. As a result, we placed nodular iron castings in the field for testing in 1949.

Our experience, plus a study of chemical requirements for nodular iron, indicated that we could not produce it economically in an acid cupola, and at the same time make quality gray iron in a mechanized foundry. Fortunately, there were pioneers in the field of basic cupola melting. Their investigations[2] showed that a basic cupola could use a very large percentage of steel scrap and produce any type of gray iron desired. The resulting low-phosphorus and low-sulphur iron made an excellent base for nodular iron.

Since the historical normal price of steel scrap is much lower than pig iron, basic cupola iron should show an economic advantage. We visited a number of foundries which, with one exception, operated basic cupolas a few hours daily for small tonnages. These visits plus generous advice and encouragement led us into a completely basic operation. We have operated for over a year with iron melted entirely from basic cupolas. During this period we averaged well over 100 tons per 8-hour day. Nodular iron and gray iron were poured concurrently from the same base iron. Both irons met or exceeded specifications. Many problems were encountered. In general it was a satisfactory operation.

Now it is proposed to examine the results, and, where possible, to compare them with those found with a year of acid melting in the same foundry. We investigated theoretical aspects and may mention or discuss some, but in the main it is hoped to confine the discussion and data to our actual results. It will be noted, from the *Refractory* discussion, transition to basic operation was gradual. We had no experimental cupolas, and regular production schedules had to be met. Many problems arose and changes had to be made in small increments to least upset our schedule. Even now our methods are being slowly improved.

Cupolas

We operate on alternate days with two no. 9½ cupolas. They are equipped with a centrifugal blower and air weight controls. A Griffin hot blast system

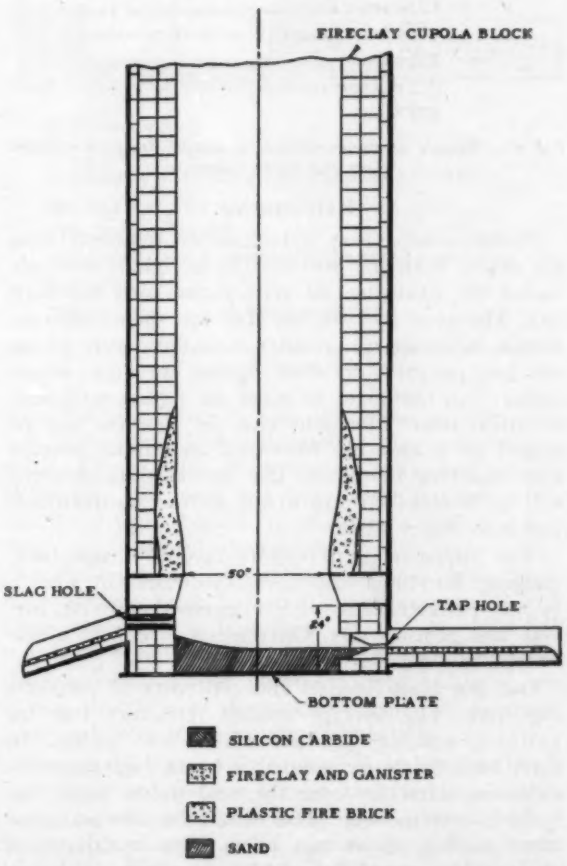


Fig. 1 — Sketch showing refractory layout in back-slagging acid-lined cupola.

*General Supt., Foundry and **Works Metallurgist, International Harvester Co., Memphis, Tenn.

delivers blast air at temperatures of from 450 to 600 F. Details of the interior are discussed under *Refractories.*

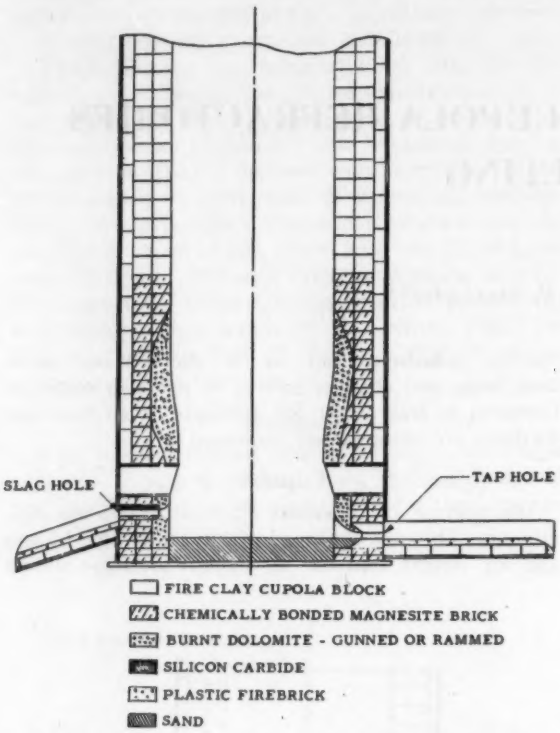


Fig. 2 — Sketch showing refractory layout in experimental short run basic cupola.

Refractories

Positive comparative statements on refractory costs are made. Both the acid and basic figures were obtained by itemizing all refractories used for each day. The total cost for the day was then easily obtained. Tonnage was divided into this figure giving the cost per net ton. Cost figures are f.o.b. source rather than our plant to make the figures adaptable to other areas. The daily cost per net ton was averaged on a monthly basis and an annual average was made for the year. The figures thus obtained will be brought up now in the *Refractory* discussion and later under *Costs.*

The approximate refractory layout of our back-slagging acid-lined cupola is illustrated in Fig. 1. It was patched daily with a gunned granular fireclay and ganister mix. On some days cupola blocks were also used.

Our hot blast lowered and concentrated the melting zone. The average annual refractory cost for patching was $.45 per net ton of iron melted. In early basic trials we gunned a burnt high-magnesia dolomite refractory over the acid patch. Some was sprayed over the well. The basic refractory was gone after melting about two hours. The remainder of the run was completed with acid practice. This was satisfactory for exploring characteristics of basic cupola iron melting, and establishing some control procedures. However, we never knew when we would convert to acid practice, and we had an unreliable slag basicity. Then we used about four inches of gunned dolomite and operated basic for four hours on a Saturday and stopped. The operation was entirely basic. After a few trials, we removed the inner layer of fireclay blocks in the well and in the melting zone to a height of about four feet above the tuyeres. They were replaced with chemically-bonded magnesite-chrome brick (Fig. 2).

Our slag hole became a major problem. After melting at the rate of about 15 tons per hour for two hours, we lost our silicon carbide slag block. The hole was about 6 in. in diameter. Shutdowns were required for repairs. We tried silicon carbide, plastic fireclay containing graphite, rammed magnesite and gunned burnt dolomite. None was satisfactory. The silicon carbide held up the longest. Burnt dolomite disintegrated in a few minutes. The well was also burning out excessively.

We tried a water-cooled bronze slag hole block developed by J. Sheley.[3] The block was entirely successful in resisting slag. Unfortunately, it burned out immediately if any iron touched it.

A plastic fireclay containing graphite was tried as a lining in the well over the basic brick. This looked good for 4-hour runs, but not for eight hours. Then we tried a mixture of crushed coal and tar (carbon paste), heated and rammed over the basic brick in the upper or slag zone of the well. When rammed in thicknesses over three inches, carbon paste was satisfactory. The lower zone or iron zone of the well was lined with rammed burnt dolomite. This zone

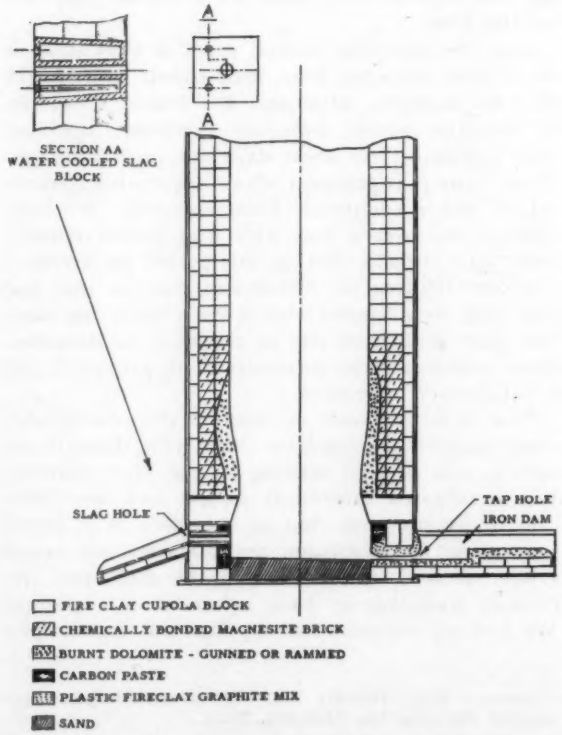


Fig. 3 — Sketch of refractory layout in experimental back-slagging basic cupola.

is quite deep in a back slagging cupola. The tap hole was still a regular acid block.

We decided to protect the bronze slag hole by controlling iron height. A dam was built in front of the tap hole, as illustrated in Fig. 3. It was necessary to have a minimum dam height of 9 in. to prevent slag from coming out of the tap hole with the iron. The tap hole was 1¼ in. in diameter. It was made of rammed burnt dolomite and was an integral part of the well and trough lining. Since the dam had to be at least 9 in. high, the iron level in the cupola was always close to the bronze slag block which could not be raised because of low tuyeres. After a number of 8-hour heats, no sign of wear was noted on the slag hole. Then the relatively small tap hole was partially blocked one day. Iron hit the bronze slag block in a few moments. A bad leak developed. At that time we discontinued our exploratory tests.

It is possible that a high tuyere rear slag cupola with a raised water-cooled slag block, large tap hole and a dam to control iron level would be the most logical and satisfactory method of operation. The slag hole should last indefinitely, while the problem of tap hole refractories would be greatly minimized with only iron to contend with.

When we converted to a regular basic operation,

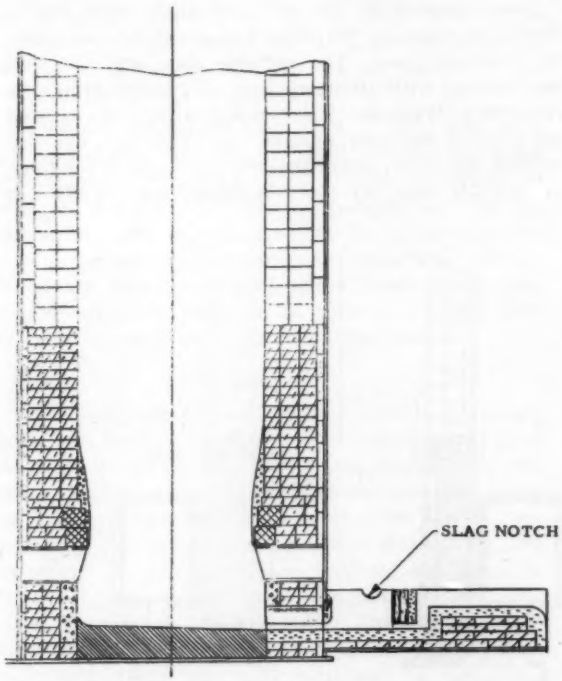


Fig. 4 — Sketch of refractory layout in experimental front-slagging basic cupola.

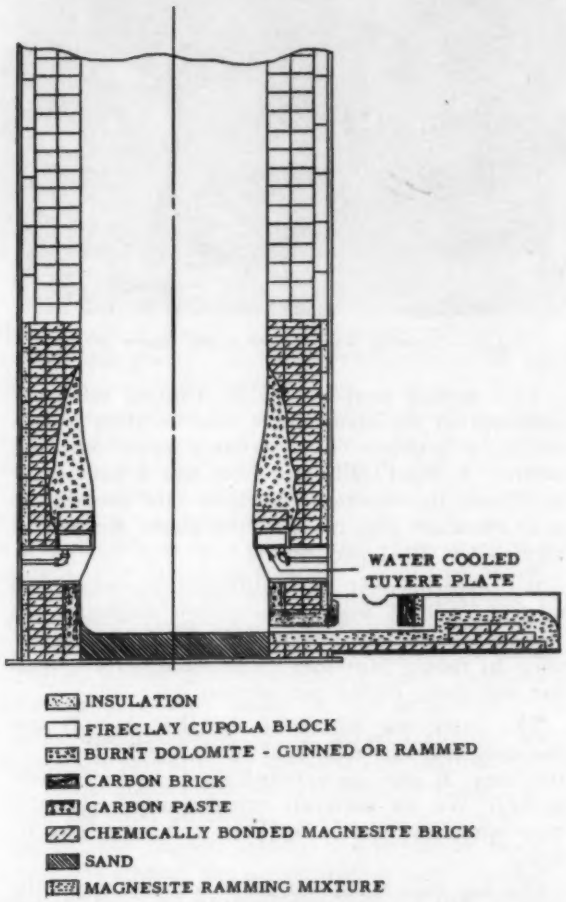


Fig. 5 — Section of experimental front-slagging basic cupola with water-cooled tuyere plate.

front slagging (Fig. 4) was used. The melting zone was lined with chemically-bonded magnesite-chrome brick. About 1 in. of insulating brick was against the shell. The well was also lined this way, but was faced with about 6 in. of rammed carbon paste. This paste formed the tap hole and inner lining of the outer front slagging well and slag notch. Magnesite-chrome brick backed the carbon paste in the outer well. The carbon paste on the slag skimmer was backed by a carbon brick over the magnesite-chrome brick. Fused chrome-magnesite bricks are shown directly over the tuyeres. These are discussed later and were used only experimentally.

Due to the lower melting zone caused by hot blast, the top of our tuyeres burned back far worse than would be true with cold blast. This made it highly improbable that we could operate more than eight hours. We made a water-cooled upper tuyere plate to control the above burnback. One plate was over each tuyere or a total of eight segments were used, as shown in Fig. 5, 6 and 7.



Fig. 6 — Showing water-cooled upper tuyere plates.

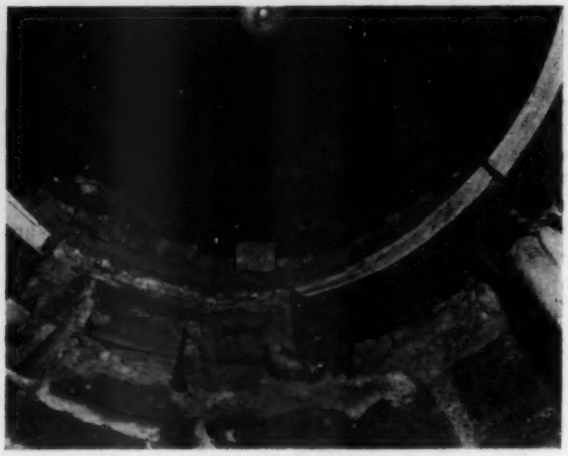


Fig. 7 — Basic brick ledge over water-cooled tuyere plates.

This method worked well in reducing refractory consumption for about three months, then leakers became a problem. This seriously upset our iron control. A heavy daily cost item was a basic brick ledge over the water-cooled plates. Our average annual refractory cost covering the above methods of melting was $2.41 per net ton.

We experimented with different materials and methods. A brick made from molten magnesite and chrome oxides was placed over the upper tuyere plate to reduce burn-back. The refractory cost that day was much higher per net ton.

The brick was successful the first day. It was the only material that held up for eight hours in this area. It may have lasted much longer if held at heat. We are currently experimenting with this type refractory in our tap hole with promising results.

Our tap hole is 2¼ x 4 in. It is made by placing a channel iron on a rammed magnesite base which forms the bottom and part of the side. Then carbon paste was rammed over it up to the tuyere plate as an integral part of the well. The carbon paste often sagged during melting. This started an air leak between the lower tuyere plate and the top of the carbon paste. A large pre-burned carbon paste block over the tap hole now minimizes this problem. We ram carbon paste into an old jacket and face it with steel plates on top and bottom. It is brought to a red heat for about 15 hours. On cooling, it is hard and remains so on reheating. The cost of basic refractories was high, so we looked at water-cooled cupolas. Foundries using them were generous in permitting us to see their practice, and to offer practical advice. The installation so well described by Levi[2] appeared most suited to our needs. We converted one cupola to water cooling (Fig. 8).

As seen in the illustration, 2½ in. EH pipe rise 66 in. above the tuyeres, about an inch apart, making the inside diameter of the cupola 65 in. Copper tuyeres 40 in. above the base, project 9 in. beyond the pipes. The tuyeres had an inside diameter of 8½ in. but we cut this to 6 in. We have 9 in. of chemically-bonded magnesite-chrome brick behind the pipe and in the well. The well is faced with carbon paste. We still use the pre-burnt carbon block for the top of our tap hole, and with the new cupola we have not had any blows.

Small patches in the well are made with plastic fireclay containing graphite. Large patches are made with carbon paste. The melting zone and the well are covered with about an inch of gunned dolomitic refractory. With the water-cooled cupola the cost of refractories for two months was $.89 per net ton melted. We feel certain that we can operate for longer periods now, so the refractory cost should fall

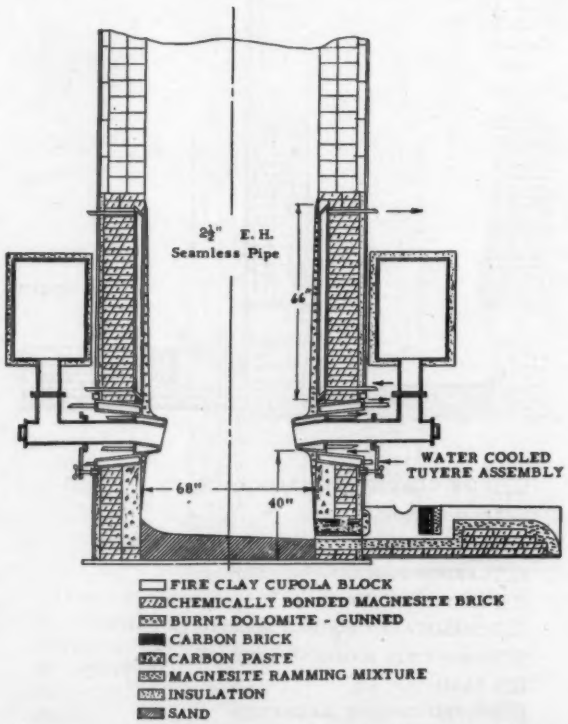


Fig. 8 — Showing basic-lined cupola with water-cooled tuyeres and melting zone.

drastically in the future. Water cost should be considered, but we hope to make this relatively insignificant with a recirculating system.

Slag

We analyzed slag and noted acid-basic ratios in our early work. This confirmed the excellent published data[2] and calculations on basic slags. This phase seems to be academic, since it is not a convenient control tool. However, slag appearance is most useful. Slag appearance may indicate to the operator whether the chill is rising due to carbon, or silicon when the carbon determination is delayed. The basicity and carbon seems to be directly related to the porosity and dullness or chalkiness of the slag fracture. The extent of gloss or glassiness directly indicated acidity or low carbon. Color does not seem to be a positive criterion.

It is important to use high-calcium limestone. High-magnesium or dolomitic stones were not effective in removing sulphur. In a basic-lined cupola sulphurs of from 0.10 to 0.13 per cent were obtained with dolomite. Switching to high-calcium stone reduced them to 0.03 to 0.05 per cent. Fluorspar has been used as a flux, but we are presently having success with a proprietary flux in brick form. This contains a large percentage of fluorspar fines or tailings in a binder. The bricks are easier to add and results appear consistent.

Limestone additions vary from 1 per cent to 7 per cent. Large amounts are used to neutralize sand when foundry droppings with a large percentage of sand are used in the charge. Amounts are derived empirically and are changed on the spot when required. Control seems easier with more slag volume. When we melt uncleaned sprue with more stone to neutralize the sand, metal control is easier than on runs of sand-free work using less stone.

Control

With acid cupola melting one metallurgical observer can easily control the iron composition, pour samples for testing and check and control raw materials, while melting more than 15 tons per hour. He cannot operate the same cupola alone with a basic operation. We found it necessary to analyze for carbon every 20 minutes. These analyses are taken at the spout. Any sudden wide fluctuations are minimized by a 6-ton holding ladle. The metallurgical observer must make more checks on the chilling tendencies of the iron and take care of variations by changing late ferrosilicon additions. A position of principal metallurgical observer was made for the coordination of effort. Thus three men are needed for basic gray iron control instead of one. Another man is needed to make and follow nodular iron.

Our charge constituents are scrap steel and silicon in the form of silvery pig plus returns. The returns are handled the same as for acid practice. Sprue is not cleaned. Foundry drop and spill is pulled out of waste sand with the yard magnet. It has a high percentage of sand. Steel scrap is regular foundry grade. The largest percentage consists

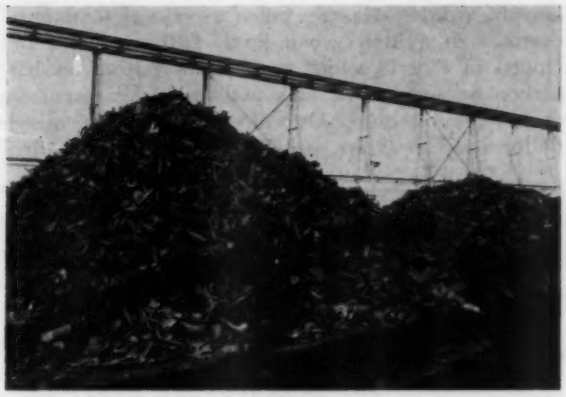


Fig. 9 — Steel scrap for basic-lined cupola operation.

of automotive frames, axles, housings, crankshafts, springs and gears. The remainder is structural, agricultural and miscellaneous iron.

A photograph of our scrap pile is shown in Fig. 9. We do segregate our spill, sprue and scrap castings. They are used individually with the steel, and limestone is increased to neutralize sand when needed. If a car of especially clean steel scrap arrives it is put aside to be used by itself. Steel from our plant is clean and kept separate. Clean steel is useful when carbons are low for some reason. As a safety factor, clean plant steel is used in the first five charges to eliminate any chance of explosive materials getting into our melting zone. Explosive materials have been found in our scrap.

Gray iron properties in the as-cast condition are determined primarily by chemical composition. Carbon is the most important element. Its control received the most attention.

In a basic cupola, carbon is affected by many factors. As a result, the scatter from a target analysis might be expected to be greater than for that found in an acid operation. Figure 10 illustrates the actual scatter in carbon analysis for one year of acid versus basic practice. Our acid aim was 3.40 0.10. and 3.50 0.10 for basic. The basic has a higher carbon to permit lower silicon levels to reduce loss and for nodular iron. The scatter in Fig. 10 if taken now would be less. Water-cooled tuyere plates, shown in Fig. 5, had many leaks. These lowered the car-

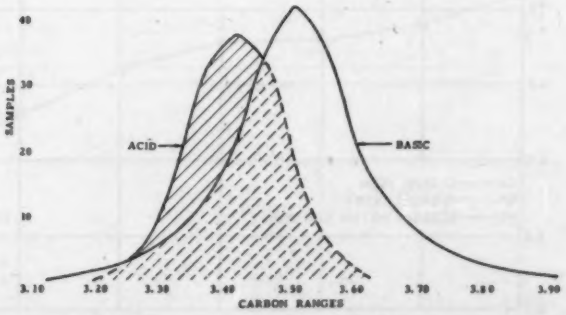


Fig. 10 — Scatter curve of carbon analysis for one year of acid vs basic operation. More basic analyses were made. Basic also includes all experimental results.

bon drastically. Measures taken to raise it sometimes overshot to a high-carbon level. Extremes were included in Fig. 10 which records all routine test-bar carbon analyses for a year with no deletions made even for good cause. Our physical level was raised deliberately as shown in Fig. 11 to insure quality.

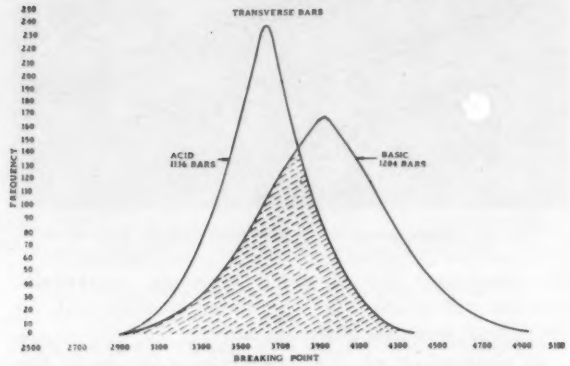


Fig. 11 — Scatter curve of transverse tests for one year of acid vs basic operation. Basic includes all experimental results also.

Carbon in a basic cupola is governed primarily by the composition and amount of slag. High carbons are obtained with very basic slags and low carbons with acid slags. The desired end effect is carbon control. As we make changes to control the carbon the slag is usually affected. Slag is of academic interest, so we will speak in terms of carbon control only.

We have found that our carbons are affected by limestone additions, charge cleanliness, blast volume, air preheat, coke amount, carbon in charge and moisture in the blast. It is felt that these factors also influence acid operations, but are emphasized by the sensitivity of a basic cupola.

We use an average of about 5 per cent limestone. Lesser amounts are used for clean charges, more for dirty charges.

Coke ratios vary throughout the year. As shown in Fig. 12, high summer humidity must be compensated for by coke. In the summer we operate the cupola with a coke ratio as low as 6 to 1, and as high as 7 to 1 in the winter. Our metallurgical observers know the theoretical amounts of coke to add for water in the blast by daily reference to AFS charts,[4] but are not bound by that figure. They shoot for 2800 F iron with a 3.50 per cent carbon. They increase or decrease coke when required, so the graph in Fig. 12 is influenced only by actual operating conditions.

Coke is raised for high temperatures. If the carbon is satisfactory we cut the limestone on raising coke. Changing the amount of coke also seems to affect the carbon in addition to the temperature.

Coke quality and size is important, changing either will directly affect practice. At present we are using 3¾ in. x 8 in. coke. Larger coke increases carbon and

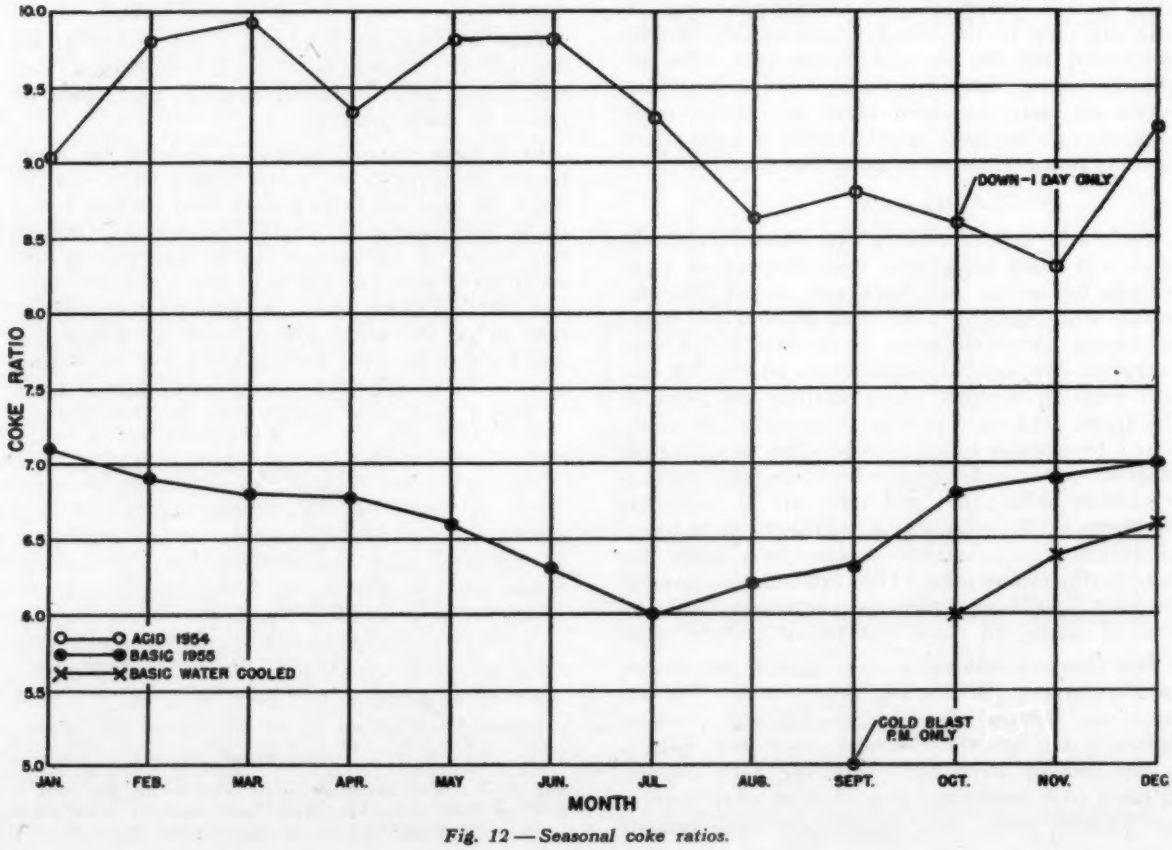


Fig. 12 — Seasonal coke ratios.

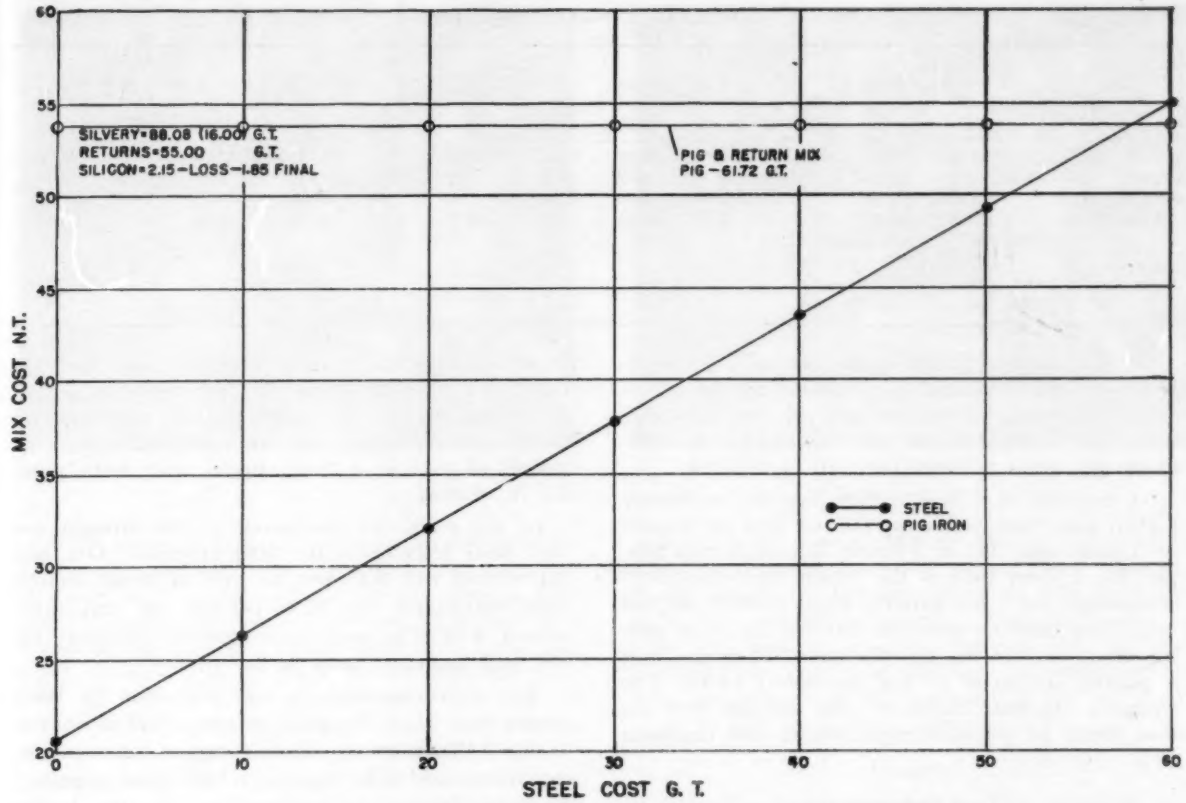


Fig. 13 — Effect of steel price on basic net metal cost.

temperature but is erratic. Smaller coke lowers carbon and temperature, but is more consistent. The larger sizes may be affected by screening methods.

Carbon may be varied by changing the amount in the charge. All conditions being equal, it is similar to acid practice.

Since limestone is neutralized by sand or rust, cleaner charges give higher carbons.

The above-mentioned methods of carbon control take effect only on reaching the melting zone or after about an hour. A rapid method of raising or lowering carbon is needed. We can do so by simply raising or lowering blast volume. In this way it is possible to compensate to some extent for irregularities in the slag composition. Following are some results of blast changes on carbon.

Time	Blast, Lb	Time Sample	Carbon, %
10:52	510	10:50	3.34
11:37	560	11:50	3.02
11:47	540		
11:52	520	12:00	3.10
11:56	500		
12:05	480	12:20	3.40
		12:30	3.44
		12:55	3.50
12:58	510	1:10	3.36
1:10	520	1:52	3.10
2:13	480	3:13	3.37
		2:35	3.37

Table 1 — Comparative Gray Iron Charge Costs
October 1955*

Raw Materials	Gross Ton	Cost Net Ton	Si	%	Acid — Cost Net Ton	Si	%	Basic and Basic Water-Cooled, Cost Net Ton	Si	%	Acid Low Phos. Cost Net Ton	Si
Pig	61.72	55.11	1.80	35	19.26	0.70						
Pig — Low Phos.	72.50	64.74	2.00							46	29.78	0.83
Silvery Pig	88.08	78.66	16.00	2.5	1.97	0.40	10	7.87	1.60	4	3.15	0.64
Steel Scrap*	43.00	38.39	0.10	15	5.76	0.02	64	24.57	0.06	20	7.68	0.02
Cast Scrap	48.00	42.86	2.00	21.5	9.21	0.43						
Returns**	55.00	49.11	2.00	26	12.77	0.52	26	12.77	0.49	30	14.73	0.57
					48.97	2.07		45.21	2.15		55.34	2.06
						0.10			0.30			0.10
					48.97	1.97		45.21	1.85		55.34	1.96

*Highest Steel Scrap Price Month in 1955.

**Arbitrary Constant Figure.

TABLE 2 — COMPARATIVE METAL COSTS PER NET TON

	Acid			Basic			Basic Water Cooled			Acid Low Phosphorus		
	1954	1955	Oct. 1955	1954	1955	Oct. 1955	1954	1955	Oct. 1955	1954	1955	Oct. 1955
Iron	46.86	47.52	48.97	37.06	39.02	45.21	37.06	39.02	45.21	52.79	53.40	55.34
Coke	2.85	2.85	2.85	4.02	4.02	4.02	4.42	4.42	4.42	2.85	2.85	2.85
Refractories	.45	.45	.45	2.40	2.40	2.40	.89	.89	.89	.45	.45	.45
Total	50.16	50.82	52.27	43.48	45.44	51.63	42.37	44.33	50.52	56.09	56.70	58.64

Steel 1954 — $28.71 Gross Ton
1955 — 32.16 Gross Ton
Oct. 1955 — 43.00 Gross Ton
Coke — 26.54 Net Ton

Calcium carbide added to the limestone is an extremely potent material as discussed in the literature.[2] Amounts approximating 1 per cent raise carbon. Even as little as 0.02 per cent or ½-lb to 2200-lb charges seems to lower the chill of the iron.

At the start of a heat, carbon may be excessively high if some basic refractory falls off into the cupola or if some other factor is wrong. This may also happen for a short time if the cupola operators overcompensate for a low carbon. High carbons are not usual, but must be provided for. We do so by adding clean steel to the transfer ladle. Then the iron is poured on top of it, and inoculants added. The wedge is checked. Extensive tests indicate that the iron meets all physical requirements and machines well.

Cost Comparisons

The basic cupola must produce quality irons at a competitive price to justify its existence. It is more expensive in some respects. Figure 11 discussed under *Control* clearly shows that more coke is needed in a basic cupola. About 9.3 lb of iron per pound of coke were melted in our acid-lined cupola. A pound of coke in a basic cupola melts only about 6.5 lb. of iron.

In the refractory discussion, it was brought out that basic refractories are more expensive. Our acid refractories cost $.45 per net ton of metal melted. Basic refractories cost $2.41 per net ton, and water-cooled, $.89. The water-cooled cupola refractory figure will probably drop in the future.

The above increases in cost are offset by lower metal costs when the price of scrap steel is normal. Table 1 illustrates the relative costs of mixes intended for an acid-lined cupola, a basic-lined cupola, a water-cooled and an acid-lined cupola iron having low phosphorus. The basic mix is cheaper than acid, even with relatively high steel prices.

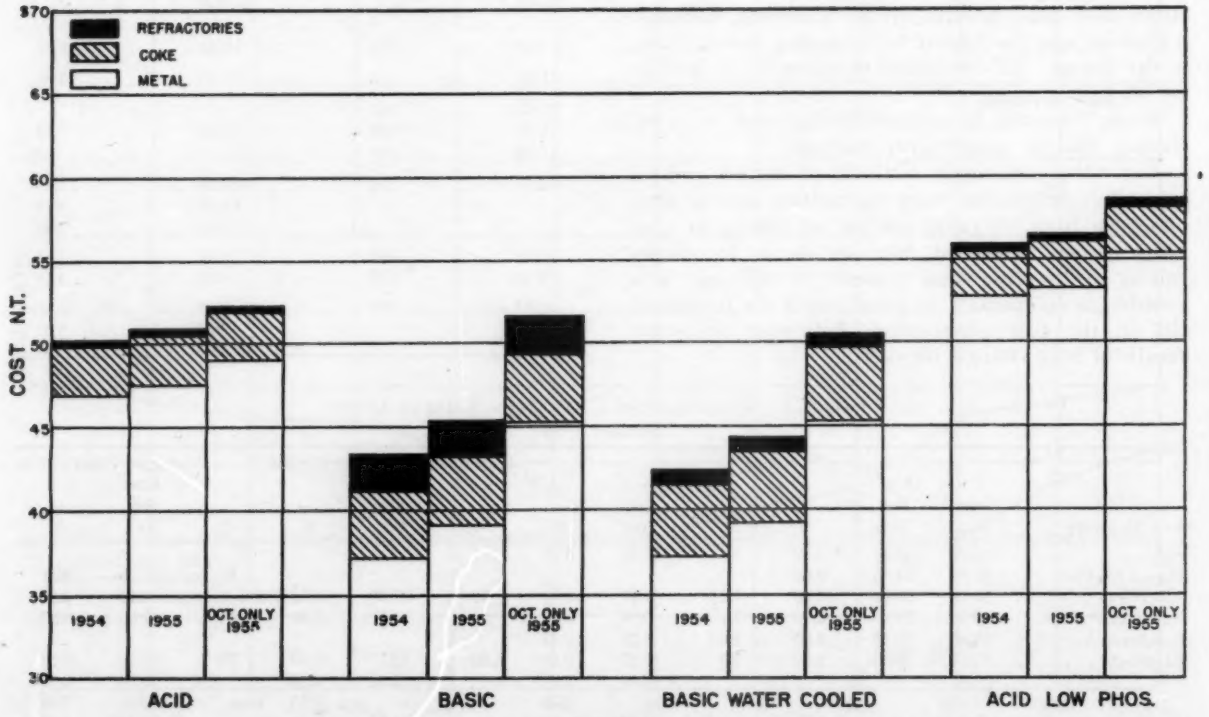


Fig. 14 — Comparative metal cost per net ton. (Scrap steel, $28.72 per gross ton, 1954 average; $32.16 per gross ton, 1955 average; $43.00 per gross ton, Oct. 1955, high; Memphis.)

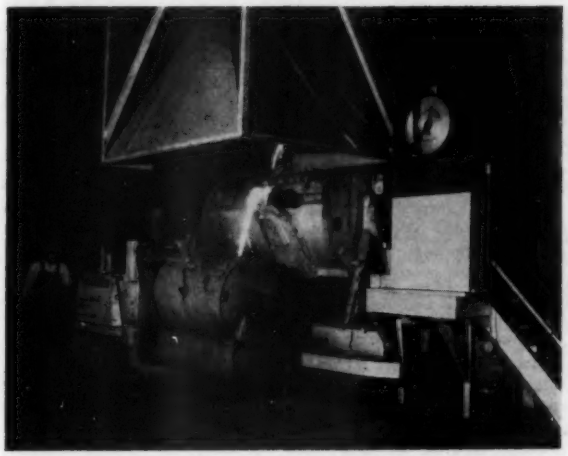


Fig. 15 — Pouring of basic gray iron. Note up position of hood.

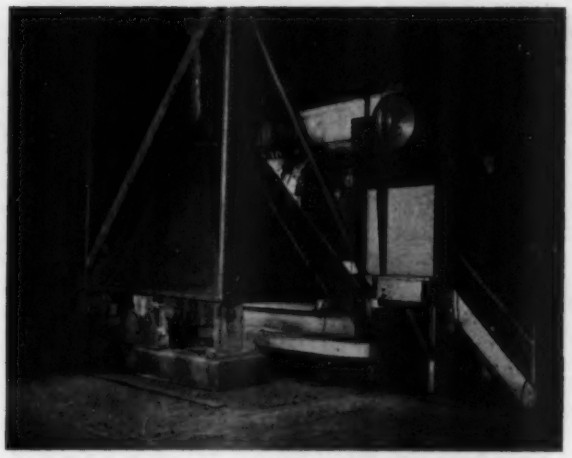


Fig. 16 — Pouring of nodular iron from same base iron as Fig. 15. Note hood down position.

Metal cost will depend on the price of scrap steel. Obviously scrap steel prices can rise to a point where it would be economical to use pig iron instead. Figure 13 graphically portrays the cost of a steel mix versus a pig iron mix. They meet at a gross ton price of scrap steel of about $58.50 when pig iron costs $61.72 per gross ton delivered. Mixtures of pig and steel, as normally used, would have a price somewhere between that of pig or steel alone.

Refractory, coke and metal costs must be combined to get a fair comparison of cost. We took the average cost of our steel scrap for 1954 of $28.71 per gross ton and for 1955 of $32.16. Then we took the cost of $43.00 per gross ton as the highest month in 1955. The mix cost for acid, basic, water-cooled and low-phosphorus acid irons are compared in Table 2. In the acid mix 35 per cent pig is used with 15 per cent steel. The basic mix has 64 per cent steel. The low phosphorus is 46 per cent low phosphorus pig. In addition our annual average costs for refractories and coke are listed in the proper category also. The totals give a true cost picture of the net metal cost. Figure 14 is a graphical comparison of these costs.

It should also be remembered that we now have three men instead of one controlling the iron. If a cupola is operated for only a few hours, it is possible two or even one man could do it. There are also other extras. We unload more coke and bulky steel. Our sprue has increased by about 2 to 3 per cent. Also, we blacken more cores because of burn in, which may be due to our higher carbon level. It is difficult to put all these extras into definite figures, but we do feel that they exist as an increase.

Nodular Iron

We have poured up to 40 tons of nodular iron in an 8-hour day. At the same time we poured larger tonnages of gray iron; this from the same holding ladle. Both irons completely met specifications. The nodular iron is made by the usual additions of magnesium alloys. Large daily tonnages would produce a nuisance of magnesium oxide smoke and blinding flashes. We devised a hydraulically lifted hood that is exhausted with a 42-in. diameter, 25,000 cfm fan to remove the smoke and cover the ladle. It is markedly successful. Figures 15 and 16 show this hood in position for gray iron and for nodular iron. Pouring of nodular is shown in Fig. 17. The effectiveness of this is shown in Fig. 18 which is an outside view of the exhaust stack.

Quality

An imperative requirement for any successful operation is quality. Under *Control* we discussed carbon analysis scatter. It was recognized that this would be greater. We lowered the silicon to offset it. The lower silicon and phosphorus raised our quality level. With an acid operation our 1.2-in. diam. 12-in. center transverse bar strengths grouped around 3700 pounds.

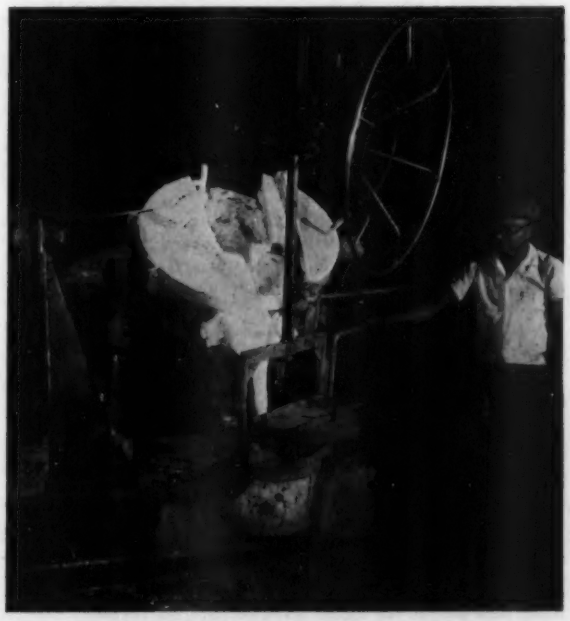


Fig. 17 — Pouring of nodular iron. Note lid on ladle to minimize splatter and reduce alloy consumption.



Fig. 18 — Outside view of exhaust stack during alloying for nodular iron as shown in Fig. 16.

Basic iron strengths were 4000 pounds. The annual scatter for acid and basic irons is shown in Fig. 11. More basic samples were taken, due to closer control.

We are quite certain that the quality of basic iron with the same carbon equivalent is equal to or better than that of acid iron. Many tensile specimens were compared. Least squares graphs were plotted, and they indicated that basic iron is stronger. However, we feel that our sample size was not large enough to be conclusive.

Acknowledgment

We wish to thank Mr. R. G. Mahal, our Works Manager, and other members of management for permission to publish this work, and their constant cooperative encouragement that made it possible.

Mr. Douglas W. Ferris, Quality Control Engineer, furnished the statistical charts and graphs.

Mr. F. Bridgewater of our Accounting Department checked our figures, and supplied cost data.

The research needed to assemble about three years of data and the drawings were done by A. F. Spengler, Jr., James Wiles, and Paul Clagett of the Laboratory.

We cannot name all the individuals in the foundry and laboratory who put forth much extra effort to make it possible to operate basic, but we are sincerely grateful. This would not have been possible without that extra effort.

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3. J. Sheley, Asst. Mgr., Black Clawson Co., Hamilton, Ohio.

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DISCUSSION

Chairman: W. R. JAESCHKE, Whiting Corp., Harvey, Ill.

Co-Chairman: J. P. HOLT, Basic Refractories, Inc., Clayton, Mo.

A. J. FRUCHTL[1]: What is the size of the tuyeres, and how many were used in the special water-cooled basic cupola?

MR. ALEXANDER: The first water-cooled tuyeres were 8½-in. in diameter. We found our melting rate slow. The temperature and carbon pick-up were also unsatisfactory. Next we tried 6-in. diameter tuyeres by merely restricting the 8½-in. tuyeres with an insert. These were so markedly superior that we replaced the 8½-in. diameter castings with 6-in. castings which we are presently using.

MR. FRUCHTL: What blast pressure was encountered in this operation?

MR. ALEXANDER: Our blast pressure will vary with the amount of blast used. Normal operations call for blast volumes of from 650 pounds to 800 pounds per minute. The pressure here is from 15 to 20 ounces. If demand is high a blast of 800 pounds or more may be used. Then the pressure is from 20 to 25 ounces.

MEMBER: Why do you prefer internal water cooler?

MR. ALEXANDER: We have no particular preference for internal water coolers. This method was adopted simply because we could not shutdown for extensive alterations. A few water-cooled pipes were placed in the cupola at a time during normal repairs. In that way we lost no production. It also appears that to externally cool our present cupola would require extensive costly engineering and alterations. Our operating cost should be equal to that of an externally-cooled cupola at present.

H. M. ASPINALL[2]: Have the authors experienced pronounced manganese sulphide segregation at gray iron castings surfaces (and particularly so toward the cope or top side of the castings?) This has been our experience in making castings from acid cupola practice. Actually we have not been able to produce any evidence to indicate that the service of the castings (ingot molds and other miscellaneous castings for steel plant use) suffers as a consequence of the segregation, however, we would like to know the experience of others on this subject.

MR. ALEXANDER: We have not noticed any manganese sulphide segregation in our gray iron castings. This should not be very likely in a basic iron since the manganese and sulphur levels are relatively low. However, in our early trials we had a few castings with kish, but this was due to obviously high carbon levels.

J. W. MOORE, JR.[3]: Our basic cupola experiences at ACIPCO indicate that a low sulfur iron (.03 to .05% S) melted at around 2800 F will generally contain considerably more carbon than 3.50 per cent. Some variations due to coke, flux, etc. are expected as well as differences in cupola size and design, but we would like to know more details as to how this is accomplished.

MR. ALEXANDER: Our sulfur level is slightly higher than 0.03 to 0.05 per cent. On the surface the difference 0.04 to 0.06 percent appears superficial. However, in practice we rarely get 0.03 per cent while we do get many 0.04 to 0.05 per cent figures. When we deliberately seek sulfur levels of 0.03 and under our carbon content also rises above 3.60 per cent.

We melt about 60 per cent steel scrap, 10 per cent silvery pig (6.00 Si) and the remainder consists of uncleaned sprue and scrap. Only high calcium stone is used to give us a basicity of about 1.80 $\frac{(CaO+MgO)}{SiO_2}$. We use only enough coke to keep our temperature at 2800 F. More will tend to raise carbons. If our carbon tends to reach and pass 3.60 our blast is raised. The carbon usually drops, then the blast is lowered. Persistent rising of carbon calls for reduced stone; coke or increased steel. The latter is true because in practice our steel may vary from 45 to 70 per cent depending on delivery of shop returns and other factors.

Carbon is checked every 20 minutes. A carbon determinator was placed in the foundry laboratory for this purpose. This permits us to make initial blast changes before variations can have much effect on iron in the 6-ton holding ladle.

S. F. CARTER[4]: Does the $0.89 per ton refractory cost for the water-cooled practice include the cost of cooling water and tuyere maintenance? Do the authors have costs of these two items?

1. Director of Research, James B. Clow & Sons, Coshocton, Ohio.

2. Metallurgist, Dominion Iron & Steel Co., Sydney, N. S., Canada.
3. Asst. Melting Supt., American Cast Iron Pipe Co., Birmingham, Ala.
4. Asst. Works Mgr., American Cast Iron Pipe Co., Birmingham, Ala.

MR. ALEXANDER: No, the $0.89 figure represents the actual cost of refractories placed into the cupola. As mentioned in the paper, this figure is being constantly reduced. Our cooling water cost was approximately $45.00 per day, and the tonnage melted was about 125 tons. However, it is planned to install a water tower and recirculate. This should make the cost of the water nominal. Our tuyere maintenance is also a dropping figure. At the start we repaired tuyeres almost daily. Now there is only occasional trouble, so a figure would be hard to obtain on this.

MR. CARTER: The two and three extra men as metallurgical observers needed to control the basic cupola seems to be a lot of manpower. Basic practice is admittedly more complicated and usually requires more diligence, but seldom requires added manpower. I wonder if the reason for the seriousness of control is the fact that a base iron is being melted that can be poured as gray iron or treated for nodular iron. This is evidently being done with success and some savings by such a practice. However, to have an iron suitable for both would narrow the margin for both irons.

MR. ALEXANDER: Manpower requirements to control iron may vary in different shops. In our shop a metallurgical observer is responsible for inspection of incoming raw material and its location in the yard. He must also keep a perpetual inventory against his disbursements. Then he submits charge compositions to the charging gang, which change throughout the day depending on materials available, etc. He checks the charges and the resulting iron after inoculation. At the end of the day, he submits and itemized report to the laboratory which in turn makes up reports to the accounting department. In acid melting he was alone, and had little free time.

In basic melting we now make checks on our carbon every 20 minutes. We also check our base iron chills and slag. A man was placed on this job to catch samples and analyze them. This accounts for one extra man.

Basic practice is more complicated. It is also subject to more variations, since the iron analysis and physicals are governed to a great extent by slag composition, which can easily vary. In the past, we checked every fourth or fifth ladle for chilling tendencies. Now we check almost every ladle. The extra work would keep our metallurgical observer tied down with little time to take care of the yard and other duties. We now have a man assisting him. This accounts for the second man on a basic operation.

There are also men assigned to nodular iron. We made as many as 90 transfer ladles of nodular iron in eight hours. Alloy preparation and checking certainly tie up a man.

It will be noted that three men are tied up on the gray iron. One is a metallurgical observer, and another is a laboratory assistant for the carbon control. We have made a position of principal metallurgical observer to coordinate the control of melting. This man also follows up the nodular iron to insure a smooth running operation.

We have operated for eight hours with only two men. It can be done. But they are so completely tied up with routine functions of basic melting that there is very little time to anticipate and eliminate possible wide control fluctuations.

MR. CARTER: We know basic cupolas cannot be expected to operate on lean coke ratios, especially when high steel charges are used. However, a 6 to 1 coke ratio sounds like a lot of coke for a cupola of this size with 450 to 600 F hot blast. The graph in Fig. 12 indicates slightly less coke (about 7 to 1) used with refractory linings. Were temperatures comparable so that this represents added coke needed to compensate for water cooling? What, exactly has been the effect of water cooling on iron temperature and coke usage?

MR. ALEXANDER: Water cooling has affected our metal temperature and coke usage. Coke consumption was greater with a refractory-lined basic cupola. We must have iron at spout temperatures of from 2800 F to 2830 F. This was no problem with a refractory-lined basic cupola. In fact our problem often was excess temperature. However, more coke seemed to be needed for carbon pick-up. We could and did get good temperatures but low carbon until we raised coke.

Temperatures of iron from the water-cooled cupola are kept at about 2800 F by using larger amounts of coke as shown in Fig. 12. We found this coke absolutely necessary. Although we meet the required temperatures, the iron is colder. We no longer get excess temperatures up to 2900 F. The bed must be more carefully burned in to assure a good start. Our temperatures drop more rapidly for shutdowns and are more difficult to bring up after starting.

USE OF EPOXY RESIN AS PATTERN MATERIAL

By

E. J. McAfee*

Plastic patterns have become important again with the introduction of the epoxy resins.

Phenolic resins, introduced during the early part of World War II, were given a thorough trial in pattern construction but were found wanting and foundries lost interest in them due to their lack of dimensional stability and their low impact strength.

The epoxy materials seem fairly certain to find a place economically between the low cost wood and the more expensive metal patterns. Considerable research is being continued by the government and by private industry.

The epoxy resins may be cast into molds made of gypsum, wood, or metal. They may be reinforced with glass cloth, or any basic metals or their alloys to increase impact strength. They will cure at normal room temperatures, thereby eliminating the need for special ovens or curing equipment.

The epoxies are applicable to most methods of pattern manufacture and combinations of materials. Up to now there has been no epoxy resin available and suitable for patterns that can be cast in much volume without an accompanying unpredictable and varying shrinkage. We soon reached the conclusion that when using the epoxy casting resins they must be confined to thicknesses of 3/8 in. or less. This meant the use of other materials in conjunction with the casting resins to give the patterns the required strength and stability.

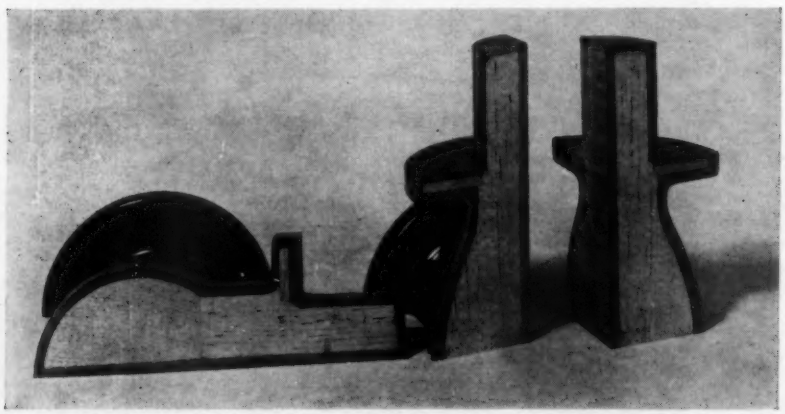
The Pattern Shop and the Materials Laboratory at the Puget Sound Naval Shipyard are now conducting a series of tests to develop and manufacture plastic foundry patterns that will justify their use both economically and technically. They do have their limitations, but when these are recognized and corrective measures taken in the manufacturing processes these limitations can well be overcome, thereby making some of the epoxy resins excellent pattern materials.

Several methods have been developed to successfully cast the various types of patterns and core-boxes.

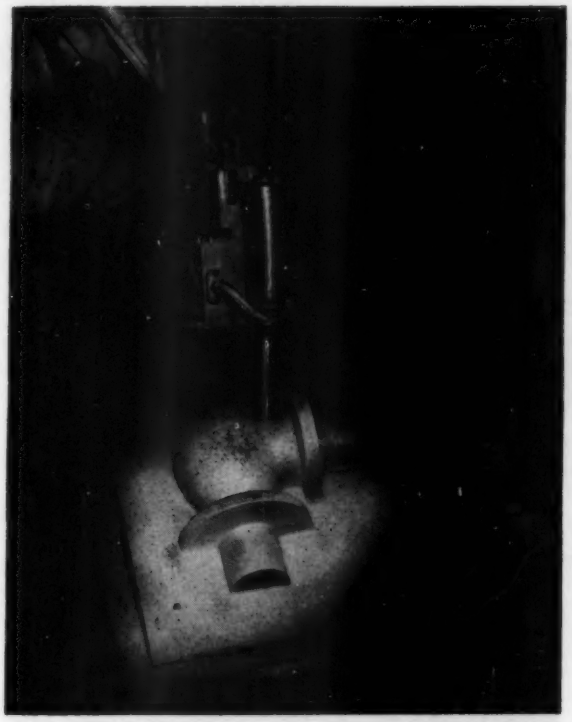
The **first method** of construction includes patterns with varying shapes and volumes but with limited dimensions of one or two feet in breadth, width, or diameter. The casting resin is poured around an inert core plug, which can be made of many materials. These include phenolic core foams, calcerite, other gypsum products, polystyrene beads, wood chips, vermiculite, and even sand and gravel.

Some manufacturers of epoxy resins have only just recently developed and put on the market new

*Master Patternmaker, Puget Sound Naval Shipyard, Bremerton, Wash.



Sectioned 2-in. globe valve pattern showing wooden core and ¼-in. epoxy surface thickness.



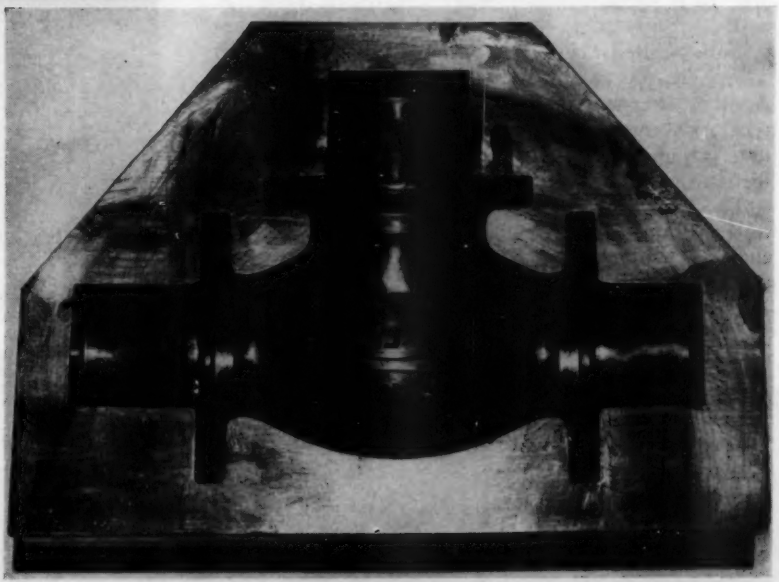
(Above)

Master pattern of 2-in. globe valve being sprayed with thin layer or coating of low-melting alloy as first step in making mold by the first method.



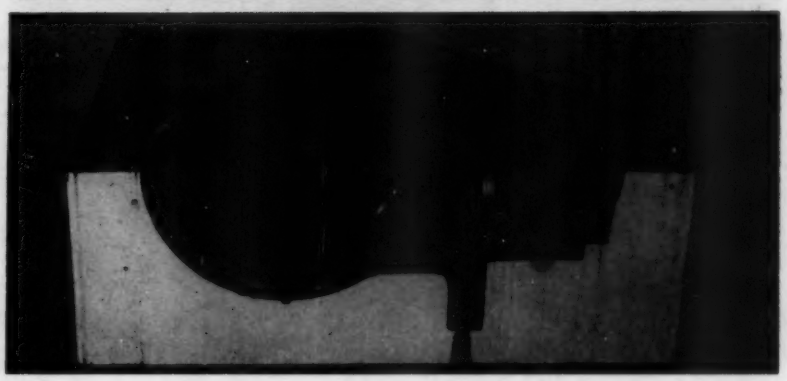
(Above)

Plaster mold of 2-in. globe valve pattern being poured with metal-coated pattern forming mold cavity.



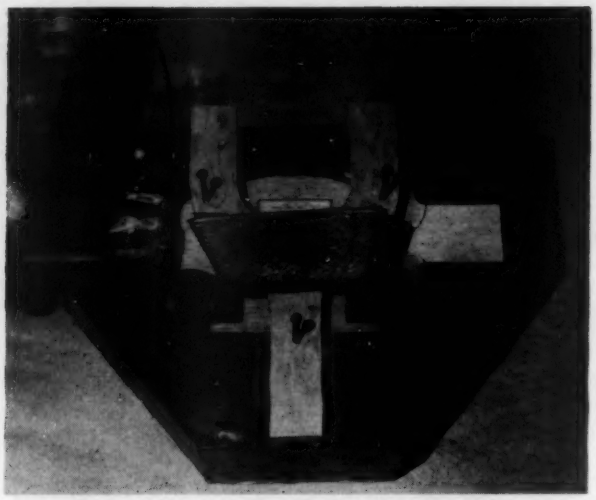
(Right)

Metal facing stays with plaster mold, making dehydration of mold unnecessary.



(Right)

Cross section of plaster mold which has been readied for casting plastic.



(Above)

Pine core is suspended in mold by bridge method. Wood screws supporting pine core are also used to withdraw pattern after casting and curing at room temperature. Pouring cycle being accomplished with pine core in proper position.

(Right)

Thin patterns such as plaque patterns, 3/8 to 5/8 in. thick, can be cast solid (by the second method).

formulas called "core resins" for this very purpose. These are too recent to have been fully evaluated. Another inexpensive inert type filler of a volcanic type rock was disappointing in several ways during tests. It absorbed moisture like a sponge and was difficult to machine.

The filler or core material we have found best up to this time for this method of construction is a good sugar or white pine, kiln dried to a moisture content of six or seven per cent.

For the first method, a master pattern is required with no extra allowance for contraction other than that necessary for the normal shrinkage of the final metal castings. The master is used to make the mold for the plastic pattern. The molds are usually made of plaster, and we prefer that they be faced with a spray of a low melting eutectic alloy. The master pattern may be of any convenient material; even an old pattern to be replaced could be used. For a parting agent between the master and the sprayed alloy, a thin coating of glycerine is rubbed on the master pattern, placed in a retaining frame, then sprayed with a thin coating of a low melting alloy. The metal thickness is not important and may vary from 1/64 to 1/32 in. Spraying any heavier is just a waste of time and material.

The gypsum plaster may then be poured, covering the metal-faced master to the height of the mold





Build-up of pattern in third method starts with brushing on resin layer, then laminating first coating.



Reinforcing second coating of die facing material with glass cloth.



First stage of filling mold with polystyrene beads used as a core resin.



Last stage of filling mold with polystyrene beads as a core.

frame. As soon as the plaster sets the mold can be turned over and the master can then be easily withdrawn. The mold surface should be polished with a fine grade of steel wool and crocus cloth. The mold is now ready for the application of the parting agent.

The advantage of using this eutectic alloy facing for the plaster molds is two-fold. First it makes it unnecessary to completely dehydrate the gypsum molds before casting the resins into them, and second, for additional castings.

The most effective parting agent we have found for this purpose (and, in fact, for all molds into which the epoxy resins are poured or applied) is a sealant manufactured from a copolymer vinyl, chloride-vinyl acetate sheet hand-shaved and cut with enough methyl ethyl ketone to produce a liquid of the consistency of a light syrup. This is a typical lacquer base. We suggest a colorant be added to make the application visible. It can be brushed or sprayed on the face of the mold. The spraying method is preferred as it can be applied more evenly.

After this has thoroughly dried, a coat of hard carnauba wax is applied and lightly polished. Over this is applied a coating of parting agent. The mold is then ready for the pouring of the plastic resin. We have found the epoxy casting resins with aluminum fillers to be the most adaptable for this type of pattern.

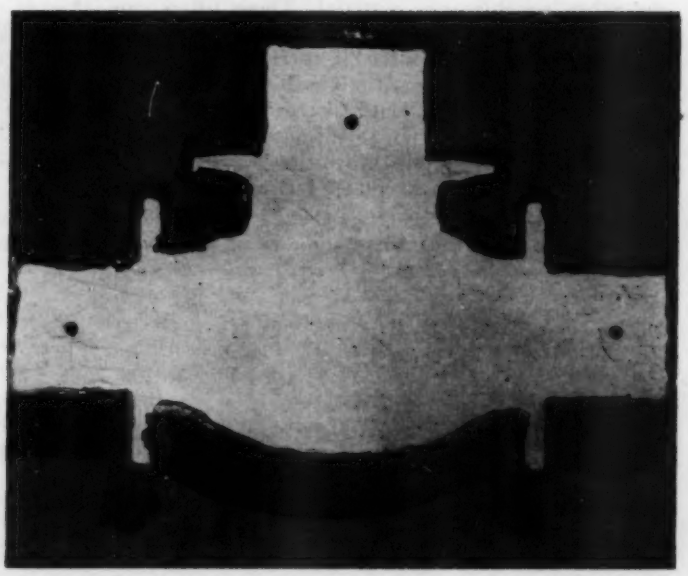
The second and most logical process is the casting of solid patterns entirely of the epoxy casting resin combined with an aluminum filler. This method is suitable for such types of patterns as plaques or flat patterns of a general uniformity of thickness up to $\frac{3}{8}$ in. and not exceeding $\frac{1}{2}$ or $\frac{5}{8}$ in. in spots.

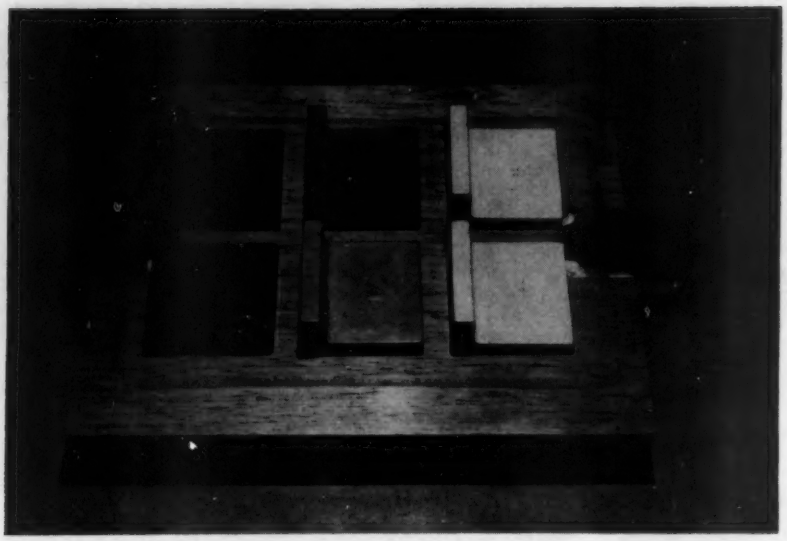
The procedure for making the mold for this type of pattern is the same as for method No. 1.

Excess material is milled from 2-in. globe valve pattern down to the parting line.



Master pattern for 2-in. globe valve built up by the die facing method and plaster of paris used as a core.





Test patterns mounted on plate were used to check ease of removal from sand. Each was different epoxy resin; one failed to release properly.

Method number three can be used to advantage on practically any size pattern, especially where the volume is great and/or there are very thin sections. It is used for the manufacture of dies and jigs. This consists of three coats of die facing epoxy resin material applied on the surface of the mold. The remaining cavity is filled with a backing material.

The same type of mold can be used as is prepared for the preceding two methods. The first layer of resin is a die facing resin applied by brush in a heavy layer. After this layer is partially cured but still tacky another layer is brushed on and small sections of glass cloth are then placed over the surface and overlapped, with the resin soaking through the fibers. A third layer is spread over the glass cloth before it has partly cured. This will give a mold surface of approximately $\frac{3}{16}$ in.

After this has thoroughly air cured, the cavity in the pattern is filled with a backing material consisting of polystyrene beads with a suitable hardener or any other suitable core material. These fillers can be poured a little above the parting line and milled off after curing with a router bit (preferably tungsten tipped) to the actual parting line. Any filler used should be of some inert material that will be dimensionally stable and will not shrink after casting. This is necessary to avoid any distortion of the facing resin. All necessary draw plates or inserts must be placed in the cavity before pouring.

The addition of the proper amount of anhydrous silicate to the casting resins, preferably to those with the aluminum filler, will reduce them to the viscosity necessary so they can be used successfully as a die facing material. This makes possible the use of the casting resin as an all purpose resin for the average pattern shop.

A plaster of paris filler having a low tensile strength should be poured up to within ¼ in. of the parting line, then covered with a facing of epoxy casting resin.

The most satisfactory filler or core material we have tested up to now is the polystyrene beads.

Various test patterns were used to evaluate the sand releasing qualities of various epoxy resins. Ell shape test patterns were used to test the vertical and flat lift. The synthetic sand used for testing is a type considered to be one of the most severe for this purpose. The mixture, given in parts by weight, was as follows:

new silica sand	95
western bentonite	5
silica flour	2.5
dextrine gums	0.3
cereal binder	0.7
perlite	0.6
fly ash	0.05
moisture content	4.6 per cent

These test patterns were left in the rammed up molds as long as 72 hours with excellent results.

ALUMINUM THERAPY IN SILICOSIS

By

J. W. G. Hannon, M.D.*

Use of aluminum therapy in the prevention and treatment of silicosis is based on the chemical theory of silicosis.

In the early part of this century it was thought that silica damaged the lungs by a mechanical process. The earlier workers believed that the physical presence of hard, sharp-pointed silica crystals in a lung that was moving almost constantly set up an irritation that resulted in scar tissue formation in the pulmonary tissues.

The mechanical theory of silicosis was accepted until the researches of Gye and Purdy[1] and later Kettle[2] supplied scientific information that cast doubt upon the accuracy of this theory. Gye and Purdy studied workers in the diamond grinding and carborundum industries and failed to find evidence of pulmonary fibrosis in the workers, even though the workers were exposed to hard, sharp particulate material that did not differ physically from silica particles. This study created doubts that silica caused pulmonary fibrosis by its physical presence in the lungs.

Later Gye and Purdy developed a special silica gel that was SiO_2 chemically but which physically lacked the presence of particulate matter. When animals were exposed to the silica gel they developed the typical lesions of silicosis without the presence of hard, sharp particulate matter in the pulmonary tissues. The researchers then postulated that it was the chemical action of silica in the animals' lungs that caused the fibrotic changes, rather than the physical or mechanical irritation by the silica.

Later Kettle studied the work of Gye and Purdy and put their new chemical theory to a test. He rendered silica particles chemically inert by placing a very thin coating of iron oxide around them. The silica particles still retained their hard, sharp points. When experimental animals were exposed to the treated silica particles, fibrosis did not develop. Kettle by means of the iron oxide had placed an interface between the silica particle and the tissue juices of the lungs, thus rendering the silica chemically inactive. This work confirmed the researches of Gye and Purdy and strengthened the chemical theory of silicosis.

It was a belief in the chemical theory of silicosis that stimulated Denny, Robson and Irwin[3,4] to efforts that resulted in the discovery of aluminum therapy.

Canadian experience had proven that dust control and medical control was not sufficient prevention to eradicate silicosis, and that some other aid, possibly a substance contained in the antidotal rock as described by Haldane was necessary.

The researches of Denny, Robson and Irwin that led to the discovery of aluminum therapy was followed by several years work to determine the efficiency and safety of the process. Their conclusions based on animal inhalation experiments over a period of many years are as follows:

1. Aluminum powder will prevent silicosis.
2. Aluminum powder is not toxic.
3. Aluminum powder will stabilize silicosis.
4. Aluminum powder has no effect on the mature silicotic nodule.

It was found that when animals were exposed to a mixture of 99 per cent silica and 1 per cent aluminum the animals did not develop silicosis. However, the animals exposed to silica alone did develop silicosis (Fig. 1).

The question then arose, does the aluminum powder have to be given as a mixture with the silica? Animals were dusted ten minutes a day with aluminum powder and then later in the same 24-hr period were dusted for 8-hr with silica dust. Another experiment was set up in which the above conditions were reversed, i.e. the animals received the silica exposure first and the aluminum later in the 24-hr. None of the animals in either study developed silicosis.

There was no evidence that aluminum was toxic, or would produce fibrosis in the lungs when animals were exposed to aluminum alone. Crombie and Blaisdell[5] confirmed this finding in humans by

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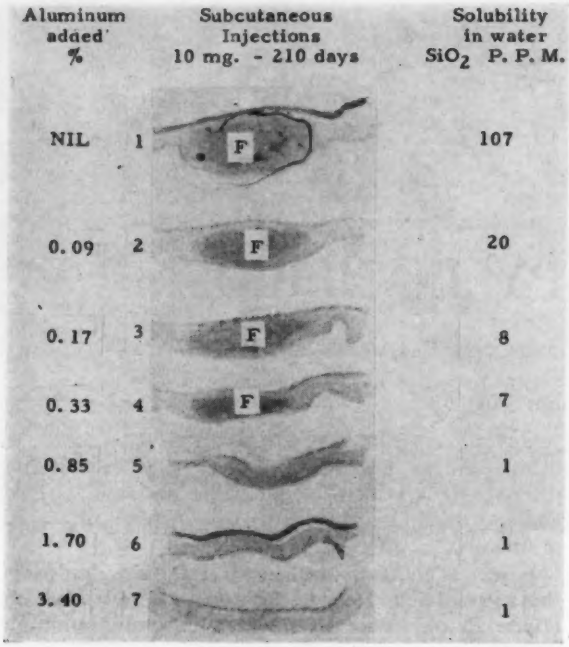


Fig. 1 — Subcutaneous injection of animal tissue with silica aline and with silica and aluminum. Fibrosis appears as F in 1, 2, 3 and 4; fibrosis absent in 5, 6 and 7.

studying men who had long exposures in a plant that manufactured aluminum powder for inks and paints. Clinical use of aluminum over 15 years has also confirmed this finding. Brown and Van Winkle[6] stated in the 1949 report of the American Medical Association that there was no evidence that aluminum was harmful to the normal or silicotic lung when inhaled in the dosages given for the therapy and prophylaxis of silicosis.

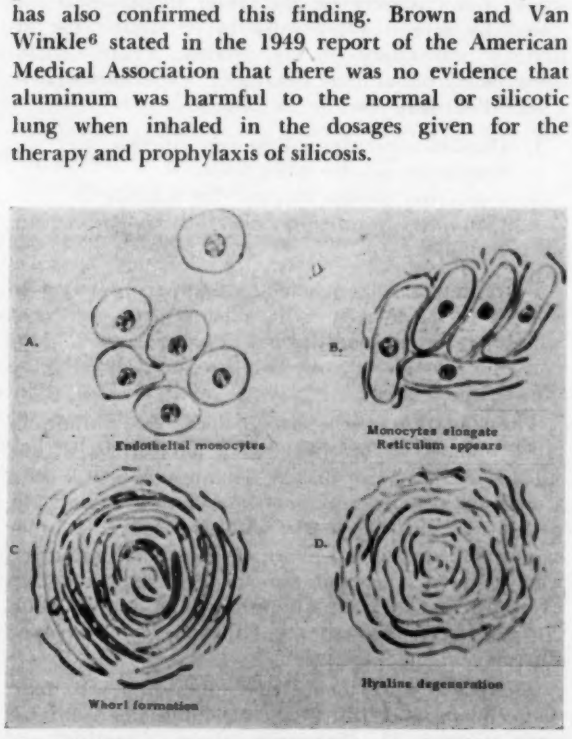


Fig. 2 — Tissue response. Stage A: Inert or foreign body response. Stage B: Intermediate or second stage in the formation of silicotic nodule. Stage C: Stage in the formation of the silicotic nodule. Stage D: The mature fibrotic nodule, stage 4.

Aluminum powder will stabilize silicosis and result in a regression of the immature stages of the formation of the silicotic nodule (Fig. 2). Denny, Robson and Irwin found that animals who had lesions midway between stage 2 and 3 in the formation of the silicotic nodule could expect regression of the process if aluminum powder was administered. Gardner,[7] et al and Dworski[8] also reported regression of the changes due to the acute inflammatory tissue response to silica.

Aluminum powder has no effect on the already established silicotic nodule and probably has no effect on stage 3 (Fig. 2) in the formation of the nodule.

To be effective the aluminum powder must be of optimum size and concentration in order that it can get down into the lungs and into the same cell that contains the silica. It is essential that there be an intimate intra-cellular mixing of the aluminum and the silica.

The modes of administration are:

1. Ball mill production of powder at the time of treatment.
2. Dry house dispersal.
3. Dispersion chamber.

In the ball mill type unit (Figs. 3 and 4) the powder is produced in the ball mill at the time of treatment. The mill consists essentially of an aluminum drum 14 in. long, 12 in. in diameter and extending through its horizontal axis is a 1-in. hollow

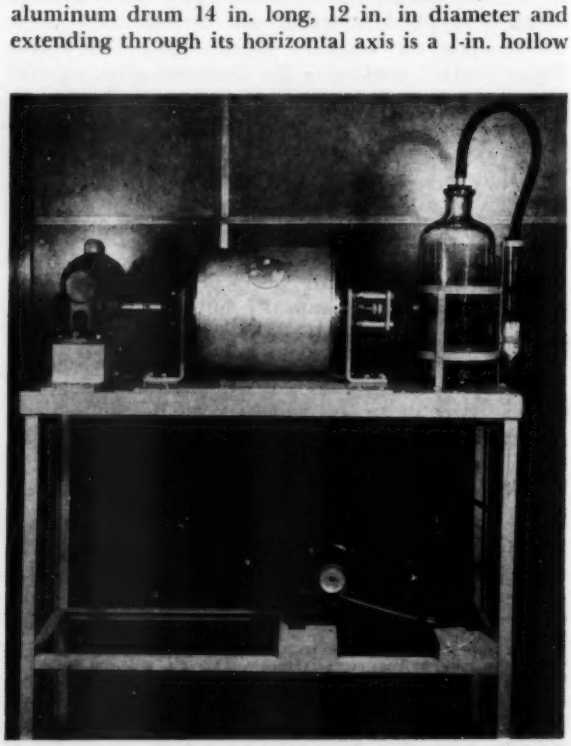


Fig. 3 — Aluminum powder manufacturing mill showing motor, drum, settling bottle, filter tube and pressovac. The mill is run for 20 min prior to starting treatments and during this time the powder produced is drawn off through the filter tube at the side of the large settling bottle by the pressovac which is mounted underneath the machine. The pressovac is stopped when the treatments are started.

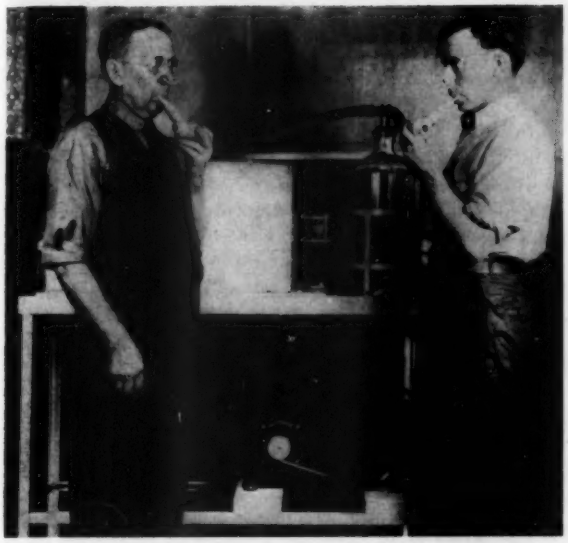


Fig. 4 — Aluminum powder treatment unit in actual operation. The men stand during treatment and wear a nose clip to prevent breathing through the nose.

shaft which is plugged in the center. On either side of the plug are forty $\frac{3}{32}$-in. holes. The drum is filled with 25 lb. of aluminum slugs and is rotated by motor at a speed of 72 rpm. The individual taking a treatment pulls the air through the holes on the shaft on the motor side of the mill in through the mill where it picks up the fine particles and out through the shaft to the settling bottle where the aggregates and large particles settle out. The air containing the aluminum powder is then pulled through the rubber tubes attached to the settling bottle through a one-way valve and is inhaled by the workmen. The mill has a production of 3.2 mg of aluminum powder per liter of air at an air flow of ten liters per minute. About 50 per cent of the powder in the air is impinged in the treatment apparatus before the air reaches the subject's air passages. The treatment dose is six minutes inhalation per week and 46 to 48 treatments per year for average exposures. Variation as to the length of treatment and frequency depend upon the environment and length and severity of exposure.

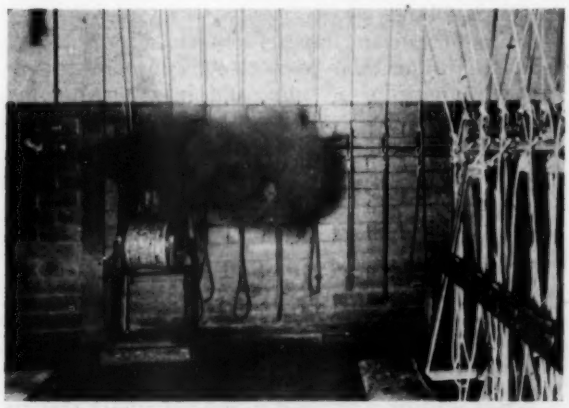


Fig. 5 — Dispersal of one, 10-gram can of aluminum powder. The dispersal valve is kept open for a period of one minute after dispersal in order to secure a thorough mixing of the powder with the room atmosphere.

The dry house dispersal (Fig. 5) method of treatment is used exclusively in mines in Canada, Mexico, and Western Australia. It is used to a lesser extent in this country as most installations are in industrial plants that do not have dry house facilities. This method of administering aluminum powder causes little or no interference with the normal routine of the workers. The powder is dispersed into the dry house either before or after the men go on shift, depending on local conditions. The men inhale the aluminum powder while either changing to street or working clothes and the duration of the treatment averages about ten minutes and is received daily for ten to twelve months per year. The dosage is 1 gram of powder per 1000 cu ft room space. This concentration is visible particularly near electric lights and the atmosphere compares to a poker room at about 11:00 p.m.

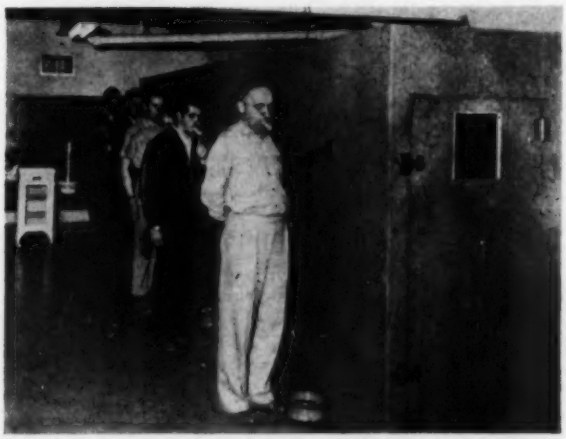


Fig. 6 — Dispersal chamber treatment. The powder is dispersed into the chamber and thoroughly mixed with the air. The men inhale the air from the chamber and thus secure adequate concentrations of aluminum in a suitable size distribution.

The dispersion chamber (Fig. 6) utilizes the principles of the dry house and the ball mill methods. The powder is dispersed into the chamber by the same method as in dry house practice but the men are outside the chamber and inhale the aluminum powder through tubes and valves such as in the ball mill method. The dispersion chamber offers advantages over the other two methods in industrial plants. The chamber can be modified in size to accommodate from six to thirty workmen at one time. The chamber can be mounted on a platform with wheels and transported throughout the plant to certain areas designated as treatment areas or centers. The frequency of treatments is about one-third of that of the dry house method. A recent improvement in the aluminum powder (McIntyre powder) has increased the efficiency of the treatment many fold and a substantial reduction in the duration and frequency of treatments can be expected.

The results of experience with aluminum therapy in humans have been accumulating and are following a favorable and predicted trend.

The use of aluminum therapy in the control of silicosis in a foundry was reported by Osmond[9] in 1955 as follows:

How Studied

Changes in the Functional Lung Capacity of Foundry Chippers (62 Workmen) Subjected to period of Aluminum Therapy — 1948-1954

Year	Lung Function Status	Men, No.
1948	Normal	34
1954	Unaltered	31
	Worse	3
1948	Subnormal	28
1954	Unaltered	0
	Aggravated	3
	Improved*	25

*Of the 25 who improved, 22 were normal at the time of the period of assessment in 1954.

Changes in the Functional Lung Capacity of Foundry Molders (39 Workmen) Subjected to Period of Aluminum Therapy — 1948-1954

Year	Lung Functional Status	Men, No.
1948	Normal	24
1954	Unchanged	22
	Aggravated	2
1948	Subnormal	15
1954	Unchanged	1
	Deteriorated	2
	Improved*	12

*Of the 12 who improved, 11 were normal at the time of final assessment in 1954.

Changes in the Functional Lung Capacity of Foundry Miscellaneous Group (188 Workmen) Subjected to Period of Aluminum — 1948-1954

Year	Lung Functional Status	Men, No.
1948	Normal	106
1954	Unaltered	103
	Worse	3
1948	Subnormal	82
1954	Aggravated	6
	Improved*	76

*Of the 76 who improved, 64 were normal at the time of final assessment in 1954.

Changes in the Functional Lung Capacity of All Foundry Workmen (289 Workmen) Subjected to Period of Aluminum Therapy — 1948-1954

Year	Lung Functional Status	Men, No.
1948	Normal	164
1954	Unaltered	156
	Worse	8
1948	Subnormal	125
1954	Unimproved	1
	Aggravated	11
	Improved*	113

*Of the 113 who improved, 97 were normal at the time of final assessment in 1954.

Osmond in his conclusions stated the following:

Of the 164 men commencing aluminum exposures with normal functional status, 95 per cent were unchanged and 5 per cent had deteriorated at the end of six years.

Of the 125 men starting with a measurable degree of functional incapacity before aluminum treatment was instituted, 9 per cent deteriorated further, while 90 per cent improved, and of the latter group 77 per cent reached normality at the end of this period.

The incidence of new cases of radiological silicosis dropped to zero during the years 1953 and 1954, which is the first time this has happened in the history of this foundry.

The use of aluminum in the Ceramic Division of the Sanitary Industry was reported by Perry[10] in 1955 as follows:

How Studied

Of a total of 139 men who received aluminum therapy for a period of 10 years, the results were as follows:

		No. Cases
(a)	Symptomatic Response	73
	Unaltered	25
	Improved	45
	Worse	3
(b)	Radiological Response	139
	Unchanged	117
	Improved	20
	Progressed	2

Thus is appears that aluminum has a markedly beneficial effect in suppressing symptoms and in arresting radiographic advance of the lesions of the men subjected to study. For instance, of 73 men who had symptoms at the start of their course of therapy, 62 per cent were improved at the end of 10 years, while only 34 per cent showed no amelioration and 4 per cent became worse in spite of therapy.

Radiologically the following results were obtained: Of 139 men who started aluminum therapy in 1944, 84 per cent remained unchanged, 14 per cent improved, while 2 per cent progressed.

Perry summarized his experience with aluminum therapy as follows:

Aluminum therapy appears to have an indisputably beneficial influence in suppressing symptoms, in arresting radiological advance, and in improving the functional reserve of ceramic workers.

The criticisms that have been aimed at aluminum therapy may be listed as follows:

1. It will be used as a substitute for dust control.
2. Aluminum therapy may be harmful.
3. Aluminum therapy may have an adverse effect on the resistance to tuberculosis.
4. The effect of aluminum therapy is psychological.
5. The use of aluminum is an admission of an existing silica hazard.

Before a plant is licensed to use aluminum therapy it must have a dust survey made by the State Department of Industrial Hygiene or a competent agent or agency and all recommendations as to dust

control followed out before the plant is given approval for its dust control program. A plant does not necessarily need to be dust proof before approval is granted. In this way it has endeavored to keep aluminum from being used as a substitute for dust control.

The criticism that aluminum might be harmful persisted for some time but Brown and Van Winkle[6] stated that aluminum therapy is apparently not harmful to the normal or silicotic human lung in the dosages given for the therapy and prophylaxis of silicosis. The experiences of the users of aluminum therapy in this country and many other countries have failed to produce any evidence that aluminum can be harmful to the normal or silicotic lung.

The feeling that aluminum therapy might alter the susceptibility to tuberculosis has been brought up but at the present time there is no scientific evidence published or otherwise that proves or infers that the use of aluminum would alter the susceptibility to tuberculosis. Dworski[8] and Schepers[11] have stated that when animals infected with tuberculosis are dusted with aluminum eight hours a day there is a transitory stimulation of the tubercular process which is followed by healing. When animals are exposed to aluminum one-half hour per day there is no alteration in the tubercular process.

The statement has been made that the improvement of symptoms in silicotics who have been treated by aluminum therapy is psychologic. Studies have been made over a period of years in which objective measurements of pulmonary capacity have been made and the improvement in symptoms parallels the improvement in these functional studies. There have been individuals and there always will be individuals who get a psychologic stimulation from any type of therapy. These individuals are in the minority and the psychological effect is short lived. Many men have been treated who were on compensation and have gone back to work and have worked many years afterwards. They were happy at work and have proven to be useful citizens. We feel that the use of aluminum can be followed by a definite physiological improvement in the lung capacities. This improvement can be measured, and when this improvement persists over a period of many years we believe that it is an actual and true response.

It has been stated that the use of aluminum is an admission of a silicosis hazard in the plant. This statement we believe to be true and the use of aluminum therapy should not be advised in a plant that did not have a silicosis hazard.

Aluminum therapy is not a substitute for dust control nor is it a substitute for a medical control program. Its main use is to act as a neutralizer for that amount of silica dust that cannot be removed from the air by a practical dust control program compatible with the economics of the specific plant. The consensus of the users of aluminum is that new cases are not developing in the treated groups. Individuals with already established silicosis are not progressing except in extreme rare instances and the beneficial effects obtained in the improvement of symptoms is lasting too long to be of psychologic origin.

The writer would like to propose a new definition for silicosis, one which is a practical, workable definition and one which is an accurate statement of facts. The definition is as follows:

Silicosis is a condition of the lungs caused by the prolonged inhalation of silicon dioxide and is characterized anatomically by nodular fibrotic lesions of the lungs that can be demonstrated by suitable x-ray films and in pathological specimens. Silicosis occurs in many stages of severity and in the more severe forms it is capable of producing cough, shortness of breath, tightness in the chest, an increased susceptibility to pulmonary diseases and a lessened capacity for work.

Conclusion

1. Aluminum therapy can and does prevent silicosis in experimental animals.
2 Aluminum therapy is not harmful in the dosages prescribed.
3. Aluminum therapy does not alter the native susceptibility to tuberculosis.
4. Aluminum therapy is useful in the prevention of silicosis in humans.
5. The use of aluminum therapy in humans with already established silicosis and symptoms is followed by an amelioration of symptoms in a significant percentage of cases.
6. The rapid type of silicosis is particularly responsive to aluminum therapy.
7. Aluminum therapy can be considered as an adjunct to our present methods of dust control and medical control in combating silicosis.
8. Aluminum therapy is not a substitute for dust control.

Questions and Answers

1. Q. Has aluminum been used in the mining industry?
   A. Aluminum is used in the mining industries of Canada, Mexico and Western Australia. It was first used in the mines of Canada in 1943. Its use was initiated in Mexico in 1945 and Australia in 1950 and at the present time there have been 64,826,000 treatments given.
2. Q. What kind of aluminum is used?
   A. Aluminum of a high degree of purity is ground in ball mills into an extremely fine state of division. 90 per cent of the particles will be below 1.5 microns, each particle consisting of a nucleus of metallic aluminum surrounded by an active oxide. Chemically, the aluminum powder is approximately 15 per cent aluminum and 85 per cent oxide.
3. Q. Would workers in an aluminum foundry have less chance of developing silicosis than in another type of foundry?
   A. As aluminum is a soft metal it does not tend to fragment into an extremely fine state of division during foundry processing operations. The atmosphere of an aluminum foundry is free from aluminum dust in respirable sizes.

4. Q. What is the acceptable limit of dustiness required in order to obtain a license to use aluminum therapy?

   A. State Departments of Industrial Hygiene set the standards on the requirements for dust control. We realize that it is impossible to obtain a dust-free plant but feel that an industry or plant should have a practical dust control program that is compatible with the economics of the plant.

5. Q. In the case of a compound exposure of silica and iron, would iron act as a neutralizer of the silica dust?

   A. We believe that iron oxide has a very slight neutralizing effect upon silica dust. If the iron dust occurs in sufficient concentrations it can modify the action of the silica by acting as a chemical diluent.

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THE NUCLEATION OF GRAPHITE DURING DECOMPOSITION OF CEMENTITE

By

E. G. Haney* and M. F. Hawkes**

ABSTRACT

White cast iron was deformed by rolling, shot peening, and twisting in torsion prior to graphitizing; the graphitizing was carried out isothermally in the temperature range of first stage malleablizing and the number of nodules of graphite per unit area was counted after reaction. In all irons where plastic deformation had taken place an increase in the number of nodules was observed.

The dependency of the number of nodules formed on the rolling temperature, per cent reduction in thickness, and reaction temperature was measured. The influence of the rate of heating to reaction temperature was investigated. The number of nodules formed in the rolled specimens is 5-15 times that of specimens not preheated; the number counted is essentially independent of the reaction temperature.

The number of nodules formed is directly related to the amount of prior plastic deformation but is inversely related to the temperature at which the white iron was previously deformed. Shot peened white iron shows an even greater increase in the number of nodules formed than does the graphitization of martensite. It was possible to "anneal out" the effect of the plastic deformation on the number of nodules formed in the torsion specimens.

Primarily as a consequence of the results of the torsion experiments, residual stresses are proposed as an explanation of the variations in the number of nodules counted in deformed irons. Residual stresses can be used to explain most of the variations in nucleating behavior of graphite in white iron.

Introduction

Decomposition of cementite (Fe_3C) in iron or steel may be represented by the following equations depending upon the temperature of the reaction:

$$\text{Cementite} \rightarrow \text{Austenite} + \text{Graphite} \quad \text{(I)}$$
$$\text{Cementite} \rightarrow \text{Ferrite} + \text{Graphite} \quad \text{(II)}$$

Graphite and austenite are the stable phases[1] from the eutectoid temperature to the eutectic temperature; at temperatures below the eutectoid reaction, ferrite and graphite are the stable phases. Reaction (I) represents the so-called "gamma-range graphitization" in all cast irons and it is the reaction taking place during the first stage annealing of white iron castings during the process of producing malleable iron. Reaction (II) is responsible for the graphite formed in some steels at subcritical temperatures.

The kinetics of the decomposition of cementite is generally considered a nucleation and growth process and has been studied as such.[2,3] It is during the first stage reaction that almost all of the nodules of graphite are nucleated in the malleable iron process. Subsequent graphite formed during the malleablization cycle generally deposits upon the existing nodules. The total time to complete the reaction is inversely related to the number of nodules formed. The condition of a given white iron casting just prior to the first stage annealing treatment has a profound effect upon the nucleation of the graphite. The number of nodules formed per unit volume is known to vary by several orders of magnitude in a given iron depending upon certain pretreatments.

The following comparisons of these pretreatments and their effect on the number of graphite nodules after the completion of the first stage reaction is given as the increased number compared with the untreated cast material having the same gamma-range treatment. The counts are made on a plane of polish and reported as the number of nodules per unit area. The formation of martensite (prequenching effect) in the structure of white irons, either as-cast or austenitized and quenched, shows the greatest increase in the number of nodules, about 8-400 times.[4,5,6]

A small increase in the number of nodules has sometimes been observed when white irons are rolled[7,8] or otherwise plastically deformed.[9] Variations in the heating rate to the graphitizing temperature can increase the number[10,11,12,13] by 2-15 times. Isothermally holding the white iron at certain subcritical temperatures (prebaking effect), most notably around 660 F (350 C) to 930 F (500 C), for 12 hours or more will give increases up to 4 times.[10,14] These are the orders of magnitude that the various factors

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have been reported to affect the number of nodules of graphite formed, hence, rate of nucleation and overall reaction rate.[6,15]

In the present paper the dependence of the nucleation of graphite in white irons on the amount of plastic deformation prior to first stage reaction, on temperature of deformation, on reaction temperature, and on the heating rate is described. It is demonstrated that the effect of plastic deformation on the number of nodules formed can be "annealed out". The experiments have led to an explanation of graphite nucleation in terms of residual elastic microstresses.

Experimental Procedure

Approach

Graphitization of martensite creates a larger number of visible nodules and hence affects the nucleation of graphite in iron alloys to a greater extent than any of the other various factors influencing the nucleation. Also, the unique effect of a martensitic structure in promoting graphite nucleation is significant when compared with a bainitic or a pearlitic structure. Hence, the prequenching effect presents itself as the first step toward separation of the factors influencing graphite nucleation.

That the condition of the alloy at room temperature before heating to graphitizing temperatures is most important has been recognized by those who have given explanations of the prequenching effect. If there is martensite present there is also a high level of residual stresses present as has been pointed out by Hultgren and Edstrom.[4] Likewise, microscopic cracks may be present. Bunin and Danil'chenko[16,17] consider such cracks and voids as the nucleation sites. On heating, the inevitable tempering of the martensite takes place so that the carbide-interface is greatly increased.

Laurent and Ferry[5] conclude that the increased graphitization rate is due to the increase in carbide-interface, since the nucleation of graphite was observed to take place wthin the tempered martensite. Some investigators[4,18] have suggested that the graphite nuclei could form directly from the supersaturated martensite during the early stages of tempering.

If it is possible to obtain the number of graphite nodules that accompanies graphitization of martensite *without* the formation of martensite, then martensite itself may not be the necessary condition. Bainite is characterized by a fine dispersion of carbides somewhat similar to tempered martensite; yet it is not effective as a graphite nucleating agent, contradictory to carbide-interface ideas.

The creation of a large magnitude of residual stresses prior to graphitizing but *without* martensite would seem the logical approach; cold working prior to malleablizing would be the obvious experimental procedure. Indeed, in some steels Austin and coworkers[19,20] have demonstrated that cold work will increase the number of nodules of graphite subsequently formed. The formation of graphite in certain hypereutectoid carbon steel wire has been considered to be the result of a specific kind of residual stress by Laszlo.[21]

If residual stresses have an active role in the formation of graphite nuclei, then it should be possible to relieve the residual stresses by suitable heat treatment and obtain a decreasing number of nodules as a result. The final set of experiments, the torsion work, show that this is possible.

Materials

The compositions of the two white cast irons used for studies of the gamma-range reaction in this investigation are given in Table 1.

TABLE 1 — COMPOSITION OF ALLOYS

	% C	% Si	% Mn	% S	% Ni	% Cu	% Mg
Iron E	2.33	0.89	0.29	0.024	0.93	0.53	0.07
Iron H	2.35	0.97	0.05	0.090	—	—	—

Both irons were from induction melted experimental heats, iron E was a 75-lb heat, iron H was a quarter of a 200-lb split heat. Iron E is a magnesium-treated hypo-eutectic iron; it was not inoculated but was superheated prior to casting into graphite molds thus minimizing the chance of any graphite forming. Iron H is the same iron used by Brown and Hawkes[2] and was cast in sand molds; it is essentially a ternary alloy with a Mn/S ratio of about unity.

Method and Procedure

A systematic study was made of the graphitization characteristics of iron H rolled to various reductions in thickness at five different rolling temperatures; specimens representing each reduction at all five rolling temperatures were graphitized at four different gamma-range temperatures, 1560 F (850 C), 1650 F (900 C), 1740 F (950 C), and 1885 F (1030 C).

The cast coupons (about $\frac{3}{8}$ x $\frac{3}{4}$ x 3 in.) were cut in half longitudinally to obtain a $\frac{3}{8}$ in. square cross section and one-half of their length was reduced in thickness by machining so that two reductions per specimen could be obtained. Since there were four reductions for a given rolling temperature, all specimens deformed at one temperature came from one finger. The specimens were reduced to thicknesses from 0.25 to 0.05 in. in a mill with 10-in. diameter rolls.

Specimens were soaked 10 minutes prior to the first pass to obtain the temperatures used for rolling, namely, 1110 F (600 C), 1200 F (650 C), 1290 F (700 C), 1380 F (750 C), and 1515 F (825 C). They were re-heated after each pass and none remained at rolling temperature more than 30 minutes. The average reduction per pass was about 10-20 per cent. Using temperatures of 1290 F (700 C) or above, a reduction in thickness of 80-85 per cent was possible with only edge cracking occurring; with lower temperatures the maximum reductions in thickness of about 60 per cent resulted in severe cracking of the specimens.

After completion of rolling all specimens were air cooled to room temperature. Each specimen of a given reduction in thickness was cut into four pieces — one to be reacted at each of the four reaction temperatures. All pieces reacted to a given temperature were treated together. These included two samples

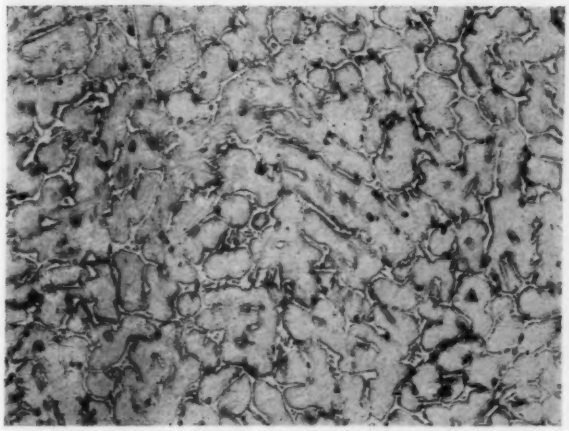


Fig. 1 — Typical microstructure of iron H as-cast showing pearlite and eutectic cementite. Nital etch. Mag. — 100X.

not rolled but having the same thermal treatment as the specimens rolled at 1110 F (600 C) and 1515 F (825 C).

Shot peening made the deformation of white iron at room temperature possible; also, the deformation near the surface is very severe, with no deformation taking place nearer the center of the specimen. A comparison is thus obtained in the given specimen.

The torsion method of deforming white iron has several advantages over deforming by other methods. It affords a much greater degree of plastic deformation at subcritical temperatures than does bending or tension. Better temperature control is possible than could be exercised during the rolling experiments. There is always a comparison in each torsion specimen of the actual increase in number of nodules formed as a result of the deformation with the undeformed metal (along the neutral stress axis). All torsion specimens used were ¼-in. square-ground and 3 in. long.

Results on iron E only are reported, although iron H exhibited similar results. The specimens were clamped at each end in a yoke arrangement. The twist was applied by a rod clamped at the center of the specimen so that two 1-in. gage length samples were obtained per specimen; two fractures are obtained per specimen. All deforming was done with the specimen and the end of the yoke suspended in a salt pot at temperatures of 1110 F (600 C) and up. The salt bath protected the torsion specimen from oxidation and helped prevent decarburization.

The specimens were soaked 15 minutes before twisting to breaking; after breaking the rod was sometimes removed from the bath at a later time than the yoke. This made a check possible on the effect of a stress-relieving anneal employing the same temperature used for twisting. Also, a check on the nodule count is possible since there were two halves of the fractured samples with the yoke and two halves with the rod. All samples were graphitized to completion at 1650 F (950 C).

All samples to be graphitized were heated to the reaction temperature at 215-270 F (120-150 C) per hour in an atmosphere of dried argon, allowed to react to completion or nearly so, and then water-quenched. The only departure from this thermal treatment was the samples of iron E used to check the effect of the rate of heating to gamma-range graphitizing temperatures on the number of nodules formed. Samples were heated at the constant rates of 25 F (15 C) and 270 F (150 C); others were inserted in a furnace in an atmosphere of argon at the 1650 F (900 C) reaction temperature. The most rapid heating rate was the result of plunging the samples in a salt pot at 1650 F (900 C).

All graphitized samples were given a metallographic polish. The plane of polish was used to count the number of nodules per unit area. One square millimeter of each sample reported was counted; this number per square millimeter represents a so-called count. No particles smaller than 2.0×10^{-3} mm in diameter could be recognized as graphite; they are not represented in the counts. All counts of rolled specimens were made by traversing the center of the transverse section of the sample. As for the graphitized torsion specimens, the nodule counts were made on a plane of polish through the neutral axis.

Results

Effect of Rolling Temperature, Degree of Deformation, and Reaction Temperature

The cast structure of iron H is compared with the as-rolled iron in Figs. 1 and 2 which show how rolling at 1515 F (825 C) breaks up the eutectic cementite network. The same two samples reacted to graphite are compared in Figs. 3 and 4; note the increase in number of nodules and decrease in nodule size resulting from the rolling prior to complete graphitization. Figure 5 shows the even greater increase in number of nodules when the white iron has been rolled at 1200 F (650 C). There was little observable difference in the graphitizing characteristics from the surface to the center of the rolled specimens.

Other photomicrographs (Figs. 6 to 11) at higher

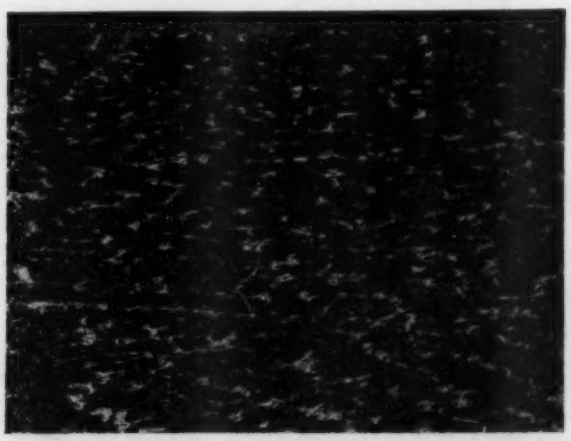


Fig. 2 — Microstructure of iron H rolled to 81 per cent reduction in thickness at 1515 F (825 C) and cooled to room temperature. Longitudinal section showing surface of specimen at top; eutectic cementite in a pearlitic matrix. Nital etch. Mag. — 100X.

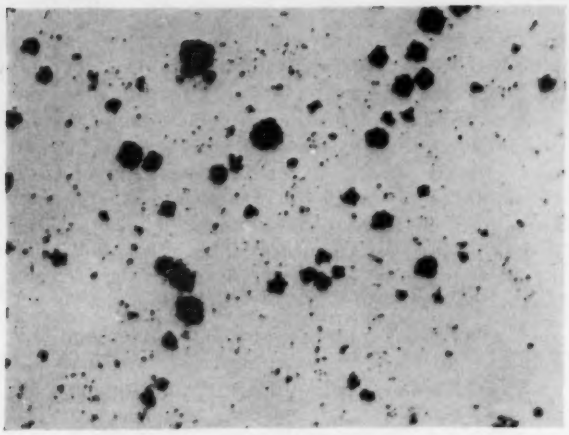


Fig. 3 — Photomicrograph showing size and distribution of graphite nodules after iron H as-cast (see Fig. 1) is graphitized at 1740 F (950 C). Unetched. Mag. — 100X.

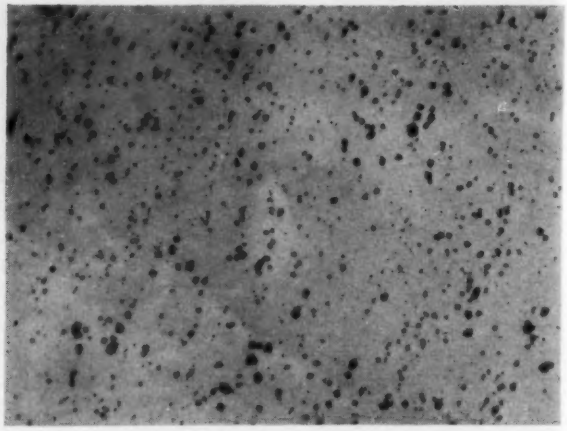


Fig. 4 — Photomicrograph showing size and distribution of graphite nodules after iron H, rolled to 81 per cent reduction in thickness at 1515 F (825 C), is graphitized at 1740 F (950 C). Unetched. Mag. — 100X.

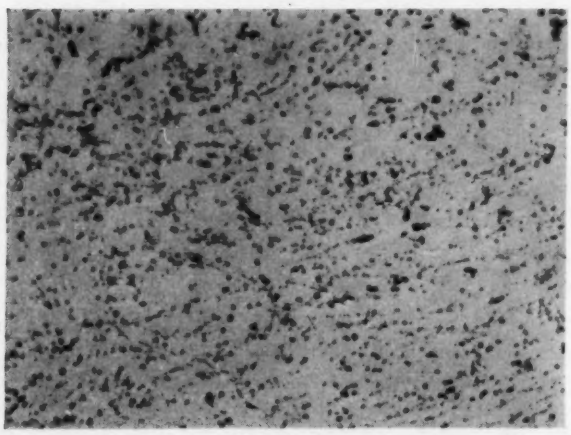


Fig. 5 — Photomicrograph of iron H reacted at 1740 F (950 C) after rolling to 65 per cent reduction in thickness at 1200 F (650 C). Note "crack shape" graphite particles. Unetched. Mag. — 100X.

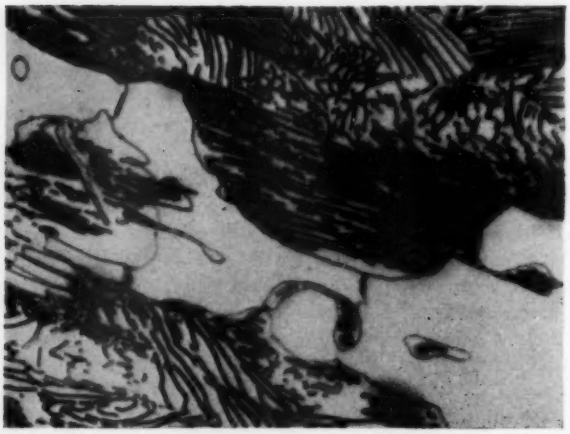


Fig. 6 — Iron H as-rolled to 17 per cent reduction in thickness at 1200 F (650 C). Transverse section showing cracks in the eutectic cementite. Electropolished. Nital etch. Mag. — 2000X.

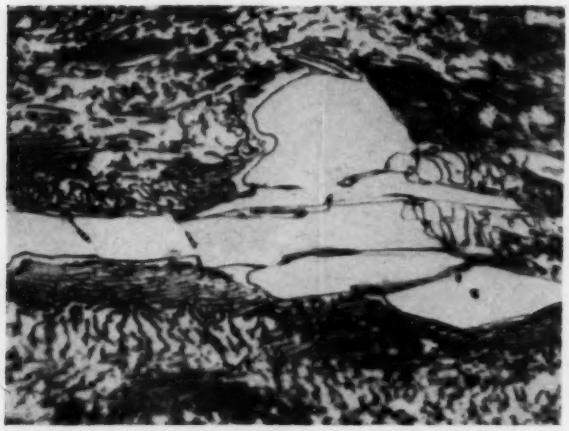


Fig. 7 — Iron H as-rolled to 65 per cent reduction in thickness at 1200 F (650 C). Transverse section showing cracks in the eutectic cementite. Electropolished. Nital etch. Mag. — 2000X.

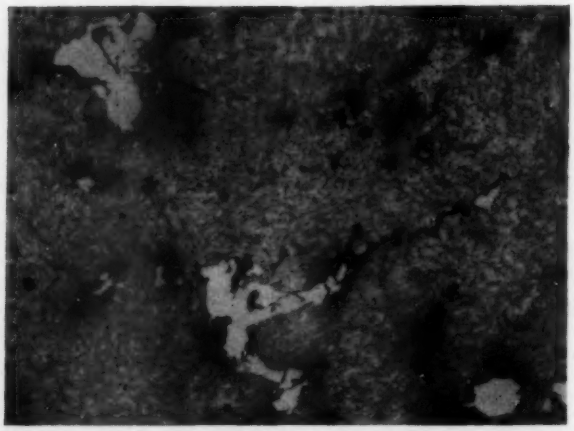


Fig. 8 — Microstructure of iron H partially reacted to graphite during first stage reaction after rolling to 65 per cent reduction in thickness at 1200 F (650 C). Note graphite growing in cracks of eutectic cementite. Nital etch. Mag. — 750X.

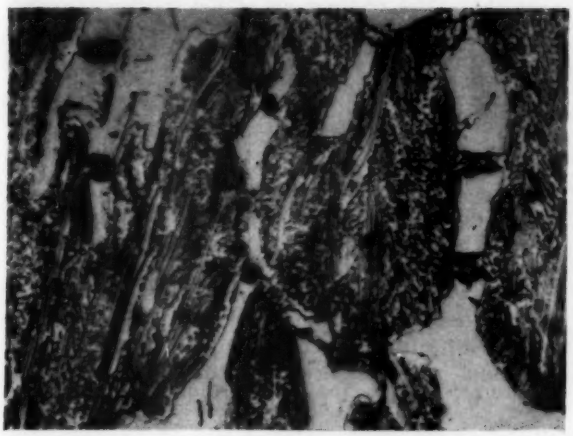


Fig. 9 — Iron H as-rolled to 66 per cent reduction in thickness at 1290 F (700 C). Longitudinal section revealing stages in process of pearlitic matrix "filling in" cracks in massive eutectic cementite. Electropolished. Nital Etch. Mag. — 750X.

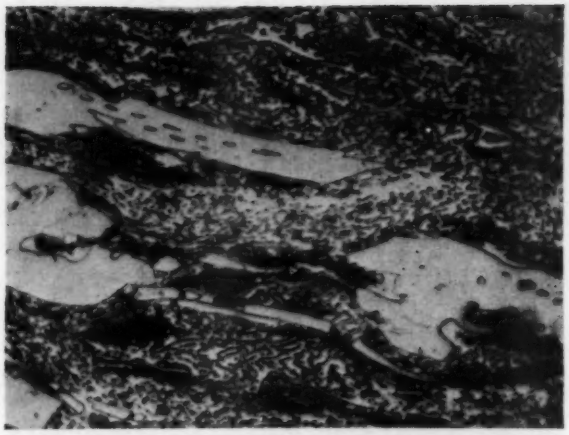


Fig. 10 — Iron H as-rolled to 85 per cent reduction in thickness at 1290 F (700 C). Longitudinal section. Electropolished. Nital etch. Mag. — 750X.

magnifications show a comparison of iron H treated in various ways. Note particularly the cracks formed in the eutectic cementite at lower rolling temperatures and that the pearlite has not spheroidized (Figs. 6 and 7). Figure 5 shows how the specimen (Fig. 7) rolled to 65 per cent reduction in thickness at 1200 F (650 C) reacts to graphite; there is considerable "crack shaped" graphite here while there is none in specimens rolled at higher temperatures (Fig. 4).

That graphite actually nucleates and grows in cracks of the eutectic cementite is more readily seen in a rolled specimen partially reacted to graphite at 1650 F (900 C) and quenched (Fig. 8). At higher rolling temperatures the cracked cementite separates and is filled in by the more plastic matrix; Figure 10 shows the "filling in" process going on in various stages. The large unresolvable voids seen in Figs. 9 and 11 are the result of the enlargement of partially filled cementite crack separations by the electropolishing.

At the highest rolling temperature, 1515 F (825 C), the matrix is completely spheroidized (Fig. 11). It was found that during the rolling treatment at 1515 F (825 C) the ferrite transforms to austenite but some of the pearlitic cementite remains for the short time at temperature. Rolling at 1290 F (700 C) is definitely below the transformation range for iron H, while 1380 F (750 C) is expected to be in the transformation range.

The nodule counts resulting from the systematic rolling experiments show no detectable dependence on the reaction temperature. If one averages the counts of a given rolling temperature at each reaction temperature, it is found that almost all of the values are within the limit of experimental error. Thus, the average value of the four reaction temperatures for each reduction in thickness is plotted in Fig. 12 from which the following features should be noted.

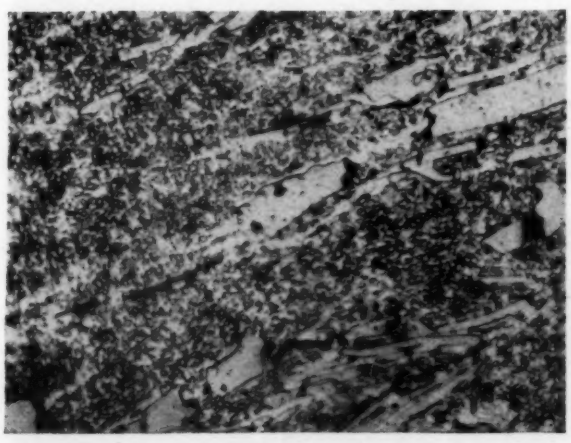


Fig. 11 — Iron H as-rolled to 81 per cent reduction in thickness at 1515 F (825 C). Longitudinal section. Electropolished. Nital etch. Mag. — 750X.

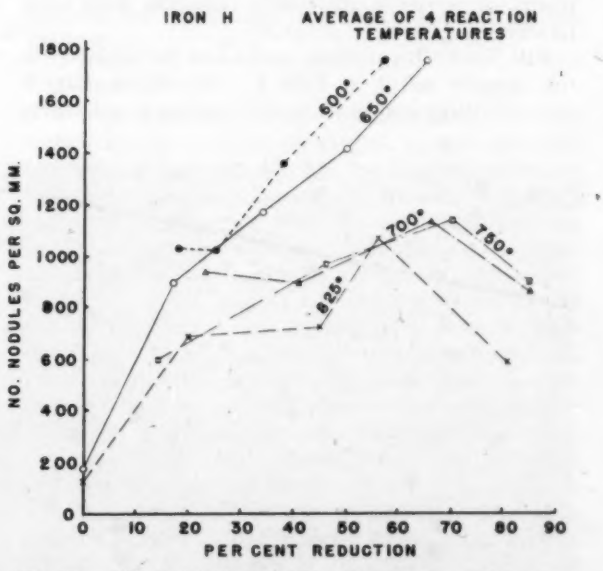


Fig. 12 — Effect of indicated rolling temperature and per cent reduction in thickness on number of nodules counted. Each point on the curve is the average value of individual points obtained at four reaction temperatures, 1560 F (850 C), 1650 F (900 C), 1740 F (950 C), and 1885 F (1030 C).

The number of nodules for the rolled samples is increased by a factor of five to ten over that of the cast material having the same thermal treatment; the exact increase depends upon the rolling temperature and the per cent reduction in thickness. In general for a given per cent reduction in thickness, the number of nodules varies inversely with the rolling temperature.

The curves for the specimens rolled at 1110 F (600 C) and 1200 F (650 C) correspond rather closely but they are separated from and differ in shape from those rolled at higher temperatures. The drop in the number of nodules at 80-85 per cent reduction at rolling temperatures from 1290 F (700 C) to 1515 F (825 C) was confirmed by the results of supplementary work when iron H was rolled at 1650 F (900 C); these are not plotted with the other data. This distinguishing feature was thought to be either a microstructural or a recovery effect.

The curves for the iron rolled at 1290 F (700 C), 1380 F (750 C), and 1515 F (825 C) do not differ significantly in shape (Fig. 12) in spite of the fact that the specimens were rolled below and above the transformation range. The microstructures of the specimens rolled to the same per cent reduction at 1290 F (700 C) and 1515 F (825 C) seem to differ only in the degree of spheroidization of the pearlitic cementite, see Figs. 10 and 11. It is indeed difficult to reconcile the drop in the curves at 85 per cent reduction in terms of the carbide-interface hypothesis when Figs. 10 and 11 are compared.

Back-reflection x-ray diffraction techniques were used to study the straining of iron H due to the rolling. The most sensitive reflections from cementite, those where 2θ is close to 180 degrees, were completely obliterated after even the smallest amount of working examined here, 14 per cent reduction at 1380 F (750 C). Nevertheless, reflections from the (310) planes of ferrite using cobalt radiation were satisfactory.

The ferrite Kα doublet could not be resolved for the samples rolled at 1110 F (600 C); at 1200 F (650 C) rolling temperatures the doublet is just barely resolvable. The photograms of the specimens rolled at 1380 F (750 C) showed a resolvable Kα doublet up to 70 per cent reduction but not with the specimen reduced to 85 per cent in thickness. It had been anticipated that recovery effects might be observed in the latter specimen but no such indications were evidenced.

The rate at which rolled irons were heated to gamma-range graphitizing temperatures was investigated. Figure 13 shows a comparison of the number of nodules formed in cast and rolled iron E as a function of the rate of heating the white iron to reaction temperature. These curves are representative of the white irons tested. It appears as though the minimum in the curve for the rolled samples might occur at more rapid rates of heating than that exhibited by the cast samples.

The possibility of "annealing out" the residual stresses introduced in the white iron by rolling prior to a graphitizing anneal was tried. Some samples from rolled specimens were reheated to temperatures of 1200 F (650 C) and held for various lengths of time before heating to gamma-range temperatures for

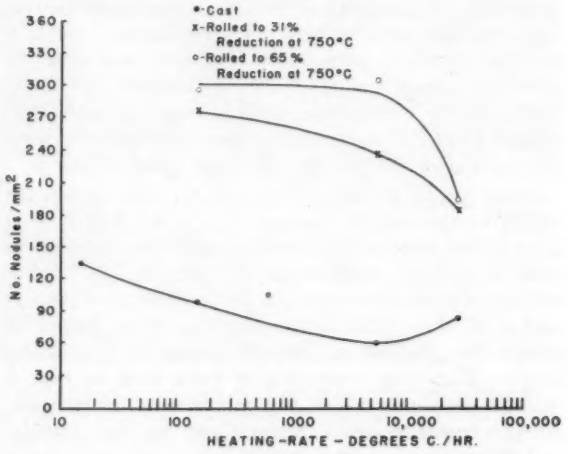
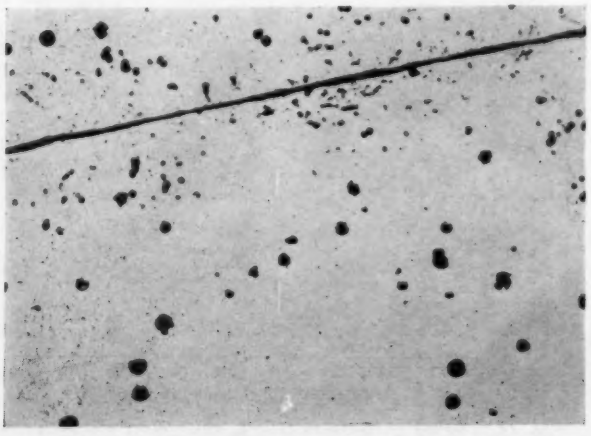
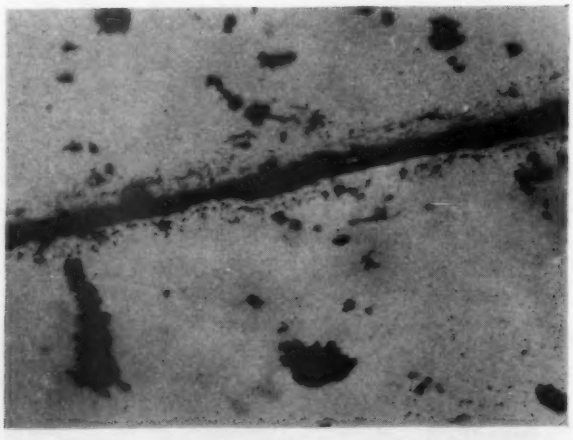


Fig. 13 — Variation in number of nodules with rate of heating iron E to 1650 F (900 C) reaction temperature.



A



B

Fig. 14 — Photomicrograph of two pieces of iron E shot peened and graphitized at 1650 F (900 C). Bakelite mounting separates the two pieces: (A) Note change in size of nodules; those closest to shot peened surface are so small and numerous as to be unresolvable at this magnification; not etched; 100X; (B) same field as (A), not etched; 750X.

graphitizing. A few specimens were heated directly to 1650 F (900 C) after rolling at subcritical temperatures. In no case was there a decrease in the corresponding number of nodules formed compared with a rolled sample reacted to graphite by the usual procedure of rolling, cooling to room temperature, and then reheating to graphitizing temperatures.

Shot Peening

Rolling prior to graphitizing does not give the number of nodules that prequenching does. However, the increasing number of nodules with decreasing temperature of deformation does indicate that deformation at room temperature might produce many more nodules. Residual stresses should also increase with the decreasing temperature of deformation. Shot peening proved to be the method by which white iron could be deformed plastically to a significant degree at room temperature.

Samples of iron E and iron H were shot peened at room temperature. They were then sealed in a vycor tube under a reduced pressure of argon and were graphitized at 1650 F (900 C). For a comparison with the effects of the shot peening, cast samples of iron E and iron H were prequenched by holding 15 minutes at 1650 F (900 C) and then quenching into ice water; these were then graphitized at 1650 F (900 C).

Results on the graphitized shot peened samples and the graphitized prequenched samples of iron E are illustrated in Figs. 14 and 15. Counts of the number of nodules in the prequenched specimen were made at a higher magnification than was previously used so that the smallest particle of graphite that can be identified is about 1.0×10^{-4} mm in diameter. The prequenching was found to increase the number of nodules from about 110 (no pretreatment) to about 18,000 per mm^2; note, however, that shot peening increases the number of nodules even more than prequenching; compare Figs. 14 and 15. This was confirmed by the results on iron H, although this alloy is not as responsive to either prequenching or shot peening as is iron E.

Figure 14 shows the extreme range of nodule sizes and number per unit area found in the shot peened iron E; the nodules farthest from the shot peened surfaces are comparable in size and number with those found in iron not pretreated, while the nodules formed at the surface are not even resolved except at higher magnification (Fig. 14). The surface of a shot peened specimen is completely graphitized before the center parts show any appreciable decomposition of carbide.

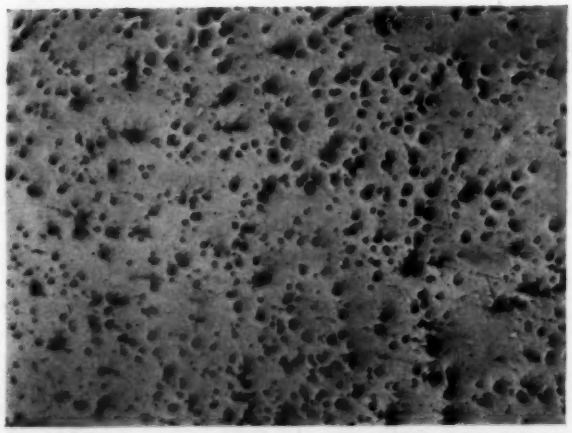


Fig. 15 — Shows effect on size and distribution of graphite nodules when iron E is prequenched to martensite and graphitized at 1650 F (900 C). Unetched. Mag. — 750X.



Fig. 16 — Macrograph of torsion specimen of iron E twisted to fracture and reacted to graphite. Sectioned in plane of neutral axis. Fracture surface on the right. Note increasing number and decreasing size of nodules from neutral axis to surface. Unetched. Mag. — 15X.

Thus, shot peening prior to graphitizing duplicates the effect of prequenching on graphite nucleation *without* the formation of martensite. Cold working is therefore as effective as martensite in nucleating graphite in the alloys tested. The effect of shot peening however can be explained by the carbide-interface hypothesis, by microscopic cracks, or by residual stresses; the three have not been separated by this experiment.

Torsion Studies

It was found that plastically deforming a bar of white iron by twisting in torsion at some elevated temperature prior to graphitizing also increased the number of nodules formed. The number of nodules increased progressively from the neutral (stress) axis to the surface of the twisted specimen, i.e., the rate of nucleation increases with increasing prior strain. In fact along the neutral axis there is no increase at all; the number of nodules formed is identical with the number formed in a specimen not pretreated. Figure 16 shows a section in the plane of the neutral axis of a torsion specimen which was twisted to fracture. The graphite nodules formed near the surface are about four to five times more numerous than those found along the neutral axis.

Results on the specimens which were twisted to rupture at 1290 F (700 C) and which broke after a 30-degree twist are shown in Fig. 17. The number of nodules counted near the surface of the specimen is plotted against the time of holding at 1290 F (700 C) after the rupture. The results show the actual increase in the number of nodules formed as a conse-

quence of the deformation and of the stress-relieving anneal compared with the count along the neutral axis. The minimum in the curve is quite significant and will be discussed shortly. A few specimens twisted at 1380 F (750 C) indicated that a similar curve would be expected at this temperature.

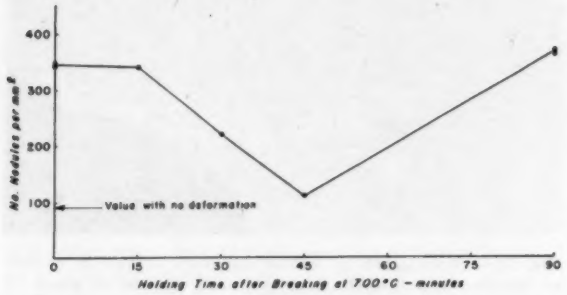


Fig. 17 — Variation of number of nodules formed at 1650 F (900 C) as a function of holding time for iron E.

Discussion

Results of the Torsion Experiments

Results of the various experiments can probably be explained in terms of the carbide-interface hypothesis as well as in terms of residual stresses; the important exception is the torsion work. It would be difficult indeed to explain the shape of the curve of Fig. 17 in terms of a changing carbide-interface or Bunin's microscopic crack hypothesis. The shape of the curve can be explained if residual stresses are considered somehow as active agents in the nucleation of graphite.

Consider the schematic diagram (Fig. 18) in terms of the number of nodules that grow to a visible size at some gamma-range reaction temperature (curve T) as a function of the holding time at 1290 F (700 C). Curve N represents the number of nuclei formed *during the holding time at 1290 F (700 C)* and which will be stable enough to grow to be visible nodules at gamma-range temperatures; the number would represent those formed from all causes including residual stresses.

Curve R is representative of the number of stable nuclei which *form and grow at gamma-range temperatures* as a result of the residual stresses not yet relieved after a given time at 1290 F (700 C). The residual stresses formed by twisting will decrease with increasing time of holding at 1290 F (700 C). Curve T is the sum of curves R and N and it represents the total number of nodules counted after gamma-range graphitization. It should be noted that there is no microscopically visible graphite even after holding 90 minutes at 1290 F (700 C).

The experimental data plotted in Fig. 17 also suggest that graphite nucleation does not take place during deformation but after a suitable time lag and/or heat treatment. That is, the conditions for nucleation are set up by the deformation but the actual nucleation proceeds in a manner to be expected from fluctuation theory.[22]

The results of the torsion work (Fig. 17) and the foregoing explanation of the results in terms of residual stresses indicates that residual stresses can be an active nucleating agent for graphite precipitation. It is therefore proposed that the residual elastic microstresses inherent in a plastically deformed white cast iron act to increase the nucleation of graphite.

Other Experimental Results

Several attempts to "anneal out" the residual stresses in the rolled irons by certain thermal treatments prior to the graphitizing anneal only resulted in a small increase in the number of nodules formed rather than a decrease. As a consequence of the torsion work it is now clear why it was not possible to "anneal out" the effect of the prior deformation. If stable nuclei are continually being formed at any subcritical annealing temperature, then only the short times of annealing immediately after the deformation would be expected to be effective.

An argument for the residual stress hypothesis is the variation of the number of nodules formed as a function of the heating rate to reaction temperature (Fig. 13); these data are indeed incompatible with both the interface and microscopic crack ideas. Note that while both the rolled and cast curves show a smaller number of nodules with increasing heating rate, the cast material does show an increase at the most rapid rate. This increase has been observed previously in white cast irons.[12,13]

One would expect that the competing effects of nucleation and relief of the residual stresses inherent in white iron as-cast would again be operative just as in the case discussed for the torsion experiments. At rates of heating which require several hours or more to reach the temperature for first stage reaction, time enough is provided for the nucleation process to begin before the reaction temperature is reached. As the heating rate increases, there is less time for nucleation to occur but time enough to "anneal out" residual stresses and their effect on nucleation. At some extremely fast heating rate, the residual stresses will not be relieved until the reaction temperature is reached and nucleation of graphite has been effected so that an increased number of nodules is formed.

The results of the rolling work are conveniently

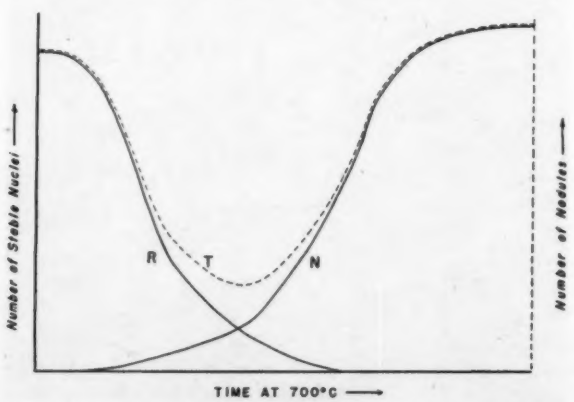


Fig. 18 — Schematic diagram to illustrate anticipated graphite nucleation in twisted white iron bar as a function of annealing time at 1290 F (700 C).

explained in terms of the residual stress hypothesis. Assuming that residual elastic microstresses are important to the nucleation of graphite, then the decrease in the number with the largest reduction at 1290 F (700 C), 1380 F (750 C), and 1515 F (825 C) (Fig. 12) could be due to a recovery effect, even though no recovery was detectable by the x-ray technique used. The difference between the 1110-1200 F (600-650 C) and 1290-1515 F (700-825 C) nodule count curves may be accounted for in part by the difference between "hot" and cold rolling, i.e., a partial recovery seems to be taking place at 1380 F (750 C) but not at 1200 F (650 C) according to the resolution of the ferrite K*a* doublet.

It is not to be implied however that the ferrite will determine the characteristics of graphitization in rolled irons. The condition of the cementite is also expected to be of importance, although the exact mechanism of the graphite reactions in strained samples of white iron is not known. The x-ray back-reflection photograms show considerable broadening of the cementite lines for all the rolled specimens. This indicates that either the cementite is being reduced to a fine grain size or the cementite lattice is strained, or both.

Since it appears that almost all of the diffracted line intensity is contributed by the eutectic carbide, the line broadening is probably due to strain in the cementite; the photomicrographs of electropolished specimens do not show fragmented eutectic cementite but rather it tends to break up into large pieces (Figs. 6 to 11). It has been demonstrated[23] that cementite can be strain hardened to about two hundred points Vickers hardness above the nominal cast value[24] of 820 Vickers hardness. The micro-hardness of the cementite in the rolled irons does not appear to differ appreciably from that of the cast material.

Another interesting consequence develops from the dependence of the number of nodules formed on the temperature of deformation. From a consideration of the general level of residual stresses formed by plastic deformation at various temperatures, the number of nodules subsequently formed is readily explained. Consider the maximum number of nodules that is the result of shot peening prior to graphitization compared with the very small effect when the deformation takes place at 1650 F (900 C). Rolling at various temperatures shows the number of nodules for a given reduction to vary approximately inversely with the rolling temperature.

Since the yield strength of the alloy is expected to go down as the temperature is raised, it is not possible to put as high a level of residual stresses in a metal worked at an elevated temperature as compared to the level possible at room temperature. Also, the higher the temperature the more opportunity there is for the removal of any residual stresses formed, that is, "hot" working is approached. The marked similarity of the curves and their numerical values for 1290-1515 F (700-825 C) rolling temperatures in Fig. 13 would indicate that structural changes due to the phase transformation are minor compared with some other nucleating factor such as residual stresses.

The fact that severe cold deformation such as shot peening will increase the number of nodules even more than prequenching to martensite indicates that it is not necessary to postulate that the tempering reaction is responsible for the prequenching effect. Cold worked structures and martensitic structures may well be nearly equivalent as far as residual stresses are concerned. A recent study[25] of x-ray diffraction line broadening as a consequence of residual strains induced by quenching to martensite and cold work concludes that these two structures differ only in degree.

It is possible that the level of residual stresses inherent in any ordinary white iron as-cast can account in part for the graphitizing behavior of the casting. Residual stresses can be used to explain certain other known variations in the nucleation of graphite not investigated in this work, e.g., the variations observed with different cooling rates from the casting temperature and the prebaking effect.

Conclusions

1. White iron specimens were rolled at temperatures from 1110-1515 F (600-825 C) to reductions in thickness up to 85 per cent; samples were reacted at temperatures from 1560-1885 F (850-1030 C); the number of nodules formed in the rolled specimens is 5-15 times that of specimens not pretreated.

2. The number of nodules formed varies inversely with the temperature at which the white iron was previously deformed. Specimens shot peened at room temperature prior to graphitizing produce even more graphite nodules (close to the shot-peened surface) than does the graphitization of martensite. Thus, cold working is as effective as martensite in nucleating graphite in white irons, i.e., tempered martensite is not a prerequisite for the formation of exceptionally large numbers of graphite nodules.

3. The number of nodules formed after complete reaction varies directly with the amount of prior plastic deformation except for the irons rolled to 85 per cent reduction which exhibit fewer nodules than the irons rolled to about 70 per cent reduction in thickness; this has been considered a recovery effect. White iron specimens twisted in torsion prior to graphitizing show a progressive increase in the number of graphite nodules from the neutral (stress) axis to the specimen surface, i.e., the number is directly related to the amount of prior plastic deformation.

4. When specimens are annealed at 1290 F (700 C) immediately after twisting to rupture, the maximum number of nodules counted per unit area after graphitizing at 1650 F (900 C) goes through a minimum when plotted against the annealing time. This variation in the number of nodules as a function of holding time at 1290 F (700 C) cannot be explained in terms of existing hypotheses of graphite nucleation. The data can be explained in terms of residual stresses.

5. It is therefore proposed that residual elastic microstresses inherent in a plastically deformed white iron act to increase the nucleation of graphite. The proposed hypothesis can explain the results obtained

in this study as well as most of the variations in graphite nucleation reported in the literature on ordinary white iron alloys.

6. In all deformed irons, the number of nodules formed at gamma-range temperatures is found to be essentially independent of the reaction temperature from 1560-1885 F (850-1030 C).

7. The temperature range which separates cold and "hot" working probably occurs between 1110 F (600 C) and 1290 F (700 C) as far as the ferrite is concerned. This may account for the difference in the shape of the nodule count curves for the irons rolled at 1110-1200 F (600-650 C) as compared with the curves for the irons rolled at 1290-1515 F (700-825 C). However, the cementite K*a* doublets could not be resolved for any of the rolled specimens which suggests that the cementite is always strained by rolling.

8. For the several rates of heating to reaction temperature investigated, the number of nodules counted after reaction goes through a minimum for the iron as-cast but not for the rolled irons.

Acknowledgments

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EQUIPMENT FOR CUPOLA-EMISSION CONTROL

By

John M. Kane*

Foundry cupola emissions present a serious problem to the foundry industry. It is difficult to define emission quantities, composition, or particle size distribution except in the broadest of terms. Even with field test data, quantities can vary throughout the day and from day to day. It becomes apparent, however, that solids discharged from cupolas can be classified in three fractions—(1) cinders, (2) dust, (3) metallic fumes and smokes. Classifications are similar to those used in coal-burning boiler stacks so a comparison takes on considerable significance because the boiler stack has been the subject of long study. Air pollution control standards for boiler stacks have been established in many communities.

Cinders: Cinders from the cupola can be defined as extremely coarse particles generally larger than 200 mesh with a lower limit in the 325-mesh screen or 44 micron range. They are aspirated into the cupola gas stream generally in spurts, which occur as a charge is dropped and again later in the cycle as channeling of the blast air releases more material during low bed. As expected, the greater the height of charge fall, the greater the variation in charge height, the higher the quantities of cinders discharged.

Composition and quantity are a function of the charge and charging methods. While cinders are basically coke fines, quantities of scale, burnt sand, and disintegrated briquets will also be found to various degrees.

Consequently, cinder concentrations can be reduced by:

(a) frequent charging to reduce height of fall,
(b) screening of coke to remove fines, covered storage to prevent weathering, minimum handling from receipt to use,
(c) melting practice that uses lowest possible coke to iron ratio,
(d) smallest possible charging opening to reduce indraft volume and resultant stack draft velocities,
(e) close control of briquets if used to assure consistent compaction and bonding,
(f) after burning in cupola stack or hot blast heat exchanger,
(g) moderate and controlled air blast volume to tuyeres,
(h) proper scrap selection and sizing to prevent voids in bed, and
(i) even distribution of charge by use of most effective charging equipment.

Dust: A considerable fraction of any cupola stack emission will be finer solid particles in the 1 to 44 micron range. This fraction can be compared to the fly ash in the boiler stack analogy. Its composition, however, is not homogeneous, and varies from cupola to cupola. Abrading of coke produces coke dust, silica fines result from charging dirty sprues and risers, heat releases adhering fines on the crushed limestone, briquets disintegrate from attrition, scale and settled foreign material adheres to scrap during its transportation to and storage at the foundry.

Smokes and Fumes: Like the boiler stack, much of the visible appearance in the cupola stack discharge can be attributed to partial combustion of carbonaceous materials (oily scrap is a major source). With the reducing atmosphere in the melting zone, complete combustion is more difficult than in a boiler with its excess air to assure maximum burning efficiency.

Incineration in the cupola stack with CO igniters or in the hot-blast heat exchanger will appreciably reduce the brown smoke fraction of the visible emissions. Such incineration also reduces many coke cinders to ash, increasing the dust fraction to a certain extent.

Since this is a metallurgical process, metallic fumes will be formed and contribute to the visible fraction. Metallic fume concentrations vary with composition of scrap and will increase with briquet percentage of the charge, with increased molten metal temperatures, and during melt down.

Particle sizes of smokes and metallic fumes are sub micron in size with usual range in the 0.2 to 0.5 micron range.

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It appears unlikely that public education will successfully explain to neighbors that visible stacks are not necessarily the source of nuisance depositions. Until such recognition is achieved, the public will continue to trace any nuisance dust problem to the visible stacks in their immediate area.

Actually, the fume concentrations range from some 1 to 3 lb/hr/ton of melt for most cupolas.

Control Equipment

Proper selection of control equipment must evaluate the results needed in each of the above ranges, if performance of control equipment is to meet expected performance. Each fraction contributes to a different phase of the emission problem, as:

(a) Cinders are so large in particle size that settlement occurs within a very short distance from the cupola stack. With favorable cupola location, they do not contribute to either public nuisance problems in the neighborhood nor are they visible in the cupola stack gases. Where cupolas are located close to plant boundaries, nuisance quantities can be blown from roof or deflected from stack with high winds. Their elimination is helpful and often economically justified by reduced maintenance to foundry roofs, and immediate plant area gutters. With a 10 ton per hour melting rate, it will be usual for cinder quantities of 500 pounds per 8-hour day to settle in the stack area. Removal costs to one foundry with three cupolas in operation melting 60 tons per hour averages $25,000 per year.

(b) Dust settles at greater distance from the cupola stack with deposition occurring throughout the foundry plant area and in most cases in adjacent areas to the foundry property. This fraction usually accounting for 2 to 6 pounds per ton melt can be a nuisance to plant or neighborhood. Oxidizing metallic particles and some fluxes can attack auto finishes and other metallic surfaces with resultant property damage.

(c) Smokes and fumes are of such small particle size that prolonged suspension and natural dispersion removes them without difficulty except in areas of stable atmospheres. Because they contribute so greatly to the visibility of the stack effluent, control is desirable when cost can be reconciled and justified.

Control equipment can be quite readily fitted into the categories described.

A. Wet Cap

Original concept in such designs was one of a cinder catcher to reduce foundry roof maintenance. For such purpose, wet cap collectors on the cupola stack have been consistently successful and represent the major present day step toward cupola control by the foundry industry. Performance comparable to the boiler-low-pressure-drop cinder catcher could be expected. Collection appears excellent on plus 200-mesh particles, with good removal down to 325-mesh sizes. Removal of particles less than 44 microns appear doubtful.

Advantages

Typical design is shown in Fig. 1. Characteristics of this group of collectors include:

(1) ***Simplicity:*** It operates on stack draft and does not require induced draft fans, protective cooling, added horsepower for operation, cupola top lids with their actuating mechanism and control equipment to open lids in case of power or equipment failure.

(2) ***Collection:*** Collection is effective in the cinder range, removing that fraction which would otherwise require manual removal from roof and gutters. Collection of material as a slurry often fits refuse disposal into existing handling systems in the foundry. (Drain water can often be incorporated with slag quenching and sluicing system.)

(3) ***Cost:*** Equipment and installation costs are lowest of methods currently used for cupola emission control. This is true even though a collector is required for each cupola and any one unit is generally in service only on alternate days.

Disadvantages

(1) ***Corrosion:*** Water requirements per cupola are in the 100 to 300 gpm range, which makes recirculation of waste water including pumps and dewatering tanks usual. Sufficient acid is picked up from the cupola gases to require chemical treatment of the recirculated water. Corrosion will also be a factor reducing life of outer shell in zone of damp surfaces on cleaned gas areas. Entrained mist can be troublesome if collector is not properly designed for gas volume.

(2) ***Protection:*** Heat damage can occur with stoppage of water supply to the collector.

(3) ***Collection:*** Particulate removal is limited to coarse particles, generally above 44 microns. Discharge appearance is not improved and substantial quantities in the 1 to 44 micron sizes escape to atmosphere.

B. Dry Type Centrifugals

Efforts to provide more effective dust removal from the cupola stack gases have resulted in application of dry type centrifugals to cupola gas cleaning—equipment that has been effective for coal burning boiler fly

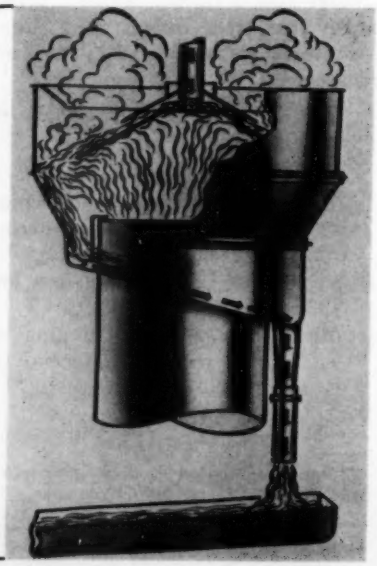
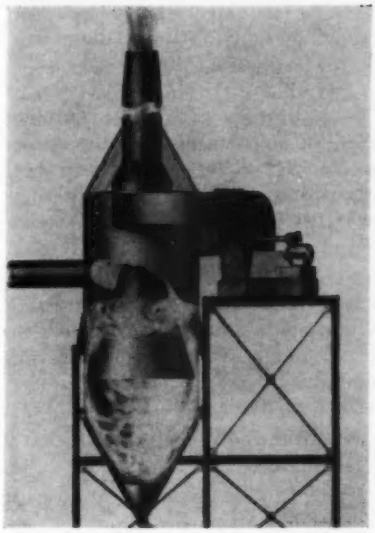
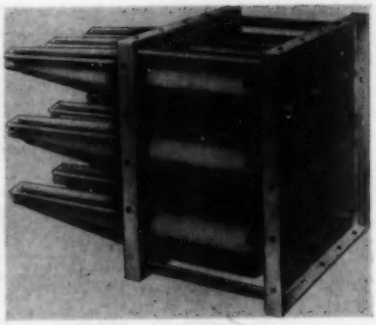


Fig. 1 — Cupola wet cap collector.



(Left)
Fig. 2 — Low pressure drop centrifugal collector.

(Right)
Fig. 4 — High efficiency centrifugal collector.



ash control. Similar removal effectiveness has been demonstrated on cupola gases where effective removal of particles down to 20 microns could be expected with low pressure drop centrifugals, down to the 5 to 10 micron range with high efficiency, high pressure drop designs.

Figure 2 illustrates low pressure drop design; Fig. 3 shows a typical installation, and Fig. 4 is typical of the higher pressure drop designs, with Fig. 5 indicating system incorporating such equipment. Characteristics of this group include:

Advantages

(1) ***Collection:*** Collection is effective in the "public nuisance" dust range. Equipment should eliminate damage or nuisance caused by solids deposition in plant and neighborhood area.

(2) ***Dry Collection:*** Dry gases eliminate corrosion problems and discharge a hot and dry gas stream, more conductive to dispersion of remaining smokes, fumes, and finer dust particles. Collected dust in dry form may aid in its disposal or can complicate its handling depending on disposal methods and refuse handling equipment used by the foundry.

(3) ***Induced Draft:*** Use of induced draft eliminates back pressure in the cupola and maintains calculated indraft control at the charge door opening.

(4) ***Protection:*** Protection devices bypass gas in case of failure.

(5) ***Cost:*** While more expensive than the wet cap, systems in this group are comparatively low in equipment cost and in replacement part and servicing required. Life expectancy of system elements should be long.

(6) ***Future Improvement:*** Elements of dry centrifugal systems are identical to those required should complete control of all solid emission be attempted. Design permits addition of aftercleaners should future regulations require without scrapping or altering equipment installed.

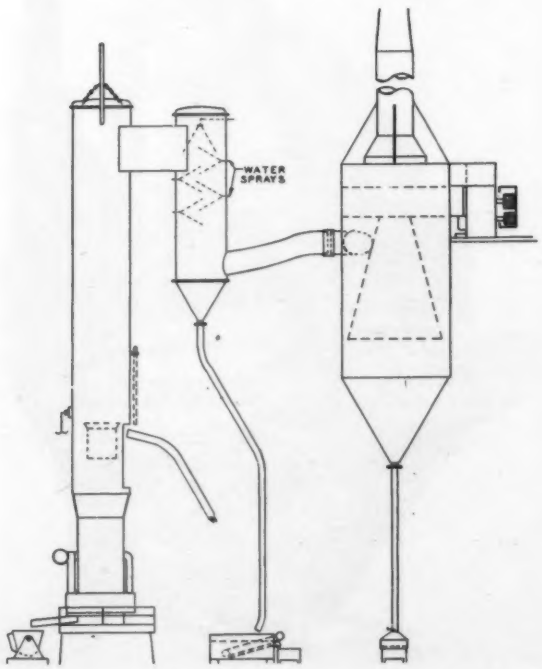


Fig. 3 — Low pressure drop centrifugal installation.

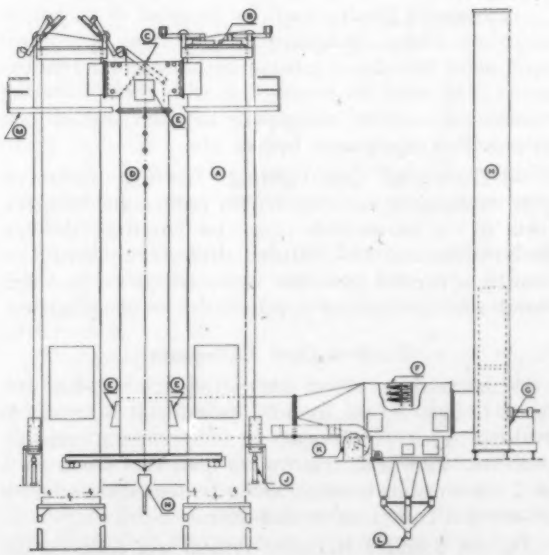


Fig. 5 — High efficiency dry centrifugal installation.

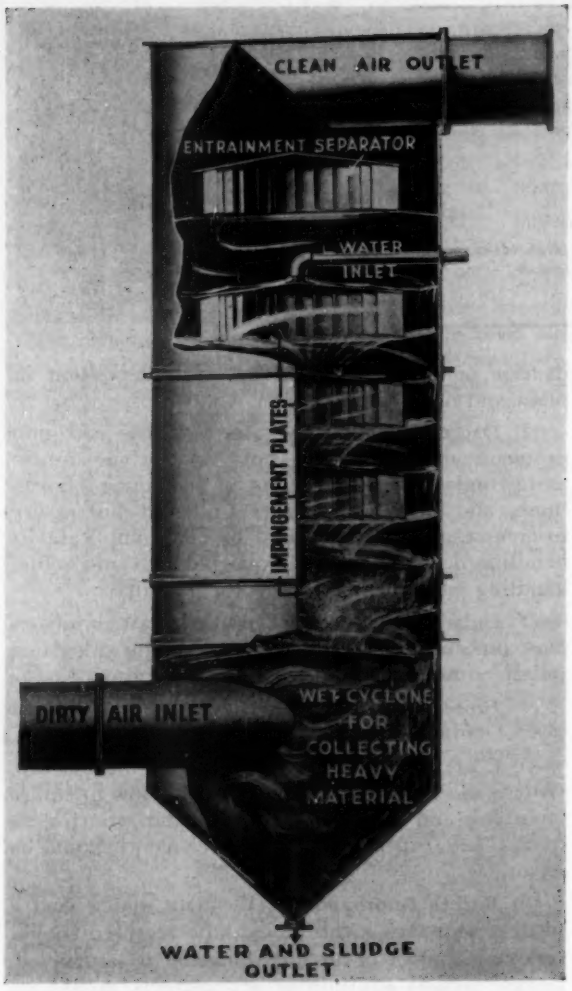


Fig. 6 — High efficiency wet centrifugal collector.

Disadvantages

(1) ***Induced Draft:*** Need for induced draft fan to overcome added resistance to gas flow of control equipment introduces substantial horsepower requirements. The need for cupola lids with their actuating mechanism, control equipment to open lids in case of power or equipment failure.

(2) ***Controlled Gas Cooling:*** Cooling towers or heat exchangers are required to reduce gas temperatures to the permissible range for handling through ducts, collectors, and induced draft fans. Controlled cooling is needed to reduce or eliminate excess water runoff and operational problems due to condensation.

C. Wet Dust Collectors

With extensive use of high efficiency wet dust collector designs by the foundry industry, it is logical to evaluate such equipment for the cupola emission problem. Excellent removal of particles down to 1 or 2 microns has been consistently demonstrated with substantial collection of sub-micron particles.

Figures 6 and 7 illustrate typical wet collector designs that have been applied to cupola gas cleaning. Performance can best be evaluated by a comparison with the dry centrifugal systems outlined above.

Advantages

(1) ***Collection:*** Collection is pratically complete on all dust particles, substantial on metallic fume, and negligible on smokes. (Discharge appearance, however, will not be substantially reduced although color will change with combination with steam plume.)

(2) ***High Temperature Handling:*** Use of water provides "flash cooling" at collector inlet permitting handling of gases throughout the cupola gas temperature range. (Precooling may be required to protect duct connections from excessive cupola gas temperatures unless refractory lined flues are used.)

(3) ***Induced Draft:*** Same indraft control at charging door opening is obtained.

(4) ***Protection:*** Gases can be bypassed in case of failure without damage to equipment.

(5) ***Wet Collection:*** Collection of material as sludge or slurry often fits refuse disposal into existing handling systems in the foundry. (Drain water can often be incorporated with slag quenching and sluicing system.)

Disadvantages

(1) ***Corrosion:*** The greater impingement, aeration, and centrifugal forces utilized in effective wet collector designs proportionately increase corrosion rates. It is debatable whether chemical treatment of water will be sufficient protection for damp surfaces

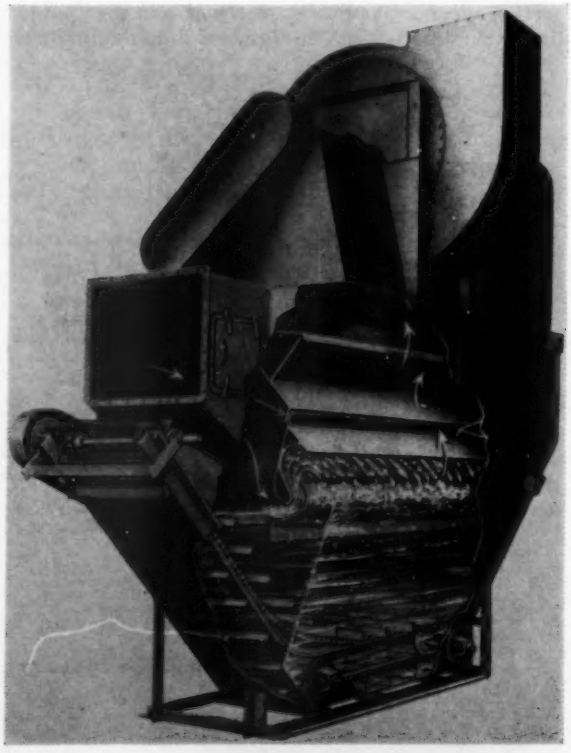


Fig. 7 — High efficiency hydrostatic collector.

in the cleaned gas, exhaust fan, and discharge stack surfaces.

(2) *Dispersion:* Exit gas conditions cooled from the extremely high temperatures of cupola gases will contain substantial quantities of water vapor. Exit temperatures will be in the 125 F-190 F range and will be close to saturation. Dispersion during stable atmospheres can be retarded. Possibility exists due to large volumes involved of condensation taking place under such conditions with precipitation of droplets in the nearby plant or neighborhood area.

(3) *Future Improvements:* Cooling in wet collectors is too effective to make addition of aftercleaners practical should complete control of all solid emissions be required at a future date. (With water vapor quantities introduced in cleaned gas discharge, reheating by bypassing a fraction of the hot gas around the collector requires bypassing of a major portion of the total hot gases handled.)

(4) *Cost:* System cost will be substantially increased over dry centrifugal if corrosion resistant materials of construction for collector, exhaust fan, and discharge stack are required. (Placing of exhauster on dry hot gas side requires controlled precooling or heat exchangers to keep gas within permissible temperature range.)

D. Electric Precipitators

The need in certain localities due to meterological or air pollution regulations for practically complete elimination of all visible cupola emissions has necessitated installation of either electrostatic precipitators or bag houses. Both collector types do provide excellent removal of the finest of solid smoke and metallic oxides. Systems involve the same general arrangements of induced draft fan, cupola lids, and controlled cooling outlined for the dry centrifugal system.

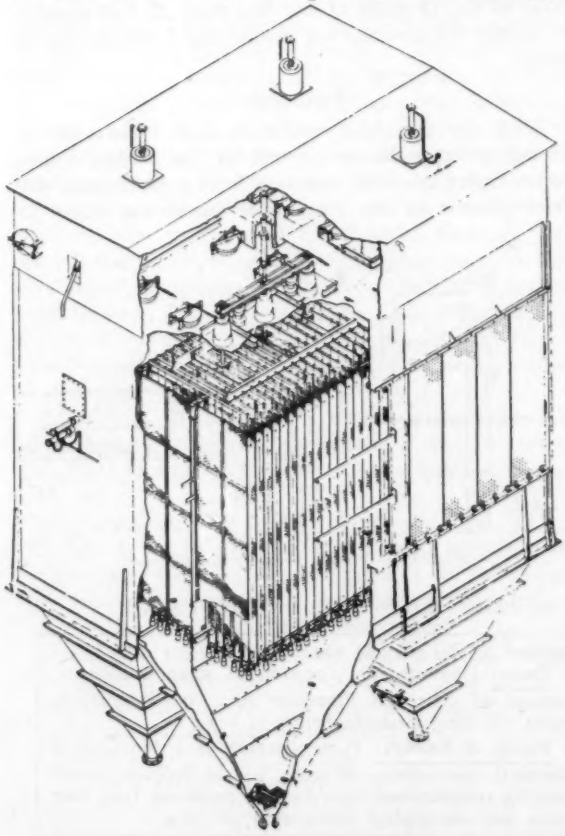


Fig. 8 — Electrostatic Precipitator.

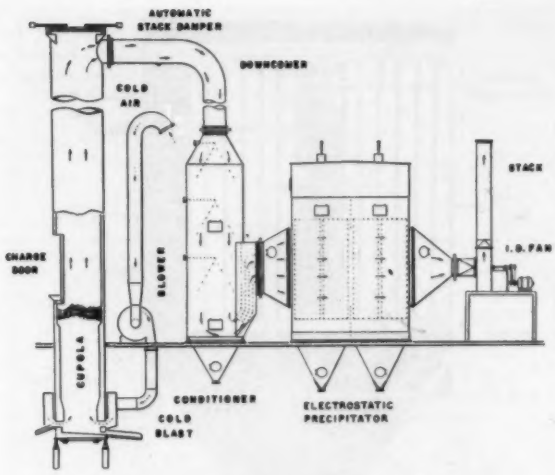


Fig. 9 — Electrostatic precipitator installation.

It is interesting to note that in boiler fly ash installations, it is rare that precipitators are employed except where gas volumes and fly ash loadings are many times those encountered in the foundry cupola. No fabric arresters have been installed for such boiler stack gas cleaning. In the electric precipitator, Fig. 8 and 9, collection is obtained by use of high voltages (50,000 to 70,000 volts) to charge the particles and attract them to grounded collector plates.

Advantages

(1) *Collection:* Excellent removal of solid particles in all size ranges when collector is designed of sufficient size to provide such order of performance.

(2) *Horsepower:* Pressure drop is low, making horsepower required substantially less than for any other induced draft collection system.

Disadvantages

(1) *Cost:* Initial equipment and erection cost is high—the cost could exceed dust control equipment costs for all other dust producing operations including shakeout, sand conditioning, and cleaning room operations.

Typical of many high performance refined designs, maintenance and servicing costs will be substantially greater than for equipment having lower order of collection.

(2) *Conditioning:* For certain fractions of cupola solids including the metallic oxides, gas temperatures and water vapor content must be closely controlled due to high resistivity of such fractions. Too efficient heat exchangers may reduce gas temperatures beyond point where needed water vapor can be cheaply introduced; fluctuations of cupola stack temperatures can complicate maintenance of optimum temperature range to collector. Desirable gas conditions for cupola

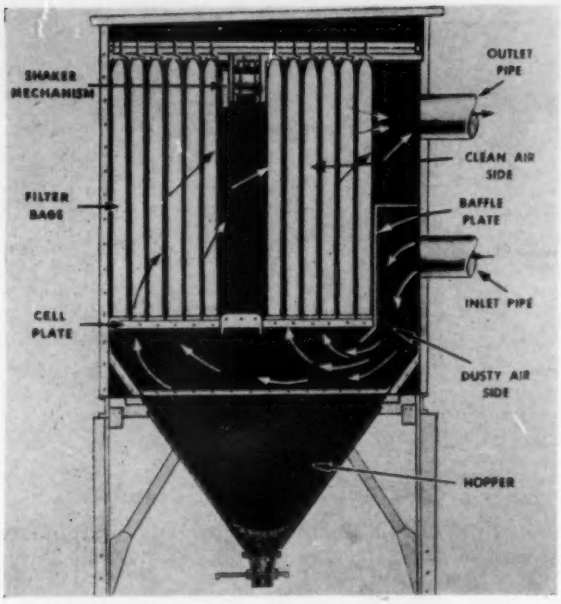


Fig. 10 — Fabric dust arrester.

type of contaminants has been reported as 200 F and 14 per cent minimum water vapor by volume.

(3) ***Disposal:*** Dry material disposal becomes most troublesome as particle sizes of collected material decrease. The ability to remove the submicron fraction introduced problems of refuse handling and secondary dust dispersion during disposal.

E. Fabric Collectors

Good fabric collectors, Fig. 10, can provide the highest order of collection efficiency on all particle sizes. Like the electric precipitator, its use has been confined to date to those areas where complete collection must be provided regardless of equipment cost and maintenance expense. A typical installation. Fig. 11, illustrates an installation designed for production foundries where length of melting period requires automatic, either continuous or periodic, removal of collected material from fabric surfaces without bypassing dirty gases during collector reconditioning cycle. Features include:

Advantages

(1) ***Collection:*** Excellent removal of all solid fractions with efficiency of removal unaffected by temperature or operational phases of cupola. Collection normally of higher order than comparable cost electric precipitator.

Disadvantages

(1) ***Cost:*** Initial equipment and erection cost is high although generally somewhat lower than electric precipitators designed for comparable performance. Replacement part costs include periodic replacement of synthetic filter fabric.

(2) ***Operation Temperature:*** Usual synthetic fabrics in use for cupola and other high temperature applications has an upper limit of 275 F. Keeping cupola gas temperatures below this figure takes accurate gas cooling equipment.

Glass fabrics can operate at temperatures up to 500 F although they appear unsuited to most fabric arrester designs due to abrading action during reconditioning.

(3) ***Disposal:*** Dry material disposal introduced same factors outlined under electric precipitator discussion.

F. Incineration

Consideration of cupola gas igniters of gas burning in heat exchangers cannot be divorced from evaluation of cupola control. With any control equipment except fabric arresters, after burning of the carbonaceous fraction is recommended to reduce discharge appearance. The advantage from a public relations standpoint will usually outweigh the formation of some additional dust particle through the reduction of coke cinders to ash.

To be effective, ignition time must be provided in the cupola stack. A distance from top of charging door to top of stack of not less than 25 feet appears indicated for combustion under usual cupola operations.

Summary

With the increasing publicity and legislation directed at air pollution throughout the United States, some degree of solids removal from cupola gases will be indicated in the foreseeable future for many lo-

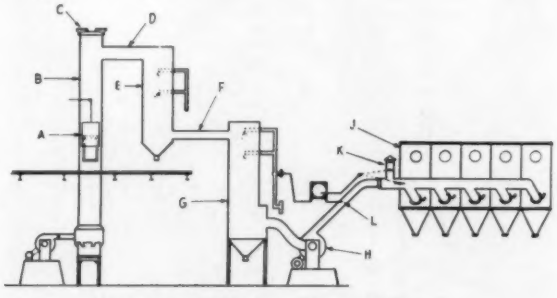


Fig. 11 — Fabric arrester installation.

TABLE 1 — RANGE OF CUPOLA EMISSIONS PER TON OF IRON MELTED IN COLD BLAST CUPOLA

Total Solids	10 to 20#/hr/ton melt
Cinders (plus 44 microns)	50 to 75% of total
Dust (44 to 1 microns)	2 to 6#/hr/ton melt
Fumes & Smokes (less than 1 micron)	1 to 3#/hr/ton melt

Cinder: Quantity is a function of blast volume per ton melted, charging methods, height of bed, quantity of coke fines. Incineration will reduce coke quantities emitted in this particle size range.

Dusts: Quantity is a function of scrap cleanliness, amount of coke and limestone fines, quality of briquets, charging methods, height of bed.

Fumes & Smokes: Fume quantity is a function of chemical composition of scrap, use of briquets, metal melting temperatures. Smokes are produced from oily scrap and incomplete combustion of coke.

TABLE 2 — COLLECTOR CHARACTERISTICS FOR 15 TONS/HR COLD BLAST CUPOLA OPERATED ON ALTERNATE DAYS

Assume Blast Volume = 8,000 cfm @ 70°	Assume Emissions 15#/hr/ton/melt
Charging Door 40 sq. ft. @ 300 fpm = 12,000 cfm @ 70°	Consisting of Cinders 8#/hr/ton/melt
Total Gas Volume = 20,000 cfm @ 70°	Dust 4#/hr/ton/melt
62,500 cfm @ 1200°	Fumes 3#/hr/ton/melt

Collector Type	Discharge Gas Temperature	H_2O Vapor @ Exit Temp. Added for Cooling	Total Vol. @ Exit Temp.	Lower Effective Collection Range (Micron)	Collector Pressure Drop W. G.	Probable H. P. (Note 1) Exit. Temp.	Probable H. P. (Note 1) 70°	Water Req'd. GPM	Probable Emission (Note 2) GR SCF	Probable Emission GR CFM @ 500°	Approx. Instal. Cost on Existing Cupolas
Wet Cap.....	400-700°F	9,850 cfm	47,850 cfm @ 550°F	44	0.25"	—	—	200	0.45	0.25	$35,000
Low Pressure Drop Centrifugal With Cooling Tower..	350-500°F	10,200 cfm	43,600 cfm @ 425°F	30	1.5"	24	40	34.3	0.36	.20	50,000
With Heat Exchanger.....	600-700°F	None	41,800 cfm @ 650°F	30	1.5"	34	71	0	0.45	0.25	150,000
High Eff. Dry Centrifugal.	350-500°F	10,200 cfm	43,600 cfm @ 425°F	7	3.0"	34.3	57	34.3	0.27	0.15	55,000
Wet Collectors....	125-190°F	9,750 cfm	33,750 cfm @ 175°F	1	5.0"	52.5	63	48-100	0.11	0.062	75,000
Electric Precipitator	250-300°F	10,200 cfm	38,000 cfm @ 275°F	0.1	0.5"	18	25	41.2	0.039	0.021	100,000
Fabric Arresters (Synthetic Fabric).	250-275°F	9,600 cfm	36,800 cfm @ 260°F	0.1	4.0"	42.5	57.5	41.8	0.0096	0.0053	75,000
(Glass Fabric)....	400-500°F	10,200 cfm	44,200 cfm @ 450°F	0.1	4.0"	40.5	69.5	33.2	0.0206	0.0114	75,000

Note 1: Horsepower calculations assume pressure loss through dusts and cooling tower of 2" added to stated collector loss except for Wet Cap. (5" with heat exchanger).

Note 2: Lower than the usual boiler stack acceptable emission regulations of 0.257 grains per cubic foot at 500°F (0.465 grains per standard cubic foot).

calities. Increased awareness to the public relations aspect of this problem is indicated by the widespread interest by the industry and the increasing numbers of control equipment installations currently in service.

The nature of the problem and some of the general ranges of control equipment characteristics have been reported in Tables 1 and 2. Installed cost figures reported are relative only and should be used with extreme caution. Where induced fans are employed, the influence of size of charging door on size and cost of equipment is substantial. Indraft velocities vary between 200 and 300 fpm with different designers. Comparison of equipment cost should recognize this possibility and costs should be accumulated on the same indraft basis.

Analysis of the substantial data existing from cupola emission studies (both published and in industrial files) would tend to describe the more usual cupolas as:

(1) Emission with cold blast usually from 10 to 20 lb per hr per ton melted; with incineration type of heat exchanger, emissions are somewhat reduced because part of the coarse fraction settles out in the heat exchanger.

(2) Cinder fraction (plus 44 micron or 325 mesh screen) often from 50 to 75 per cent.

(3) Dust fraction (1 to 44 microns) usually from 2 to 6 pounds per ton melted.

(4) Metallic fume and smoke fraction while some 99 per cent by number of particles account for 1 to 3 pounds per ton melted.

(5) Collection efficiency of fabric arresters 98 to 99 per cent; electric precipitators 92-99 per cent depending on size selection; high efficiency wet collectors 85 to 90 per cent; high efficiency dry centrifugals 70-85 per cent; low pressure drop dry centrifugals 60 to 70 per cent; wet caps 45 to 60 per cent.

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HOT-BLAST CUPOLA PRACTICE

By

Wilhelm L. Heinrichs*

ABSTRACT

The author describes development of basic slags cupola practice from cold-blast to continuous hot-blast operation. The inter-relations of the melting process, the thermal and metallurgical reactions are explained in light of work done by Jungbluth and Korschan.

After a thorough study of acid- and basic-lined hot-blast cupolas existing at home and abroad, Krupps installed last year a hot-blast cupola which has been running continuously under a basic slag. Advantages and disadvantages of the unit, production of gray iron for various uses, and the fundamentals of melting conditions are discussed. Dimensions and data on tuyeres, lining, recuperator, bast temperature, fire-box and top-gas temperature, and water consumption are given.

The inter-relation between combustion rate and metallurgical reaction rate, Si losses, desulphurization of the slag, dephosphorization, carbon pick-up, and intermittent operation in a water-cooled unlined cupola are treated. According to the author's experience, a number of good grades of iron can be produced in the basic hot-blast cupola suitable for various uses in the foundry, and also the liquid metal required for conversion into nodular iron. A well-run hot-blast cupola with water-cooled tuyeres and jacket and an average diameter of 3.3 ft (40 in.) at tuyere level can be operated continuously under a basic slag and can produce up to 5000 tons of liquid metal per month.

A hot-blast carbon-lined cupola which can be run on acid, basic, or neutral slags, has been in operation for a year at the Krupp foundry in Essen, Germany. This paper discusses metallurgical reactions in this cupola operated with basic slags.

Piwowarsky[1] in his book *High-Grade Cast Iron* stated that the cupola is a shaft-type melting unit that gives the most ideal interchange of heat between solid or gaseous fuel and the metallic charge; today it still represents the most important furnace for the production of cast irons.

Much has been written about the thermal and metallurgical processes taking place in the standard cupola. Among many others, Jungbluth and Korschan[2] investigated the fundamental inter-relations in cupola operation. They were the first to summarize systematically in the cupola-behavior diagram the results they obtained.

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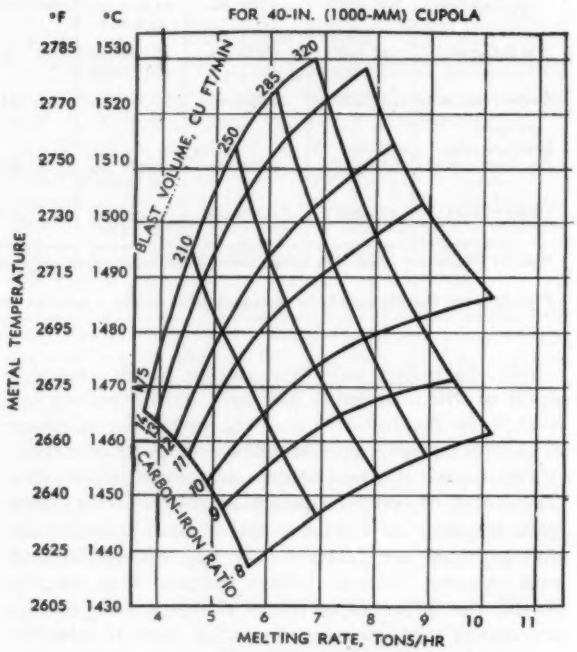


Fig. 1 — Cupola behavior diagram by Jungbluth and Korschan for 40-in. cupola.

Figure 1 shows a diagram for a 3.3-ft (40-in.) cupola converted from a 1.9-ft (23-in.) cupola according to work done by Jungbluth. This diagram shows the relationship between the four variables, melting rate, metal temperature, carbon requirements, and blast volume; with two of these factors fixed the other two can be determined.

The following rules can be deduced from this cupola-behavior diagram:

1. With a fixed coke charge and increasing blast volume, melting rate and metal temperature will increase until the iron has reached a maximum temperature. The heat losses of the melt due to radiation decrease as the throughput rate increases. Exceeding the optimum throughput rate will result in considerable loss of iron and associated elements.

2. With constant blast volume, an increase in coke charge will result in decreased melting rate with increase in metal temperature. This increase in temperature is due to the fact that less iron flows through the cupola in a given time and that the hot layer of coke loses less heat.

3. Large amounts of coke and large blast volumes are necessary to obtain highest metal temperatures.

4. To maintain a constant melting rate and yet increase the temperature of the metal, both coke charge and blast volume must be increased at a ratio which can be read from the behavior diagram (Fig. 1).

However, with a fixed coke charge and a changing blast volume, the combustion ratio remains the same. With an increasing coke charge, combustion ratio decreases and the combustion to carbon dioxide is more incomplete. The lines indicating a constant percentage of carbon (Fig. 1) are, as the figures on the left-hand side in the diagram show, also indicate a constant combustion ratio.

When studying the operation of a cupola, the investigator must run the tests under similar conditions. Composition of the charge, size and carbon content of the coke, and the fluxes, etc., must remain the same. Since in practice these requirements are never fully met, the values obtained from the cupola behavior diagram are never obtained with mathematical accuracy.

In extension of the behavior diagram (Fig. 1), formulas can be set up to explain the mechanism of the melting operation. In this connection, the various formulas available to determine melting rate, blast volume, amount of air per kg of carbon, coke quantity, etc., which Jungbluth[2] discussed in detail in several papers, may be mentioned. The formula for calculating the melting rate, for instance, is as follows:

$$S = \frac{60}{10} \times \frac{W}{L \times \frac{K \times k}{100}}$$

where

S = melting rate in metric tons per hour
W = blast volume in cubic meters per minute NTP
L = amount of air (blast volume in cubic meters NTP per kg of carbon)
K = coke quantity per 100 kg of iron
k = % carbon in the coke.

The numerator of the fraction indicates the blast volume per hour and the denominator the blast volume per ton of iron.

Knowledge obtained from the cupola behavior diagram (Fig. 1) and from the above formula refers to the standard cold-blast cupola, but permits further calculations to be made for hot-blast or oxygen-enriched blast operation.

The first large-scale attempts to obtain greater economy in cupola operation by using higher blast temperatures did not take place until after the first world war. The following advantages were expected:

1. An increase in blast volume that could be delivered per unit of cross-sectional area of the cupola,
2. An increase in the carbon combustion rate,
3. Increased melting rate,
4. Better thermal efficiency,
5. A narrowing down of the heat-absorbing (oxidizing) zone of the cupola which would have the effect of increasing the metal temperature,
6. A decrease in losses by narrowing down of the oxidizing (oxygen-bearing) zone.
7. Better desulphurization due to improved thermal efficiency and a narrowing down of the cooler furnace zone, and
8. Less slagging up of tuyeres and less tendency to form "skulls."

Utilization of the considerable amount of heat available from the cupola top gases for preheating the air blast suggests itself. Figure 2 by Jungbluth and Korschan[2] indicates the heat available from the top gases and shows, in calories per 100 kg (2200 lb) of iron, the amount of heat escaping with the top gases as latent and sensible heat. With small percentages of coke, the heat available from the top gas is half the amount of the heat contained in the metal, but this can be tripled with larger quantities of coke.

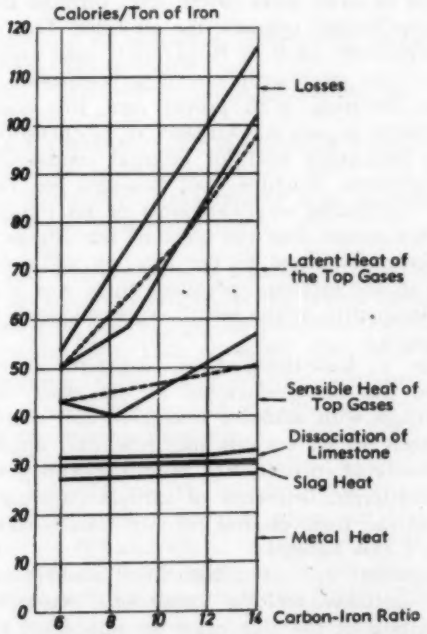


Fig. 2 — Heat balance at a blast volume of 70 cfm with respect to 100 kg (220 lb) of iron for a 23-in. cupola.

The idea of returning some of the heat escaping with the top gases to the cupola by preheating the combustion air in a recuperator or regenerator led to the development of the hot-blast cupola.

By using hot blast, the temperature in the combustion zone is bound to increase so that hotter metal is obtained. Because of the higher temperature in the combustion zone, the reduction to carbon monoxide $CO_2 + C \rightleftharpoons 2\,CO$ will go further than in

cold-blast operation. According to Boudouard's Law, the conditions of equilibrium will change toward CO with an increase in temperature (Fig. 3). This means that according to Jungbluth's formula an increase in melting rate takes place. When keeping the metal temperature constant comparably to cold-blast operation, the amount of coke used with hot blast can be reduced.

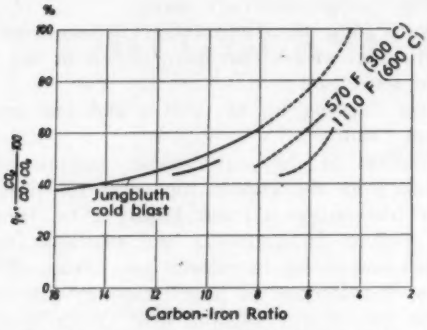


Fig. 3 — Relationship between Kk/100 and N_v with hot-blast operation (Bader).

One type of hot-blast cupola is the Schuermann furnace. In the Schuermann cupola waste gases are removed by draft from tuyere level through the row of tuyeres located opposite the air inlet. This cupola ran successfully on 9 to 10 per cent coke and 40 to 60 per cent steel scrap, melting consistently good quality hot iron. With proper care, this acid-lined Schuermann cupola was satisfactory in operation producing high-grade iron for internal combustion engine cylinders, machine tool castings, etc., with a charge containing small amounts or no pig iron at all. This meant that the costs of the charge could be reduced by $5 to $8 per ton. In addition to a saving in the melting operation, there was improvement in quality of the metal, especially with regard to strength.

Since a basic-lined 5-ton capacity electric-arc furnace was being operated at the same time, a comparison with acid-slag operation and a study of the effects which various slag practices, applied to the basic-lined furnace, had upon the iron produced was of interest. Influence of various slags on properties of the basic electric furnace iron is shown in Figure 4 and Table 1.

These tests run in a basic-lined electric furnace under uniform melting conditions, showed that composition of the slag plays an important role in the electric furnace. Advantages of superheating the melt can be completely offset by deficiencies in slag control. This suggested applying basic slag practice also in the cupola.

After the second world war a number of basic-lined cupolas were put in operation in the United States, Britain, and Italy in order to be able to charge a high proportion of scrap and thereby decrease costs of the charge and to obtain an iron low in sulphur for the production of nodular iron.

Results of these tests showed that metallurgically, the operation was highly promising, but that a basic lining is relatively expensive because of its poor erosion resistance. Moreover, the question of refractories for the tap-hole and slag notch has not been settled to date, because of the high rate of wear of the refractories used so far, some of which are very expensive. Mention should also be made of papers published in American technical literature during the past two to three years, especially of those by Holt,[5] Levi,[6] and Frazell and Sheley.[7]

To overcome these difficulties, Ford Motor Co. in Detroit probably was the first to attempt operating an unlined cupola using a basic slag. The basic lining was replaced by water-cooled cast steel boxes arranged at and above tuyere level, the service life of which is theoretically unlimited. This cupola operated satisfactorily for more than two years. The first step was thus made in the development of a cupola having an unlined stack which enables continuous operation.

Stimulated by the great amount of work done in America on hot-blast cupola operation, the author, about two years ago, installed a basic-lined hot-blast cupola. Apart from our metallurgical experience with a basic-lined electric furnace, we drew upon the experiences of a few Italian and British foundries concerning the metallurgical and thermotechnical potentialities of a basic-lined cold- or hot-blast cupola.

TABLE 1 — SLAG COMPOSITION ON TAKING SAMPLE 4 (FIG. 4)

Type of Slag	Composition, % SiO_2	Al_2O_3	CaO	MgO	Fe_2O_3	MnO	S
A. Aluminiferous	47.7	10.8	16.9	8.55	4.80	0.36	0.55
B. Calcareous	5.3	0.3	71.5	4.90	6.80	—	0.26
C. Lime, fluorspar, coke	21.8	9.4	54.8	9.98	2.00	0.20	1.62
D. Limestone (Stage D_2)	25.2	3.54	52.3	9.00	2.48	0.68	1.51
E. Carbide Slag	2.48	0.12	69.3	3.62	2.48	0.09	0.60
F. Carbide Slag	6.78	6.80	65.8	12.81	2.40	0.19	0.71

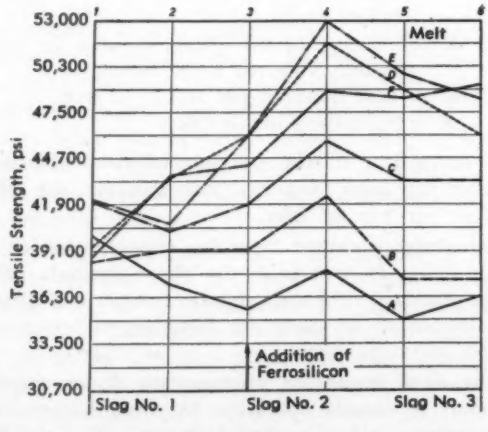


Fig. 4 — Tensile strength in relation to slag practice (Piwowarsky and Heinrichs[4]).

Only the Doat metallurgical blast cupola operates on a similar principle. Since little information was published on metallurgical and thermotechnical experiences, we were faced with fundamental development work.

A few months before our plant was in operation, two similar cupolas were put in operation in two West German steelworks. These two cupolas were intended to supply hot metal to open-hearth furnaces, the object being to produce, from a charge containing the maximum proportion of scrap, a hot metal low in sulphur, phosphorus, and silicon, of a constituency which would allow an increase in the output of the furnaces.

In the gray iron foundry, however, where a high-quality cast iron must be poured directly into sand molds, the composition of the iron must be kept within definite ranges, especially with regard to carbon and silicon. In addition, shop practice requires continuous operation of the cupola which can be achieved only when a number of requirements has been met.

Since all these requirements had to be met in our foundry, we selected for our cupola a neutral lining using carbon brick. This enabled us to run the cupola on an acid, neutral, or basic slag. However, the information available to us on basic slag practice was insufficient to assure obtaining the desired carbon and silicon content in the liquid metal.

After six months of basic slag practice in a hot-blast cupola, we have not gathered enough experience to form or corroborate definite conclusions. In examining the chemical reactions occurring in the basic hot-blast cupola we have referred to a paper by Fleming,[8] "Inter-relation of Combustion and Metallurgical Reactions in the Cupola," and incorporated Fleming's findings in our considerations and conclusions.

In his paper, Fleming dealt with combustion processes in the acid and basic hot-blast cupola and with the carbon reactions. He not only studied the results obtained by Jungbluth and Korschan on acid-cupola practice, but also included the hot-blast cupola in his considerations.

He summarizes conditions governing the reaction temperature as follows:

With a given type of fuel and at a constant combustion rate, it is possible to consider the heat balance at a plane in the furnace which is at a definite reaction temperature T_R by drawing up the total heat balance for this plane, provided that the fuel, burden, and gas phase are all at temperature T_R in this plane. By the reaction temperature T_R is thus meant that point at which the $CO:CO_2$ ratio can no longer undergo any change. The heat balance is then that for the whole of the reactions occurring in the reaction zone up to (in the physical sense) the temperature level T_R in the reduction zone.

The heat supplied to this zone is then made up as follows (Fig. 5):

a) the total heating value of the fuel,

b) the heat content of the fuel at temperature T_R

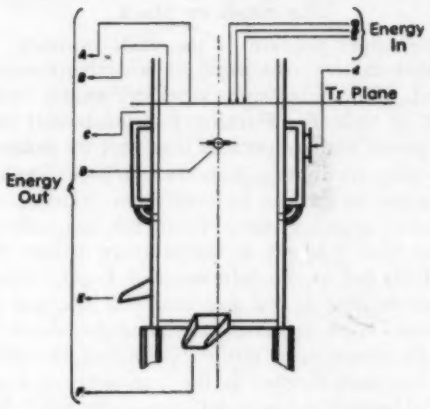


Fig. 5 — Cupola heat balance at plane of reaction temperature T_R.

c) the heat content of the charge at temperature T_R,

d) the heat content of the blast.

The heat loss is this zone is made up as follows:

a) heating value of the gas generated (composition in equilibrium with temperature T_R,

b) enthalpy of the gas generated at temperature T_R,

c) heat loss by radiation, conduction, and that removed by cooling water,

d) heat required for slag formation and metal oxidation,

e) heat content of the tapped slag including the heat of fusion at temperature T_R, and

f) heat content of the tapped iron including the heat of fusion at temperature T_R

For this concept of heat balance in this reference plane, an equation can be set up which enables the different heat energies to be plotted in relation to temperature T_R. This equation then enables the corresponding values of T_R to be computed for a variety of furnace data, and a graph can be obtained for the reaction temperatures up to point T_R = 800 C, (1470 F) and for the various fuel ratios, as may be seen from Fig. 6. The graph in Fig. 6 has been calculated by Gumz[9] for cold blast and blast temperatures of 300 and 600 C (570 and 1110 F). The curves by Gumz agree with results obtained in experimental work by Jungbluth and Heller,[10] Bader,[3] and De Bock.[11]

In his theoretical studies of the acid-lined cold-blast cupola, Fleming has divided the cupola into three sections:

1. The shaft or stack (i.e. from charging level down to the level of the reaction temperature at which the temperature of the charge is raised from the normal temperature to the temperature T_R and at which the $CO:CO_2$ ratio, which again depends on the temperature T_R, remains constant),

2. The zone above the tuyeres, and

3. The zone below the tuyeres.

The Shaft or Stack

In the upper portion of the shaft or stack, water and other matter volatilized at low temperatures is removed. To determine whether oxides will be formed or reduced, Fleming first examined the reaction plane with respect to iron and its oxides.

The diagram in Fig. 7 shows the reduction of the iron oxides by carbon monoxide in relation to the Boudouard reaction curve. From this diagram it can be seen that iron at a temperature below 575 C (1070 F) is led to the formation of Fe_3O_4, since the CO content lies at 8.2 per cent, i.e. the gas causes oxidation. With increasing temperature, above 600 C (1115 F), there is a tendency toward deoxidation, which increases further as the temperature is raised. To what extent there is sufficient time available for complete reduction in the gas phase is questionable in view of the speed at which gases permeate through the charge.

In this connection, the effect the furnace atmosphere has on silicon and manganese may briefly be mentioned.

Silicon can only be oxidized in the stack since, with temperatures available in the stack, the highest CO: CO_2 ratios obtainable will still be too low for reduction of SiO_2 to Si.

Manganese, likewise, can only undergo oxidation since the reduction occurring in the cooler region of the stack will be reversed as the reaction temperature is reached.

Zone Above the Tuyeres

The chemical reactions in the melting zone will always produce oxidation, since the air blast provides an excess of oxygen. The degree of oxidation is varied not only by chemical, but also by physical components such as fuel quantity, fuel size, and blast temperature.

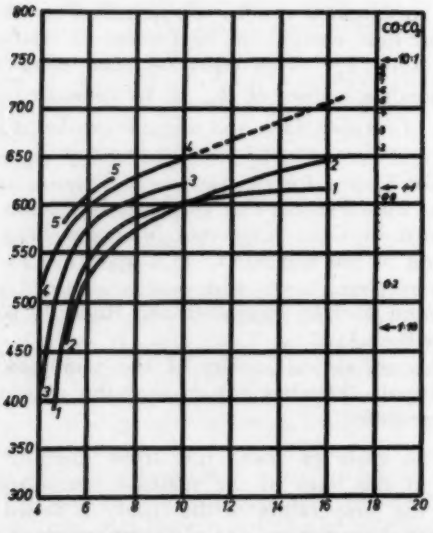


Fig. 6 — Inter-relation between temperature T_R and fuel ratio: 1. Cold Blast (Gumz); 2. Cold Blast (Jungbluth and Heller); 3. Blast Temperature 570 F (300 C) (Gumz); 4. Blast Temperature 1110 F (600 C) (Gumz); 5. Blast Temperature 1110 F (600 C) (Bader); 6. Blast Temperature 930 F (500 C) (DeBock).

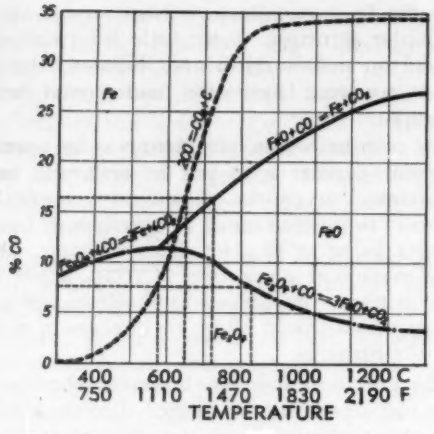


Fig. 7 — Reduction of ferrous oxides by CO in relation to Boudouard reaction curve.

Since the reaction rate depends on the surface area of the fuel which reacts with unit volume of gas, it increases at high temperatures, not only due to a decrease in fuel size, but is in some cases also influenced by the cellular structure of the fuel. From the network of the remaining ash it is apparent how strongly the rate of reaction is affected by the cellular structure.

The velocity of the air blast affects the rate at which the reaction takes place:

1. by the cooling effect of the air. The higher the velocity of the air in the furnace, the stronger will be the cooling effect in the middle temperature zone, which will thus slow down the chemical reaction rate.

2. by higher turbulence in the regions of higher temperatures where the physical factors predominate. With increasing blast velocity the contact time will be shortened, thus preventing a vigorous reaction taking place.

At the upper limit of the oxidizing zone, the reaction $CO_2 + C = 2\,CO$ may be expected to proceed at a slower rate because of a decreasing number of actual contacts between the two reactants with only a small amount of oxygen remaining in the gas phase.

Figure 8 (after Fleming) illustrates the conditions in a cupola. Areas 1 and 2 show the zones where cooling occurs due to the incoming blast and the restrictive limits of the free-oxygen zones. As already mentioned, the point where free oxygen ceases to exist is also the point of the highest flame temperature. In an operating cupola, no downward gas flow can be detected, and the main flow of gas will be upward, as indicated by the arrows. The arrows also show the direction of the heat carried by the gas stream.

The region enclosed by the broken line (between and above the boundaries of area 2) is the peak temperature zone and, consequently, the zone of CO formation. Above the upper boundary, the heat will be used to melt the metal and the slag components. In this upward stream of gas of almost uniform flow rate, the CO_2 content decreases while the CO content increases until the upper limit of the reduction zone (section ABC and EH) is reached.

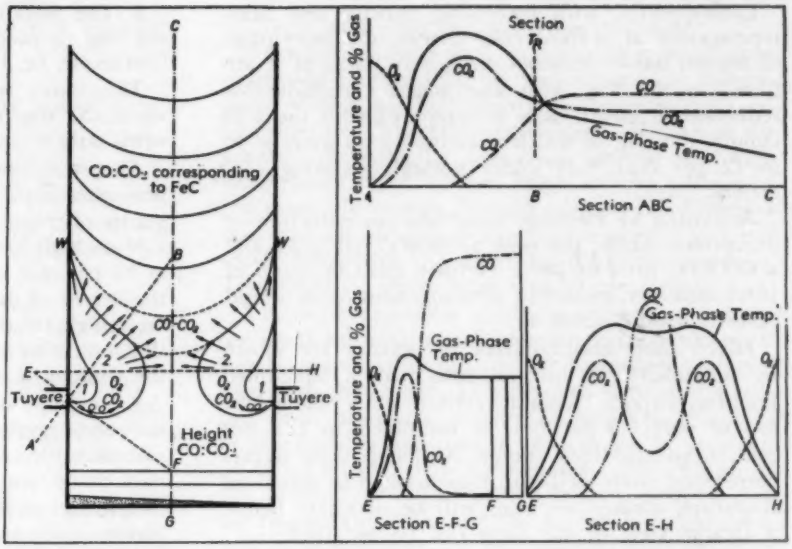


Fig. 8 — Representation of CO:CO_2 ratios obtained in various cupola zones.

Zone Below the Tuyeres

This zone, in which free oxygen can no longer exist and in which a high temperature is coupled with a very high CO:CO_2 ratio, is perhaps the most important in the whole cupola. Here, reducing conditions again exist and the oxidation which has occurred is reversed to a certain degree. The degree of this reversal, coupled with the final metal-slag reactions will determine the final melting losses of the cupola. Fortunately, the principal factors governing efficiency of this zone are fairly obvious: they are the high temperature and the high CO:CO_2 ratio.

In the region below the tuyeres, graphically represented by the section EFG, heat is removed by radiation and conduction and by the descending column of molten material. In this range the gas composition will consist mainly of CO. The conversion of existing CO_2 to CO will require little heat and it can be assumed that the ratio CO:CO_2 almost approaches equilibrium with the temperature reached.

Now, since no gas flow can be assumed to exist in this zone, the optimum efficiency of this zone is dependent only upon temperature. In and below the slag, the reactions occur at the slag/coke interface and at the metal/coke interface. A proper well depth appears to be of importance in order to keep, on the one hand, free oxygen or carbon dioxide from re-oxidizing the slag components at this interface, and on the other hand, to prevent unnecessary heat losses by an excessive depth of the well.

Maintaining an adequate and uniform slag depth is important to ensure sufficient time for the reaction between the slag and the falling drops of metal. It is also obvious that a fluid slag is preferable to a viscous slag and that, wherever possible, complex and stable compounds of the slag with the oxides to be reduced should be avoided and that the slag should contain catalytic materials which accelerate the reaction. The slag must combine with the ash network remaining after combustion of the coke. The slag should therefore be examined for its effectiveness with respect to the reaction, because the reaction can take place only either by direct oxidation of CO to CO_2 or indirectly as oxide/slag/coke reaction.

The reductions occurring in this zone under this high CO:CO_2 ratio increase in intensity with increasing temperature, reduction of ferrous oxide already taking place at 1800 C (3270 F), at a CO:CO_2 ratio of 10:1. In the basic-lined hot-blast cupola, much of this reduction takes place.

With proper furnace control, the slag produced is almost white, resembling carbide, and contains only 0.5 to 1.0 per cent FeO, as can be seen from slag analyses in Fig. 9.

Color:	Blackish-brown	Blackish-gray	Brownish-gray
% S in Iron:	0.111	0.089	0.065
Corresponding Slag Analysis:	12.5 % Al_2O_3 38.0 % SiO_2 39.6 % CaO 1.83% MnO 0.8 % MgO 2.5 % FeO 0.69% S	12.6 % Al_2O_3 39.0 % SiO_2 38.1 % CaO 2.22% MnO 0.6 % MgO 2.0 % FeO 0.73% S	6.4 % Al_2O_3 28.9 % SiO_2 47.5 % CaO 1.10% MnO 0.6 % MgO 1.0 % FeO 1.04% S
Color:	Grayish-Beige	White	Resembling Carbide
% S in Iron:	0.43%	0.033	0.022
Corresponding Slag Analysis:	8.5 % Al_2O_3 29.5 % SiO_2 51.6 % CaO 0.79% MnO 0.5 % MgO 0.7 % FeO 1.37% S	9.0 % Al_2O_3 29.0 % SiO_2 51.2 % CaO 0.61% MnO 1.1 % MgO 0.8 % FeO 1.27% S	6.8 % Al_2O_3 27.0 % SiO_2 51.4 % CaO 0.55% MnO 0.9 % MgO 0.6 % FeO 1.10% S

Fig. 9 — Analysis of some basic slags.

Consequently, with increasing basicity and blast temperature at a fixed coke charge, the percentage of ferrous oxide decreases until only traces of it are found in the slag. This also means a considerable reduction in metal losses as compared with the acid cupola, the slag of which contains on an average 10 to 12 per cent FeO under normal operating conditions.

According to Fleming, there was no reduction of manganous oxide, not even at 1800 C (3270 F) and a $CO:CO_2$ ratio of 10^3:1. With a $CO:CO_2$ ratio of 10^4:1, however, reduction of manganous oxide would occur at 1360 C (2480 F).

Effects these findings have in practice are shown by Schmidt.[12] In his test runs with a basic-lined hot-blast cupola, Schmidt arrived at a manganese loss of only 3.5 per cent as compared to 22.8 per cent in an acid-lined cupola. Although these figures correspond fairly well with those quoted in American literature, further test runs will be necessary before a clearer view of the facts can be obtained.

Reduction of silicon from silica is also possible with the conversion of carbon to carbon monoxide $2C + O_2 = 2CO$, at temperatures above 1540 C (2800 F) and a $CO:CO_2$ ratio of $(1.8 \times 10^4) : 1$, whereas reduction by the complementary oxidation of CO to CO_2 would require a $CO:CO_2$ ratio of more than 10^5:1 at 1300 C (2370 F) or 10^4:1 at 1620 C (2950 F).

In acid hot blast practice, the theoretically difficult reduction of SiO_2 goes almost to completion in the slag zone, in fact at times leads to a silicon gain. In the basic furnace, however, despite continual reference in the literature to a greater silicon loss, we have been able to develop a certain control. Thus, operating for a week with an iron of 1.5 per cent silicon content, we have reduced the silicon loss to 6 to 9 per cent, whereas H. Schmidt in his previously mentioned work obtained an Si-loss of 34.4 per cent in a basic operation versus a 17 per cent loss in acid. In these experiments varying coke charges and blast volumes were used and hence no clear-cut reason for the high Si-losses could be determined.

In summary, when considering the reactions for iron, manganese, and silicon in the cupola hearth, it was found that all three can be reduced under possible cupola conditions (in the sequence in which they are named) and that reduction is favored by increasing hearth temperatures.

Desulphurization

Very little can be done about desulphurization in an acid-lined cupola. Many attempts have been made to use a higher proportion of limestone in the cold-blast cupola or to encrust the coke with lime. However, nothing is gained by this because of the considerably higher consumption of and wear on the lining. If, however, the neutralizing effect of the lining is eliminated and the cupola operated on a basic slag, the sulphur content can be consistently kept low. Desulphurization can proceed in two ways:

1. The slag dissolves the sulphides from the metal until the amount of these sulphides is in equilibrium with the amount in the metal, i.e. FeS (metal) = FeS (slag)

2. The metallic sulphides react with limestone in the slag to produce calcium sulphide together with iron oxide, i.e. $FeS + CaO \rightarrow FeO + CaS$.

The latter reaction can be fully effective only when the slag contains an excess of limestone and when only a low per cent of iron oxides exists.

Almost all the sulphur contained in the drops of iron goes into the slag, and while melting several grades of iron in current shop practice, we obtained a Si as high as 1.7 per cent in the slag and as low as 0.1 per cent in the iron tapped. In this connection, the degree of desulphurization is not only dependent on a high temperature and degree of basicity of the slag, but other factors are also important, for instance, the content of magnesium oxide in the slag.

A thorough investigation was made of the individual and aggregate effect of the slag components silicon oxide, calcium oxide, aluminum oxide, ferrous oxide, and manganous oxide by Drachmann.[13] According to his findings, ferrous oxide and manganous oxide have no direct influence on the sulphur index. Based on knowledge concerning the three desulphurizing factors (silicon oxide, aluminum oxide, and calcium oxide), curves can be drawn up in a "ternary" diagram by means of the regressive equation. In his considerations, Drachmann came to the conclusion that, compared with silica and lime, the alumina content is of only little influence and that the $CaO:SiO_2$ ratio is an indication of the influence the composition of the slag has on the sulphur index.

However, Drachmann in his work on desulphurization considered only the basicity ratio $CaO:SiO_2$ and attached no importance to magnesium oxide.

We agree with Fleming that the optimum degree of desulphurization can be obtained in the basic-operated cupola and that a water-cooled furnace is the most economical type for this purpose.

Dephosphorization

Not much can be said about dephosphorization, although with certain high-phosphorus charges we have, in some cases, observed considerable decreases in phosphorus. However, at this stage we cannot give detailed information and data on the conditions governing these reactions.

Carbon Pickup

In regard to carbon control, it has been shown by many workers that a higher carbon pickup can be obtained more easily in cupolas operating with basic slags. It is believed that this is due to more rapid removal of the ash network which is acid in character, and by which means the carbon surface is kept active. It is also believed that the non-wetting character of the basic slag with respect to the coke assists in this direction. Cupola operation with a controlled basic slag is, therefore, best suited for use where a high degree of carbon pickup is required. The extent of the carbon pickup obtained and, hence, the final carbon content of the melt depend on:

1. the original carbon content of the charge,
2. the temperature of the melt,
3. the iron-coke ratio, concordant with the points just mentioned,

4. the iron-oxide content of the slag,
5. the coke-ash percentage and the type of coke used,
6. the basicity of the slag,
7. the extent of the free-oxygen zone in the tuyere areas, and
8. the blast temperature.

At first sight, this appears to be a considerable list of variables, but in a steadily running installation with a given type of burden, most of these variables can be kept constant.

From a metallurgical point of view, in basic hot-blast operation it is important that the operator can regulate the desired carbon content and C:Si ratio (the sum of C plus Si), with certainty of producing a high-grade cast iron with high tensile strength and good transverse strength. We found no difficulty in meeting these requirements if special attention is paid to the charge, the blast volume and blast temperature. We are thus able to produce high-grade cast iron for machinery castings with a tensile strength of more than 26 kg per sq mm (37,000 psi) containing the desired carbon and silicon.

In summarizing the ideal melting requirements of a basic hot-blast cupola for melting soft gray iron, the following should be kept in mind:

Iron, silicon, and manganese losses can be kept at a minimum.

It is possible to produce an iron with a high, medium, or low carbon content.

It is possible to produce low-sulphur iron containing as little as 0.01 per cent S.

To obtain these conditions, the cupola must be run at a high reaction temperature:

The reaction zone, especially the free-oxygen zone, must be kept physically small and generate a high temperature to obtain a hot metal.

The slag must be sufficiently basic to favor desulphurization and carbon pickup. The slag must depend on materials charged only, i.e. coke ash and fluxes must not sporadically attack the lining.

The slag depth must be constant to ensure constancy of metal/slag and slag/coke reactions.

Thermal efficiency can, in spite of a lower coke charge, be increased by the use of hot blast as compared with a standard cupola operating at the same temperature. This offers the simplest solution to the problem of obtaining a reaction zone of very high temperature and minimum height in which free oxygen exists. It appears that the effect in this direction rises with an increasing blast temperature. The latter should be as high as present day recuperators permit. The economic aspect must also be considered.

The requirement for a sufficient degree of basicity for desulphurization and adequate carbon pickup is met by proper adjustment of the percentage of fluxes used with the charge. This depends only on the burden, since the neutral graphite-bearing lining requires no slag-forming constituents, and can best be obtained by using a completely water-cooled cupola in which the melting and tuyere zone is water-cooled, so that the tuyeres also withstand the temperatures required.

The requirement for an adequate slag depth is met by the use of continuous tapping through a syphon, since the slag depth is determined by the level of the external metal dam above the tap-hole as is usual for systems where both iron and slag are tapped continuously.

Operation of Basic Hot-Blast Cupola

Following the theoretical considerations, let us now turn to the actual operation of our basic hot-blast cupola (Figs. 10a, b, c).

The shaft or stack of the author's cupola has an ID of 1.94 m (76 in.), the thickness of the shell being 3 cm (1.2 in.). The initial lining gave the hearth an ID of 1.2 m (47 in.) plus 65 mm (2½ in.) for the protective layer of fireclay brick. The carbon brick bottom is 62.5 cm (25 in.) thick. The distance between hearth bottom and the underside of the tuyeres is 1.1 m (43 in.). Measured from the bottom, the rammed up carbon lining is 3.37 m (133 in.) high.

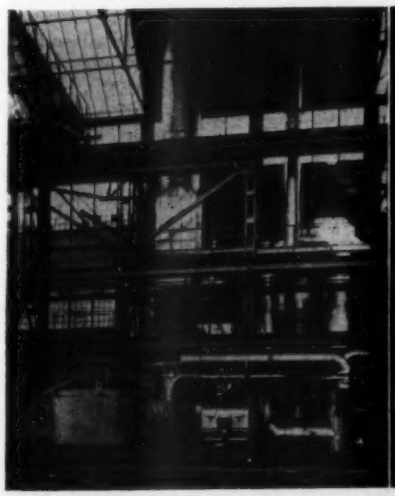


Fig. 10A — Hot blast cupola plant.

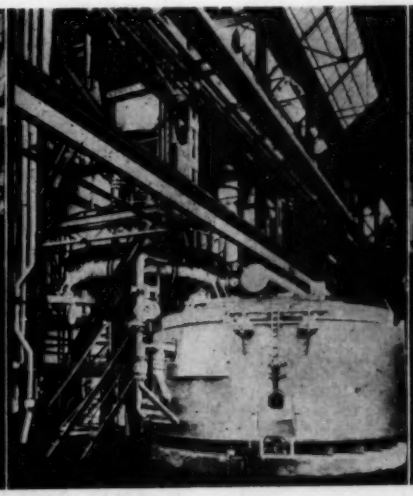


Fig. 10B — Side view of hot-blast cupola with gas-fired 35-ton metal receiver.

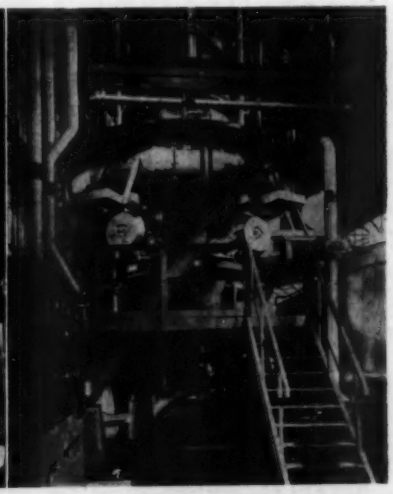


Fig. 10C — Side view of hot-blast cupola with control panel.

From this level on, the shaft or stack is lined with fireclay brick up to the top-gas offtake, this section being 2.37 m (93 in.) long. Below charging level, the stack is protected by 1.435 m (57 in.) long stockline armor 1.25 m (49 in.) in diameter. Overall height of the cupola is approximately 24 m (79 ft). Stack and other waste-gas mains are lined with light fireclay brick.

The cupola has six water-cooled tuyeres, each 61.5 cm (24 in.) long with a diameter of 18 cm (7 in.). Hot-blast enters the cupola from a wind belt through a tuyere stock of 13.5 cm (5.3 in.) diameter. The diameter at the tuyeres can be varied from 98 to 140 cm (39 to 55 in.) by which means the melting rate can be adjusted between 6 and 11 tons per hour.

The section of the shell between well and top-gas offtake is cooled by four cooling belts.

Hot blast is supplied by a recuperator. With a maximum fire-box temperature of 900 C (1650 F), this recuperator is capable of delivering 6000 cu m/hr (3,530 cfm) at the maximum permissible hot-blast temperature of 600 C (1110 F). The average top-gas temperature is 400 C (750 F). A rotary blower arranged on the intake end of the recuperator is rated for 8000 cu m/hr (4,700 cfm). The necessary combustion and cooling air is provided by a fan with a capacity of 5000 cu m/hr (2,940 cfm).

The top gases used for preheating the recuperator are exhausted through a gas take-off belt below the stockline armor. The exhaustor arranged on the discharge end of the recuperator has a capacity of 20,000 cu m/hr (11,770 cfm). The vacuum developed in the piping and the recuperator is approximately 120 mm (0.47 in.). The volume of top-gas desired is coarsely regulated by a top-gas valve, fine control being provided by butterfly valves arranged on the discharge end of the recuperator.

The total water consumption of the cupola ranges between 90 and 110 cu m/hr (330 and 400 gpm). Our installation does not permit all of the cooling water to be recovered, as some of it is used for granulating the slag (approx. 20 cu m/hr or 70 gpm). Two pumps with a capacity of 240 cu m/hr (885 gpm) are arranged at the supply main to ensure adequate cooling. Return water is pumped by another two pumps with an aggregate capacity of 140 cu m/hr (520 gpm).

To supervise and control the operation and to gather further data for metallurgical developments, the following instrumentation is provided to obtain a constant record of the variables mentioned below:

1. blast volume,
2. blast pressure,
3. hot-blast temperature at the cupola inlet and at the recuperator,
4. waste-gas volume leaving cupola and recuperator,
5. waste-gas temperature (recuperator),
6. fire-box temperature (recuperator),
7. top-gas temperature,
8. $CO:CO_2$ content,
9. H_2 content (at the same time as a precautionary measure against formation of H_2 in the cupola),
10. water temperature (at tuyere outlets and at the shell cooling belts),
11a. total quantity of cooling water while melting, and
11b. total quantity of cooling water while damping.

At first sight, this appears to be a formidable list of process measurements. However, this degree of instrumentation is necessary because of the large number of reactions occurring in the furnace which cannot be determined off-hand for each cupola zone.

The plant is required to produce every desired grade of iron in a sufficient quantity to meet the needs of our foundry. This requires exact cupola control, since the specified grades often differ only slightly, especially with regard to carbon and silicon content. In addition, the percentages of silicon, phosphorus, and manganese are decisive factors in the production of nodular cast iron obtained from a basic-operated cupola.

Another important factor is that our basic hot-blast plant is to be operated continuously to utilize its inherent economic advantages, although, normally, the foundry operates on a one-shift basis. Continuous operation will very much decrease the high rate of amortization. From information obtained on the same cupolas operating in steelworks, we know that there is a relationship of about 2:1 between the cost per ton of liquid metal produced during an 8-hr day and an 24-hr day.

We cannot give any definite cost per ton of liquid metal produced at this stage, since these costs depend on such variables as charge coke, cost of cooling water, metallic charge, type and quantity of fluxes, on which we have not yet been able to obtain conclusive information within the relatively short period of our operation. According to our present experience, the pecentage of charge coke can be kept between 12 and 18 per cent depending on the type of charge used, especially the proportion of steel scrap in the charge. Fluxes, however, depend on coke charge as well as on the metallurgical conditions under which the cupola must be run for a given grade of iron.

According to operating experiences gathered in the steelworks, no insurmountable difficulties will be encountered in continuous operation, provided the problem of the slag separator (the piece of equipment most liable to breakdown in continuous operation) is solved. In several days of continuous operation, we have obtained satisfactory results and confirmed the possibility of continuous working.

The metal is separated from the slag in a syphon box and flows continuously into the holding ladles, one holding 5 and the other 35 tons. The slag is immediately granulated and flushed into a settling basin. Normally, the syphon box lasts for a week and must then be removed for repairs which can be carried out without difficulty.

Above the tuyere level, the cupola has no refractory lining and need therefore no longer be emptied on weekends. Need for lining repairs is eliminated. Like the blast furnace, the cupola can be throttled down as required. After melting down and tapping the re-

maining metal and slag, bed coke must be charged and maintained at the normal level while the air blast must be cut off. The furnace can be run in the throttled down condition as long as is necessary. On starting up, a suitable amount of bed coke must be charged, after which charging and blowing can be resumed. During the period of damping, the flow of cooling water around the tuyeres and the shell must be maintained.

To obtain the necessary high blast temperature in continuous operation requires careful cleaning of the recuperator. A knowledge of the level of the melting zone and of the metal and slag baths and experience on temperature control are of particular importance. Figure 9 shows the composition of various types of slag encountered in basic slag practice.

Test runs carried out by the author, some of which were continued over several shifts, were intended for the study and establishing the control of the cupola for melting every grade of iron required in our foundry, including the grades to be converted into nodular iron.

At this stage, it may be stated that during the initial 1½ years of operation, we experienced no considerable breakdown of the cupola installation, not even of the recuperator.

After 18 months of actual operation, it is a well established fact that with this carbon-lined, hot-blast cupola we are able to produce economically a number of grades of iron required to meet, after completion of our foundry expansion program, our total requirements of liquid metal (5000 to 6000 tons per month) by operating the plant continuously.

The following research and development work is being carried out by the author to obtain perfect control of the melting operation in the hot-blast cupola:

1. A thermodynamic study to set up a heat balance and to determine what further use could be made of the top gases not required for preheating in the recuperator.

2. The influence of hot-blast operation on the trend of the curves in the cupola-behavior diagram by Jungbluth and Korschan.

3. Investigation of physical and chemical properties of the metal melted under a basic slag.

4. Exact determination of the inter-relation between losses and increases, respectively, of carbon, silicon, manganese, phosphorus and sulphur and furnace control and slag practice.

5. Solubility of hydrogen, oxygen, carbon monoxide, carbon dioxide, and nitrogen in the iron in relation to the basicity of the slag and to the melting temperature.

6. Comparison of the iron melted in acid and basic slag practice with regard to:

a) solubility of the gases,

b) castability, shrinkage, and piping,

c) physical and technological properties of the metal.

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DISCUSSION

Chairman: DR. A. E. SCHUH, U.S. Pipe & Foundry Co., Burlington, N. J.

Co-Chairman: R. A. CLARK, Electro Metallurgical Co., Union Carbide & Carbon Corp., Detroit, Mich.

D. E. KRAUSE[1] *(Written Discussion):* This paper quite well summarizes the operation of hot blast cupolas and the increased versatility of operation with a water-cooled cupola. There are a few statements which might need clarification and for economic reasons hot-blast operation in America might be somewhat different from that followed in the author's plant.

Some of the captions in Fig. 1 appear to be in error. The blast volume in cfm appears to have a zero missing so that 210 cu ft/min should read 2100 cu ft/min. It is also felt that carbon-iron ratio should read carbon- per cent of charge.

The marked influence of slag composition on the tensile strength of iron melted in a basic electric furnace as shown in Fig. 4 has not been observed in cupola operation with various slags providing the composition of the iron is the same. Iron "A", sample 4 in Fig. 4 has a carbon equivalent of 3.82 per cent. Since the slag composition for sample 4, iron "A" is quite similar to an acid cupola slag, the strength of an iron of similar analysis melted in an acid cupola would have a tensile strength of about 42,000 psi instead of approximately 38,000 psi as shown. Iron "E", Sample 4 has a carbon equivalent of 3.60 per cent which in acid cupola melting would give a tensile strength of about 49,000 psi with a silicon inoculation. Some of the marked effects of slag composition shown in Fig. 4 may therefore be attributed to basic electric furnace melting conditions and may apply to only a limited extent in cupola melting operations.

The statement is also made that slag from an acid cupola contains from 10 to 12 per cent FeO under normal operating conditions. Possibly as a result of better coke quality in America, the FeO content of the slag from acid cupolas melting an ASTM class 30 iron will more likely be in the range of 3 to 6 per cent instead of 10 to 12 per cent.

A coke charge of 12 to 18 per cent of the metal charge seems rather high with a blast temperature of 1100 F unless the coke is of much lower fixed carbon content than is customarily obtained in America and the amount of steel in the charge quite high.

Probably as a result of a more favorable ratio of cost of gas and oil to coke in America than in Germany, the majority of the hot blasts in this country are externally fired. For a blast temperature of 500 F, the cost of firing the heater with gas in the Middle West is about 10 cents/ton of iron melted at a melting rate of about 20 ton/hr. The use of gas or oil for firing the heater eliminates the need for cleaning the heater as is the

1. Technical Director, Gray Iron Research Institute, Inc., Columbus, Ohio.

case when gases from the cupola stack are used. The use of gas or oil as a fuel also eliminates a great deal of expensive duct work and also results in a much more compact unit.

S. F. CARTER[2] *(Written Discussion):* I would like to congratulate the author on an excellent paper and a valuable contribution to the foundry literature. He has presented some useful and interesting theoretical data on cupola behavior. His cupola incorporates hot blast, basic slag, and continuous operation, three most progressive developments in cupola practice.

I have only a few things to confirm and a few to question:

Mr. Heinrich reported FeO contents in acid slags as high as 10 to 12 per cent. This is much higher than generally encountered in America, where 1 to 5 per cent FeO is usual and only 5 to 8 per cent FeO might be found in special cases with high steel charges or coke ratios around 14 to 1. Possibly charge materials or coke are different in the author's country.

Mr. Heinrich indicated dephosphorization was uncertain. Possibly some dephosphorization data in my paper "Basic-Lined Cupola for Iron Melting" in AFS TRANSACTIONS, vol. 58, pp. 376-392 (1950) might be helpful. A number of various mixes and flux combinations were found successful under oxidizing conditions, low silicon contents, and low melting temperatures. Typical phosphorus contents were lowered from 0.79 to 0.49 per cent, from 0.58 to 0.28 per cent, and from 0.38 to 0.12 per cent. However, at higher temperatures and more reducing conditions favorable to good casting practice, little or no phosphorus was removed.

I question the author's statement, quoting Drachmann, that ferrous oxide has no influence on desulfurization. Our experience, as indicated in papers of 1950 and 1952 (AFS TRANSACTIONS), suggest that FeO content is one of the important factors in desulfurizing efficiency of basic slags, along with basicity, fluidity, temperature, etc. We consider magnesia along with lime in the degree of basicity.

We have had no experience with blast temperatures beyond 450 F, but verify the general tendencies of preheated blast described by the author. However, I would like to suggest the effect of cupola diameter. We find preheated blast far more influential on a 48-in. cupola than on an 84-in. cupola. We believe the great difference in the ratio of perimeter to cross-sectional area explains the difference.

In the author's cupola a melting rate of 6 to 11 tons per hour for 39- to 55-in. inside diameters and coke charges of 12 to 18 per cent are approximately what we would expect from a cold-blast cupola. With blast preheated to around 1100 F we would expect considerable reduction in coke and increase in output. Is that amount of coke required for the hot blast to function or is it necessary to counteract the cooling effect of the water? There is the further possibility of differences in coke quality. Does the author have comparative coke ratios and melting rates on a conventional cupola of comparable size?

Again, I would like to thank Mr. Heinrich for his valuable contribution to our cupola literature.

MILTON TILLEY[3] *(Written Discussion):* This excellent paper on the theory and practical use of heated cupola blast is greatly appreciated since at this time the foundry with which the writer is connected is planning on an installation.

Since the theory of what goes on inside the cupola remains somewhat in the realm of speculation, I will confine my remarks to exchange of views and questions as to the import of some points in the paper which are obscure to me, yet have a fundamental bearing on the subject.

Take the points, on which I seek enlightenment, in order of occurrence in the paper. The first advantage expected was "an increase in blast volume." Was this meant to be an increase in weight of air in spite of the increase in volume due to hotter air? In view of the fact that at 10 per cent coke and 35 per cent efficiency in cupola melting the iron would contain about 1,000,000 Btu I am at a loss to evaluate the heat content in Fig. 2 as expressed in calories per 2200 pounds.

The statement is made that "the chemical reactions in the melting zone will always produce oxidation, since the air blast provides an excess of oxygen." The wording would lead one to think that there was free oxygen present, but if the author means the excess oxygen is present as CO_2 I would agree. Item 2 under velocity of air blast, states that "with increasing blast velocity the contact time will be shortened, thus preventing a vigorous reaction." I cannot understand why the speed of any particular atom or molecule should affect the reaction as long as another is immediately available to carry on the reaction.

The statement is made that "the zone below the tuyeres is perhaps the most important." I have always considered the melting zone as the most important because there is where the quality of the metal is controlled with the zone below the tuyeres taking second place because of the relatively small effects taking place at the slag-metal interface. I agree that these reactions are taking place but the question is of the magnitude of the results on the economy and quality of the metal compared to what happens in the melting zone.

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RECENT DEVELOPMENT OF THE CORELESS LINE FREQUENCY INDUCTION MELTING FURNACE IN EUROPEAN FOUNDRIES

By

Otto Junker*

ABSTRACT

An outline of the development and present use of coreless line frequency induction furnaces is presented. Pioneered in Germany, this furnace is suitable for non-ferrous metals, including magnesium and its alloys. It is advantageous for turnings and chips. The furnace has found wide use for melting cast iron, particularly spheroidal graphite cast iron. More recently, it has also been used for steel melting. Various furnace types and applications are described and operating data given.

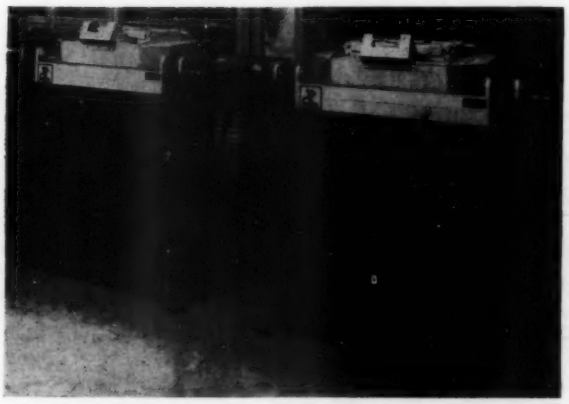
Induction melting furnaces may be classified broadly into the following groups:

1. Core-type line frequency induction furnaces.
2. Coreless high-frequency induction furnaces.
3. Coreless line frequency induction furnaces.

While the first two furnace types are well known in this country, and amply covered by existing literature, the third furnace type has only recently become established as a practical melting tool in European foundries. It is, however, not yet well known in the United States. The purpose of this paper is to acquaint American foundrymen with the construction, electrical and operating characteristics of the coreless line frequency induction furnace, as well as to present some actual operating data.

The coreless induction furnace operates on the principle of an air core transformer without a closed iron coupling. Its main characteristic is a simple melting chamber of cylindrical shape resembling the conventional crucible. It has no melting channels, or loops, and can be adapted to intermittent operation since no molten metal heel is required to remain in the furnace. Thus, alloy changes are easily made.

Due to the lack of ferromagnetic coupling the transmission of electrical energy from the primary coil to the molten metal charge is a much greater problem in the coreless furnace. This problem can be solved in two ways. One is to use a high frequency, usually in the range from 500 to 10,000 cycles; the other is to use a very high magnetic field density at ordinary line frequency of 50 or 60 cycles. Thus, a line frequency coreless furnace is usually operated with magnetic field densities in the neighborhood of 4000 ampere-turns per centimeter (amp/cm), whereas the magnetic field density used in high frequency furnaces is usually in the range of 500 to 1000 amp/cm.



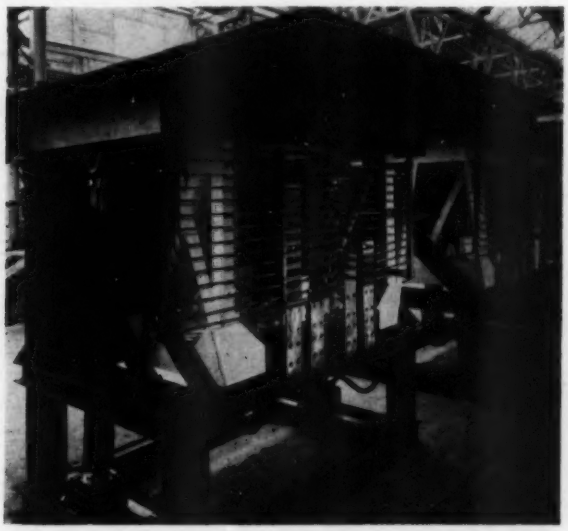


Fig. 1 (Above) — Front view of coreless line frequency induction furnaces for melting cast iron and steel.
Fig. 2 (Right) — Rear view of furnaces in Fig. 1.

*President, Otto Junker G.m.b.H. Lammersdorf, Germany.

The principal incentive for the development of a line frequency coreless induction furnace has been to eliminate the costly motor generator type frequency converter, and the energy losses attributable to that unit.

During the last 20 years, many unsuccessful attempts have been made to develop an economical coreless line frequency induction furnace. The failures of the past were due mainly to two causes:

In the first place, the coreless induction furnace requires large capacitor banks for power factor correction. Until recently, the price of these capacitors was prohibitive. However, improvements in manufacture and increased volume of production have considerably reduced this price.

Secondly, the bath movement in the melting crucible of line frequency coreless furnaces is inherently more violent than in the high-frequency type, due to the greater magnetic field strength required. Latest developments have overcome this difficulty by proper arrangement and dimensioning of the coil, and the bath movement can now be controlled so as to make the furnace a practical tool even for melting light metals.

As a result of these developments, coreless line frequency induction melting furnaces have been installed in many European foundries in recent years. Figures 1 and 2 illustrate these furnaces.

Furnace Characteristics

Electrical Considerations

It has been established that the capacity of these furnaces should be at least of the order of 1200 lb for iron and steel, and 800 lb for light metals. Smaller capacities would be unsuitable for line frequency connection and require operation with higher frequencies.

Since the power supply is usually on a three-phase basis, it would be logical to subdivide the furnace coil into three sections, each separately connected to one phase of that supply. Unfortunately, when attempting such a subdivision, it was found that the efficiency of each coil section was decreased. Also, larger capacitor banks were required for power factor correction. It was, therefore, concluded as a practical matter that coreless line frequency furnaces should be constructed as single-phase units. This is also the established practice with line frequency core-type induction melting furnaces in the United States.

Should there be a special requirement for a balanced three-phase load, a phase balancing system may be provided by connecting reactors and capacitors to the two other phases. The theoretical considerations leading to this balancing system have been described in several publications.* In group installations, the connection of a number of single-phase furnaces to the same power transformer practically eliminates the question of phase balancing.

Figure 3 shows cross-sectional views and electrical diagrams comparing a high-frequency furnace with a coreless line frequency furnace including a phase balancer. The line frequency furnace does not require a frequency converter. The picture also indicates that the current penetrates deeper into the molten metal in the line frequency furnace and that only part of the molten metal is surrounded by the coil.

While no ferromagnetic core is used in the coreless line frequency induction furnace to couple the primary with the secondary, the arrangement of a number of yokes of laminated silicon steel on the outside of the coil has been found advantageous. This provides a slight improvement in power factor, but the main purpose is to shield the supporting steel frame from stray magnetic fields, thus preventing the heating of the frame.

*See bibliography.

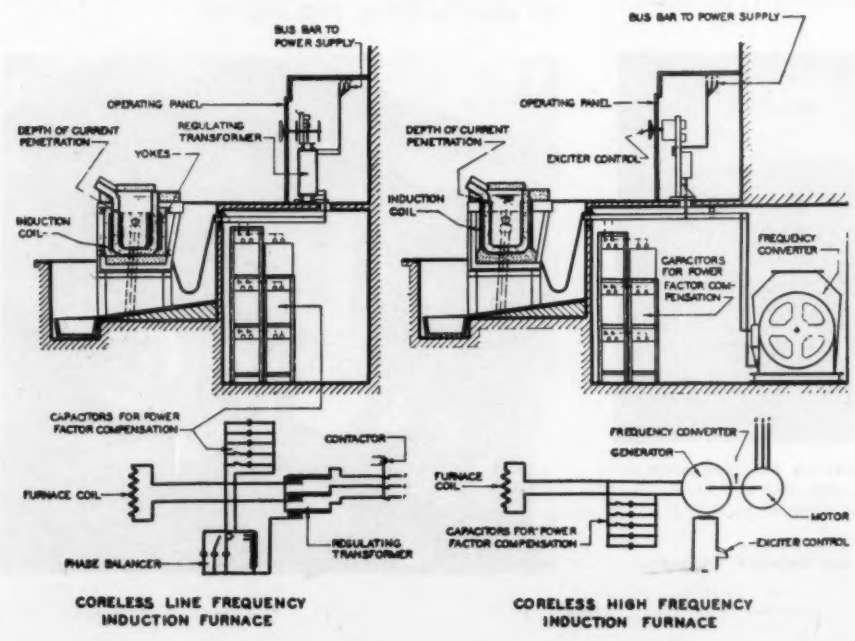


Fig. 3 — Schematic comparison of high frequency and line frequency coreless induction furnaces.



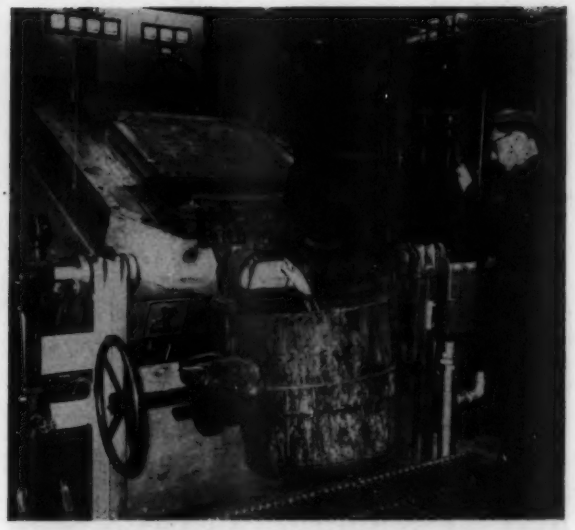


Fig. 4 (Above) — Coreless line frequency induction melting furnace for aluminum.
Fig. 5 (Right) — Pouring from the aluminum melting furnace.

The use of a lower frequency in this furnace results in a higher power factor, ranging from 0.15 to 0.17 when melting aluminum, and from 0.18 to 0.24 when melting iron. By comparison, the power factor of the high-frequency furnace ranges from 0.07 to 0.12.

Another advantage in many cases is the small space requirement of the line frequency furnace, due to the elimination of the frequency converter. Also, power consumption runs about 10 to 15 per cent lower than that of the high-frequency furnace using the frequency converter.

Operating Characteristics

Starting of the line frequency furnace may be achieved in two ways: When operating small size furnaces on an intermittent basis, they can be started by charging one solid block of metal (about 15 to 20 per cent of the crucible content) which has the approximate shape of the crucible. Such a block may be prepared at the end of a shift by casting ladle residues and other left-over metal into a suitable mold partly filled with miscellaneous small scrap, such as gates, risers, etc. Larger furnaces may be started by charging metal in large shapes directly into the crucible.

As a rule, the line frequency furnace requires charging larger lumps of cold metal than the high-frequency furnace. This, of course, applies only to the initial charge until a pool of molten metal is established.

The strong bath circulation inherent in the line frequency furnace is of particular advantage when melting turnings or chips which may be charged without briquetting. Circulation is also important in the recarburizing of steel and cast iron with pure carbon.

In the beginning it was feared that the strong bath circulation would be detrimental to the refractory lining. Operating experience over a period of four years on some 80 units with holding capacities up to 18,000 lb, has shown that lining life remained well within the range considered normal for high-frequency furnaces. As an average, one may count on a lining life of about 4000 charges when melting light metals, about 250 charges when melting cast iron, and about 130 charges when melting steel. The strain on the refractory furnace lining is not as great as one might expect, as the interaction of electro-magnetic forces is entirely between the current carrying induction coil and the molten metal charge.

The proper installation and care of the refractory lining is of particular importance in the operation of any induction furnace. However, it is hardly possible to avoid the formation of minute cracks in a large refractory lining. An interesting feature of the use of line frequency is the fact that relatively small amounts of power are induced in the metal that may seep into such cracks. One may count on such fins freezing safely before reaching the water-cooled induction coil. By contrast, a high-frequency current will heat small particles in the vicinity of the coil much more effectively.

Typical Installations

Melting Aluminum

Figures 4 and 5 show a coreless line frequency induction melting furnace for aluminum, with a capacity of 1800 lb, installed in the aluminum foundry of a leading German automobile manufacturer. The furnace is rated 240 kva, at 50 cycles; it is connected to a three-phase supply at 380 volts. It has a melting rate of 880 lb per hour and a power consumption of 460 kwh per metric ton (4 lb per kwh).

The furnace is charged cold with aluminum ingots or scrap with a minimum cross section of 4 in. x 2½ in. If no alloy changes are contemplated for a period of time, a molten metal heel of about 200 lb is left in the furnace between charges.

Metal losses are remarkably low. When melting solid aluminum scrap, losses range between 0.3 and 0.8 per cent. Water-free chips, often heavily con-

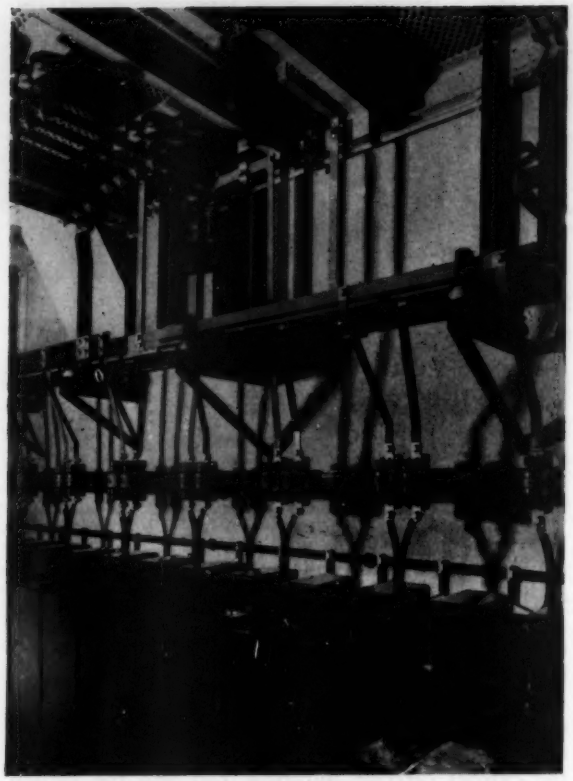


Fig. 6 — Capacitor bank for power factor compensation of the aluminum melting furnace.

taminated with oil, will show metal losses between 3.3 and 8 per cent. The operation requires no fluxing. As a comparison, metal losses in gas or fuel fired rotary furnaces are of the order of 2 per cent for solid scrap, and a multiple of the above figure for chips.

The refractory material used in this case is of a special semi-acid composition containing about 30 per cent alumina and is mixed with 4 to 5 per cent water. The average lining life is about 4000 charges. There has never been a case of a molten metal break-through from these aluminum melting furnaces.

Figure 6 shows the capacitors used in this installation; Figure 7 shows the electrical control unit and the regulating transformer.

The coreless induction furnace has been used with equal success in the melting of wrought alloys which are generally more delicate than casting alloys. Experience has shown that the slight bath movement obtained when operating on lowest power is beneficial in elimination of slag inclusions; metal from the coreless line frequency melting furnace has been of particularly good quality and machinability.

Melting Magnesium

A special application for the line frequency furnace is the melting of magnesium contained in a steel crucible. Such a unit is shown in Figs. 8 and 9. The cross section, Fig. 8, illustrates a thin-walled steel crucible which may be lifted out of the coil for direct transfer to the casting floor, thus eliminating the use and heating of ladles. This furnace has a power rating of 90 kw and a capacity of 330 lb magnesium. Its melting rate is 330 lb per hour; average power consumption is of the order of 600 kwh per metric ton (3.7 lb per kwh), which includes the reheating of the removable crucible. The superheating temperature is about 1500 F.

The use of a thin crucible wall (about ½ in.) will cause 60 to 70 per cent of the heat to be induced directly in the melt, the remainder in the crucible wall. Thus, overheating of the steel wall is low, and crucible life is increased.

A special coreless line frequency induction melting furnace for magnesium uses a thick-walled crucible that absorbs about 75 per cent of the electrical energy. The main reason for this construction was the larger capacity of the unit which increased the strength requirements of the crucible. Also, for metallurgical reasons, a higher supeheating temperature coupled with the least possible bath movement, was required. Due to these factors, the furnace, rated 180 kw, has a power consumption of about 750 kwh per metric ton (3 lb per kwh), and an average melting rate of 460 lb per hour.

Coreless Furnaces for Cast Iron

The coreless line frequency induction furnace has been of increasing importance in cast iron foundries. Figure 10 shows the smallest commercial unit built at present, with a capacity of 1800 lb. This furnace is used in the production of cast iron valve stems for the automotive industry, which are cast with a hard ledeburitic layer, with very narrow tolerances as to depth of hardness. Accurate analysis of the melt and controlled casting temperature are of primary importance in this operation, and the coreless line frequency induction furnace has been found to be the ideal tool for the job. This 50-cycle, 250-kw unit is connected to a 380-volt three-phase power supply and has a melting rate of 570 lb per hour. When melting cast iron and superheating to 2640 F, the power consumption is of the order of 600 kwh per metric ton

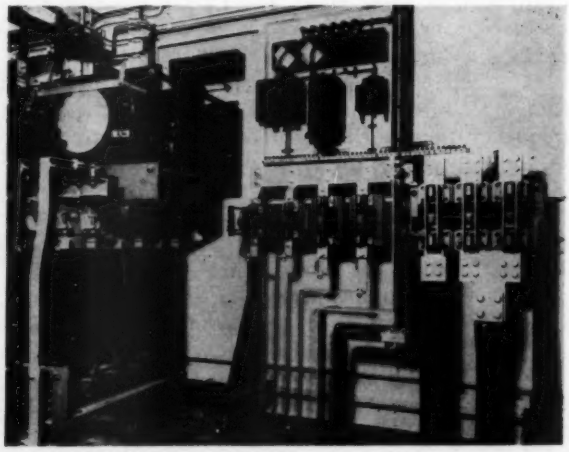


Fig. 7 — Regulating transformer for coreless line frequency aluminum melting furnace.

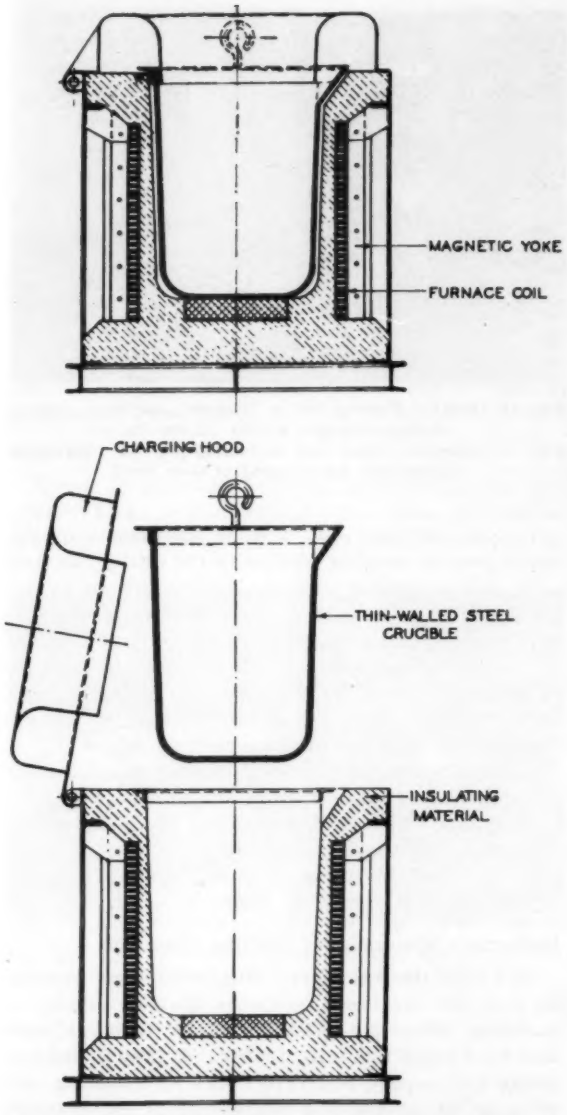


Fig. 8 — Cross section of coreless line frequency induction furnace with thin-walled steel crucible for melting magnesium alloys.



Fig. 9 — Coreless line frequency magnesium melting furnace with thin-walled crucible.



Fig. 10 — Line frequency induction furnace for cast iron; Includes capacitors and transformer with furnace and tilting frame in one compact unit.

(3.7 lb per kwh).

In this particular installation, the entire electrical equipment, including capacitors, was incorporated with the furnace and tilting frame to form one packaged unit.

Furnaces of similar design with larger capacities from 3300 to 4000 lb, suitable for cold charging, have been in operation in foundries of the automotive and heavy machinery industries. Figures 11 and 12 show a 1000 kw unit with a crucible capacity of 18,000 to 20,000 lb, installed by one of the leading German heavy machinery manufacturers, for the production of chilled cast iron rolls. This unit, operating at 50 cycles, is connected by a transformer to the three-phase line of 10,000 volts. The furnace is capable of melting 3300 lb per hour with an average power consumption of 550 kwh per metric ton (4 lb per kwh), casting at 2640 F. The installation of this furnace was so successful that an even larger unit of 44,000 lb capacity is now under consideration.

The intensive bath movement inherent in this furnace makes it particularly suitable for melting charges consisting entirely of chips. This may be done simply by charging loose chips into the furnace. By locating the furnace opening directly at floor level, chips are charged manually and will be melted as fast as they can be charged. Metal losses amount to a maximum of 2 per cent. The analysis of the charge does not change while melting, thus enabling the casting of high quality rolls directly from scrap.

Figures 13 and 14 show a 6600-lb line frequency furnace installed at a large metal-working plant in Berlin, Germany. The furnace is charged with sheet scrap from stamping operations, such as transformer sheet, to produce high quality synthetic cast iron. It is known that recarburizing is more difficult with higher silicon contents. However, the strong bath movement in the coreless furnace enables raising the

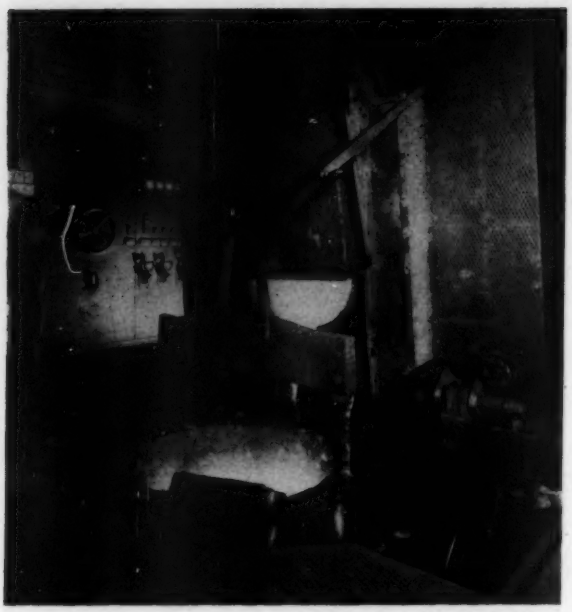




Fig. 11 (Left) — Pouring off a 1000-kw cast iron melting furnace, capacity 18,000 - 22,000 lb.
Fig. 12 (Above) — Same unit in melting position. Note easy accessibility for charging at floor level.

carbon content from 2.3 to 3.3 per cent C in a matter of 30 minutes, even with silicon contents up to 1.8 per cent. This corresponds to a recarburizing speed of 2.0 per cent C per hour. Recarburizing is performed continuously while melting. If the silicon content is lower, for instance 0.4 per cent, recarburizing is faster, and an increase in carbon content from 0.1 to 3.4 per cent C was obtained in 50 minutes, corresponding to a recarburizing speed of 3.96 per cent C per hour.

The coreless line frequency induction furnace has further found a field of application as a superheating unit in cast iron foundries, operating with liquid metal charges. Here, this furnace type has proven to be of advantage because of its reliable operation even at temperatures above 2640 F. When superheating a charge of molten cast iron, power consumption is of the order of 50 kwh per metric ton per 100 C temperature increase.

Production of Spheroidal Graphite Cast Iron

The most important factor in the production of spheroidal graphite cast iron is attainment of the lowest possible sulphur content of the initial cast iron melt. When working with the coreless line frequency furnace, the initial charge may consist of low-cost steel turnings or chips which may be gradually recarburized during melting. Thus, a cast iron is obtained which has a sulphur content not higher than steel. This low sulphur content results in saving considerable amounts of the expensive Ni-Mg alloy.

Latest developments have shown the possibility of desulphurizing cast iron by means of a fluoride containing lime slag in the basic-lined coreless furnace. Extremely low sulphur contents have been obtained after a period of about 60 minutes, which is still within practical operating limits.

Refractory Materials for Melting Cast Iron

As a rule, the refractory linings of coreless furnaces for cast iron are based on either siliceous fireclay or quartzite. Siliceous fireclay is rammed with the addition of a small amount of water. A 6600-lb furnace lining will require about 10 hours for ramming and 48 hours for drying and pre-heating of the crucible. Siliceous fireclay is an efficient lining for temperatures up to 2640 F. For higher temperatures, a quartzite lining has been successful. This material is

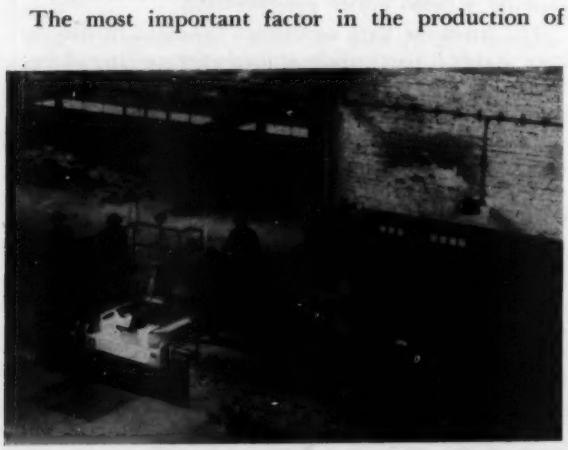


Fig. 13 — Coreless line frequency induction furnace for synthetic cast iron.



Fig. 14 — Skimming the furnace in Fig. 13.

rammed in the dry state, a procedure similar to that used for high-frequency furnaces. A simple ramming mold about 0.2 in. thick is left in the furnace to be melted out with the sintering charge. The quartzite used is of very fine (0 to 0.08 in.) grain size and contains about 0.6 to 1.8 per cent boric acid (H_3BO_3).

Depending on the nature of the melting operation, from 7 to 17 lb of siliceous fireclay per ton of cast iron are required; in the case of quartzite, from 4½ to 13 lb per ton of cast iron are used.

The advantages of the quartzite ramming mixture are its safe use at temperatures up to 3000 F, and the possibility to start the furnace immediately after ramming, since the material does not require any drying. On the other hand, quartzite costs about twice as much as siliceous fireclay.

Melting of Steel

Figure 15 shows an example of a coreless line frequency furnace for melting steel. This installation produces austenitic steels of very low-carbon content,



Fig. 15 — Two-furnace installation operating alternately from one 500-kw transformer, for melting steel.

usually about 0.04 per cent C. As a rule, the operation consists of melting steel scrap of normal carbon content (about 0.20 to 0.40 per cent C), decarburizing this metal to 0.03 per cent C, and then alloying the melt. The power consumption in this installation is 700 kwh per metric ton (3.1 lb per kwh). This particular plant is of interest because it operates two 3300-lb furnaces alternately through a transfer switch from a common control unit rated 500 kw.

For the melting of steel two lining materials have been used, quartzite and fused magnesite.

If kept in good repair, a quartzite lining will have a life of about 120 charges of steel. The molten steel will attack the crucible more severely, and a consumption of quartzite will be 24 to 33 lb per metric ton of steel. This comparatively high figure is due to the fact that a quartzite lining must be patched and repaired before complete relining is necessary.

Magnesite will give a shorter lining life, of about 80 charges. This is due to its greater sensitivity to temperature shock. A magnesite consumption of about 7 to 13 lb per metric ton of steel is to be expected. Magnesite is by far the more expensive material and, in Germany, will cost about five times as much as quartzite.

TABLE 1 — CORELESS LINE FREQUENCY INDUCTION FURNACES IN EUROPEAN FOUNDRIES FOR FERROUS AND NON-FERROUS METALS

Metal	Number of Furnaces	Capacity, lb	Kilowatt Rating, kw	Melting Rate lb/hr	Power Consumption kwh/ton*	Power Consumption lb/kwh
Cast Iron and Steel	1	1800	120	440	570	3.5
	3	1800	250	900	540	3.7
	10	1800	300	1100	540	3.7
	1	3500	250	900	550	3.6
	2	3500	300	1050	540	3.7
	9	3500	320	1100	550	3.6
	9	3500	450	1670	530	3.7
	1	4500	250	900	560	3.5
	1	7000	300	1100	540	3.7
	2	7000	320	1140	550	3.6
	4	7000	450	1700	520	3.8
	3	7000	600	2050	510	3.9
	7	7000	720	2900	500	4.0
	1	11000	500	1950	500	3.9
	3	13000	300	1000	590	3.4
	1	13000	1100	4400	500	4.0
	1	18000	1000	3750	490	4.1
	59		24,980 kw			
Aluminum and its Alloys	1	650	75	275	500	4.0
	3	1300	150	570	500	4.0
	3	1650	170	620	490	4.1
	1	1750	240	820	500	3.9
	5	2200	250	930	490	4.1
	2	3800	480	1750	480	4.2
	1	4400	360	1200	510	3.9
	3	4400	450	1750	490	4.1
	3	4400	500	1950	490	4.1
	1	6600	600	2150	470	4.2
	23		7,295 kw			

*Short ton of 2000 lb

Figure 16 shows an interesting installation which is at present under construction. The 1800-lb induction furnace, rated 300 kw, will be used to melt tool steel which will be cast directly into ingot molds, also shown in the cross section. This installation will be completed at the plant of a leading German electric steel manufacturer who, some 30 years ago, pioneered the use of high-frequency induction furnaces.

Table 1, finally, surveys the number of coreless line frequency induction furnaces in operation as of August 15, 1955, in the ferrous and non-ferrous industries of European countries.

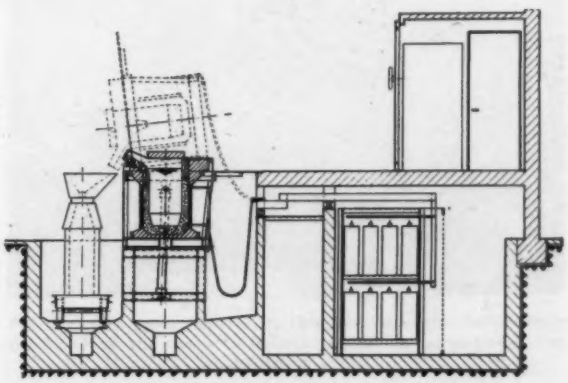


Fig. 16 — Coreless line frequency induction furnace for melting tool steel.

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DISCUSSION

Chairman: JAMES THOMPSON, Continental Foundry & Machine Div., Blaw-Knox Co., East Chicago, Ind.

Co-Chairman: H. C. WEIMER, Beardsley-Piper Div., Pettibone-Mulliken Corp., Chicago.

W. E. LEWIS[1] *(Written Discussion):* It is interesting to note practically 60 of the small line frequency furnaces are used on ferrous work. Of these the 1800-lb, the 3500-lb, and the 7000-lb sizes seem to be the more popular.

As to the 7000-lb size one sees that the highest powered units in this class are backed with 720 KW and these can melt 3½-ton heats in about 2½ hours.

Arc furnaces of 7000-lb charge size are generally backed with 1700 KW or higher. One furnace of this charge size—averaging 7800-lb heats—melts those heats in 65 minutes tap to tap. This means a rate of 120 lb of acid foundry steel per minute. The power consumption, using oxygen, averages under 450 KWH per ton. Roof life 450 heats; sidewall life 350 heats.

The charge in this case is foundry sprues, having some adherent sand, and purchased steel scrap. This scrap normally is rusty enough that a melting loss of say 4 per cent applies. Being top charged by bucket, the furnace can use fairly good sized pieces of scrap in as much as the inside lining diameter of the furnace unit is about 6 feet.

We shall not comment particularly on the use of the coreless line frequency induction furnace on non-ferrous work as we confine our activities to iron and steel melting and refining furnaces.

We note the relatively low power factor of the electric crucible type of melting furnace described by Mr. Junker. While some arc furnaces use capacitors for power factor boosting this ordinarily is not the case. The direct-arc furnace generally melts down at about 85 per cent power factor or higher. After a charge has been melted to fluid, the power factor ordinarily exceeds 90 per cent.

In reading about the 1800-lb furnace on cast iron valve stems it seemed that the 2640 F tapping temperature given is rather to the low side. A small arc furnace making alloy iron heats for this type of work has been found on checking to be tapping at 2850 F.

Concerning single phase operations we ordinarily find power companies rather unsympathetic to single phase in any capacities over 300 to 400 KW in size.

Referring to the 9- to 10-ton furnace powered with 1000 KW the writer would say that this seems low powered. A 9- to 10-ton usual charge size arc-furnace would normally be backed with at least 4000 KVA. The 5-ton usual charge size arc-furnace has been run with as high as 5500 KVA power imput.

Metal losses at 2 per cent indicate an extremely clean charge. A direct arc furnace user in the United States continuously cold melting cast iron borings at a 3-ton per hour rate reports average melting loss of 6 per cent. Tapping temperature, 2750 F.

One thought in connection with the coreless line frequency induction furnace is that this construction scarcely could favor the use of the oxygen lance to speed the operation. Apparently, too, these furnaces need a construction where the depth of bath is relatively deep. Hence, the operation is one of melting a select charge rather than one of making metal from a considerably less select charge.

Observation of the furnace layouts shown indicates the coreless line frequency induction furnace could deliver quite nicely into ladles operating either on a car or handled by crane. This arrangement minimizes head room over the furnaces themselves.

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SUB-SURFACE BLOWHOLES IN GRAY IRONS AND THEIR ASSOCIATION WITH MANGANESE SULPHIDE SEGREGATION

By

W. G. Tonks*

Some casting defects may be apparent immediately on cleaning the rough casting, but, on the other hand, defects may not be revealed until during or after the machining operation. Sub-surface defects are not only costly, due to wasted process time, but if there is a marked time-lag between casting and the machining operation, foundry production proceeds unaware that waste is occurring. Trouble due to sub-surface holes is not necessarily revealed only by machining, as stringent shotblasting may remove the surface skin to show up the defects. The presence of sub-surface holes may cause defects in castings which, although not machined, undergo a subsequent finishing process, such as vitreous enamelling, chromium plating or galvanizing.

Following an original observation by F. J. Cook,[1] R. Jolly[2] has shown that as the manganese content approaches the silicon content, there is a greater tendency to form a slag-like inclusion and a segregation of manganese sulphide. It was suggested that the manganese sulphide concentration was a secondary effect caused either by mechanical trapping or due to a more complex slag metal reaction.

A. D. Morgan[3] has discussed the occurrence of gas holes and has established that the rate of formation of ladle slag was higher during periods when gas holes occurred. An increase in the sulphur content of the metal resulted in a higher MnO content in the ladle slag. When the manganese content of the metal rose above 0.9 per cent there was an increase in the incidence of gas holes.

The first part of this paper shows some typical examples of the occurrence of sub-surface blowhole defects. The second part details experimental work designed to determine and investigate the cause of the trouble.

The term "blowhole" is often used in its widest sense to describe a cavity with a smooth surface,

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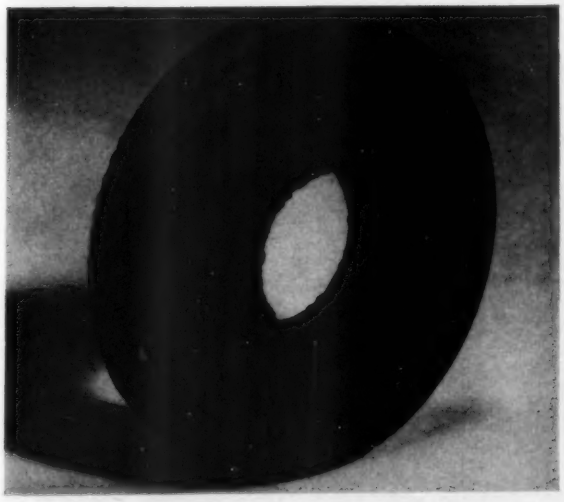


Fig. 1 — Gross blowhole resulting from excessive moisture content and low permeability molding sand; overall dimensions: 6-in. diameter, ¾ in. thick.



Fig. 2 — Gas holes shown in this casting were deliberately produced by the use of a ladle in which the sand lining was inadequately dried.



Fig. 3 (Left) — Casting A — Gland; overall dimensions: 5-in. diameter, ¾ in. thick; Fig. 3A (Right) — Microstructure at 100X, unetched.

which is generally taken as an indication that the void was formed and retained by a gas bubble. In order to define the type of blowhole under discussion in this paper, blowholes may be regarded as falling into the following categories:

1. Gross blowholes.
2. Gas holes.
3. Small blowholes, usually sub-surface. These are the type dealt with in this paper.

Gross Blowholes

In the example of this type of blowhole shown in Fig. 1, the defect was not revealed until after machining had begun. This defect is sometimes attributable to the use of a low permeability molding sand which retards the escape of mold gases, and in this connection it is often forgotten that increased mold hardness markedly reduces permeability. High gas evolution may be due to a high moisture content of the molding sand, or to corebinders with a high gas content. Under-baking will also give rise to cores with a high gas content. Vents in molds and cores which become blocked by metal penetration are a potential cause of gross blowhole trouble. Further sources of this type of blowhole may be unsuitable chaplet coatings, and rusty chills and metal inserts.

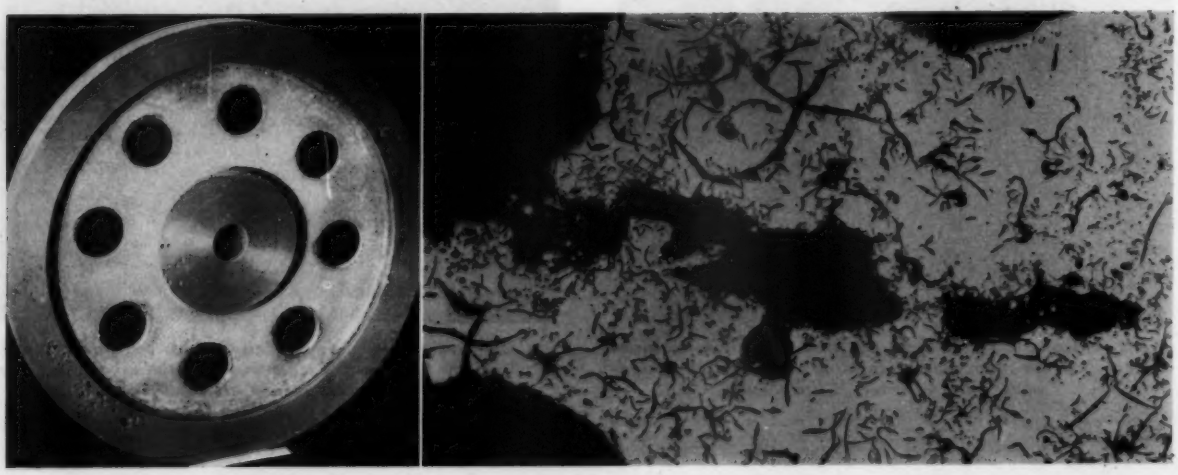
Gas Holes

Gas holes in cast iron are generally attributable to hydrogen, and may be recognized as smooth, elongated holes with bright, shiny surfaces. The holes may be evenly distributed throughout the casting section, but they often tend to migrate to the top surface of the mold.

Sources of hydrogen pick-up may be found in a damp cupola bottom or launder, or in the use of damp ladle inoculants, or the moisture from the mold. The defective casting shown in Fig. 2 was deliberately produced by the use of a ladle in which the lining was inadequately dried.

The foregoing types of blowholes are not dealt with in this report.

Fig. 4 (Left) — Casting B — Pulley Wheel; overall dimensions: 2 ft, 0 in. diameter, rim 2 in. thick; Fig. 4A (Right) — Microstructure at 100X, unetched.



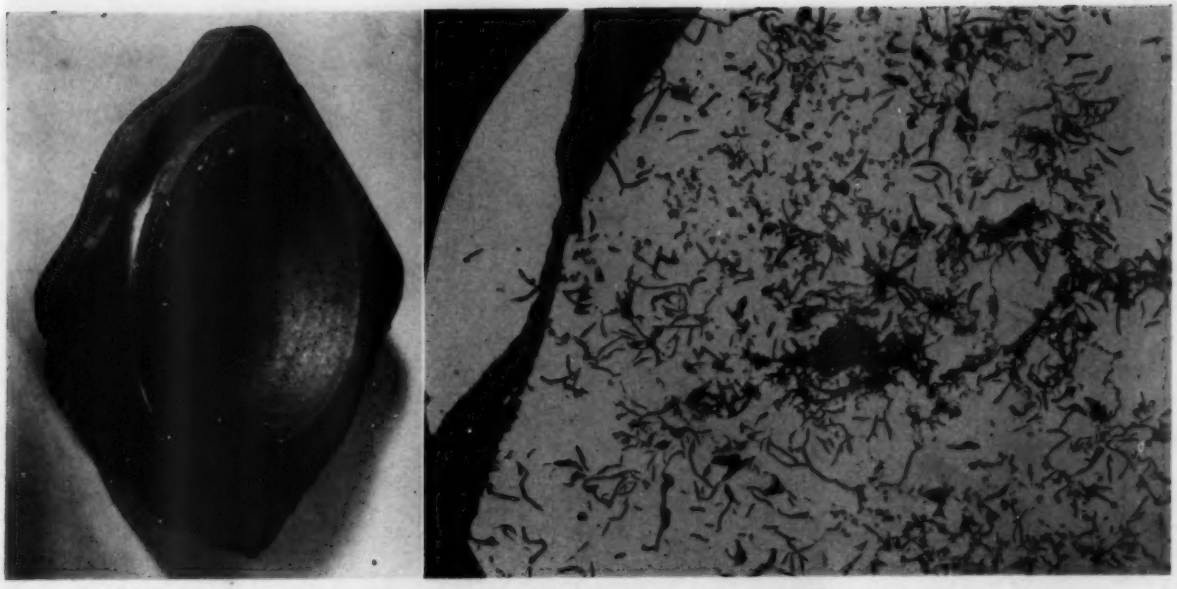


Fig. 5 (Left) — Casting C — Cover Plate; overall dimensions: 5 in. square, ½ in. thick; Fig. 5A (Right) — Microstructure at 100X, unetched.

Small Blowholes

Blowholes of this type are usually spherical, with smooth, though generally dull, surfaces. They often contain a slag-like material and are generally not fully revealed until the surface metal has been removed by machining. They are usually to be found in clusters beneath the top surface of a casting, as poured. These blowholes may make sporadic appearances in the foundry, to disappear without apparent change in standard conditions.

The examples of small sub-surface blowholes illustrated in Figs. 3 to 9 were collected from seven different foundries, and are in every case typical of the defect under discussion, which also occurred in other types of casting produced. The section size of the castings concerned ranged from ⅝ in. to 2½ in., and with the exception of the cylinder casting, which was produced by a core assembly, the castings were poured in green-sand molds. All the castings were machined on the face in which the defects were present.

As shown in Table 1, this type of defect can occur over a wide range of metal composition. With the exception of Casting F, it will be seen that the manga-

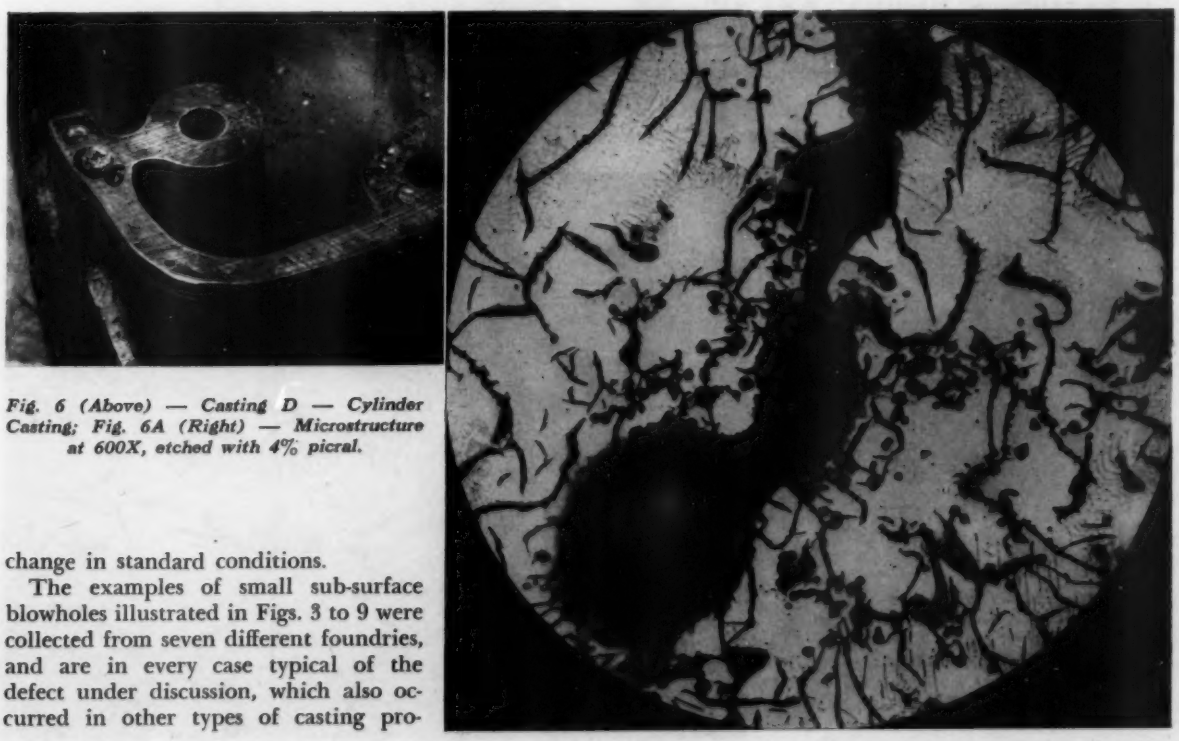
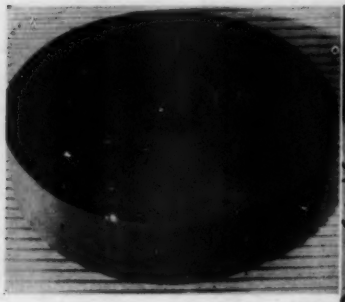


Fig. 6 (Above) — Casting D — Cylinder Casting; Fig. 6A (Right) — Microstructure at 600X, etched with 4% picral.



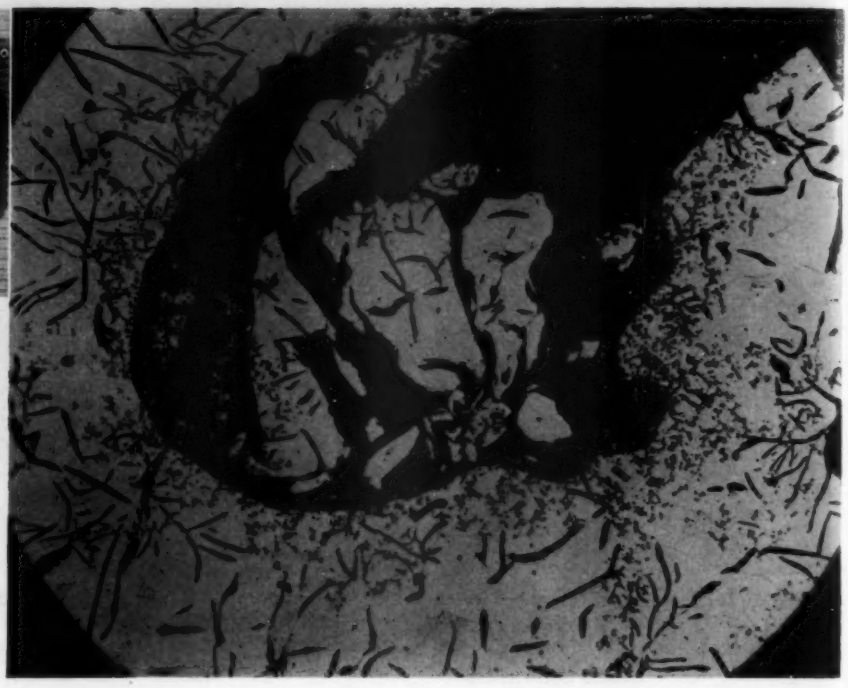


Fig. 7 (Above) — Casting E —Gear Blank; overall dimensions: 8-in. diameter, 2½ in. thick; Fig. 7A (Right) — Microstructure at 100X, unetched.

TABLE 1

	Casting	T.C., %	Si, %	Mn, %	S, %	P, %
A	Gland*	3.58	1.82	0.77	0.064	0.56
B	Pully Wheel	3.35	2.68	0.65	0.069	0.58
C	Cover Plate	3.14	2.19	0.63	0.109	0.81
D	Cylinder Casting	3.25	2.82	0.74	0.087	0.33
E	Gear Blank	3.38	2.74	0.84	0.145	0.84
F	Bracket	3.17	2.31	0.49	0.149	0.56
G	Bearing Housing	3.17	1.79	0.58	0.121	1.01

*This casting was known to have been poured extremely cold.

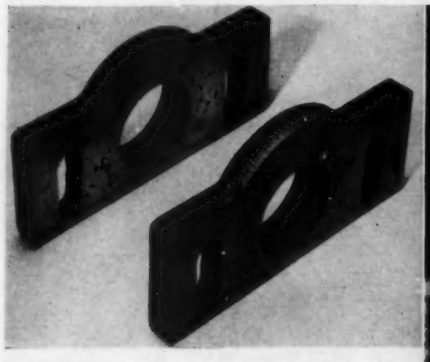
nese content is in excess of the amount necessary to balance the sulphur according to the well known formula:

$$Mn\ \% = (1.7 \times S\ \%) + 0.3\%^{4}$$

and the sulphur content appears to be representative of the whole range of variation likely to occur with cupola-melted gray iron.

On examining photomicrographs from castings showing this particular type of defect (Figs. 3A to 9A), taken in most cases at right angles to the blowholes, it will be seen that there are two common features:

(a) the appearance of a slag-like material in the defective area; and
(b) a segregation of manganese sulphide within the defective area.



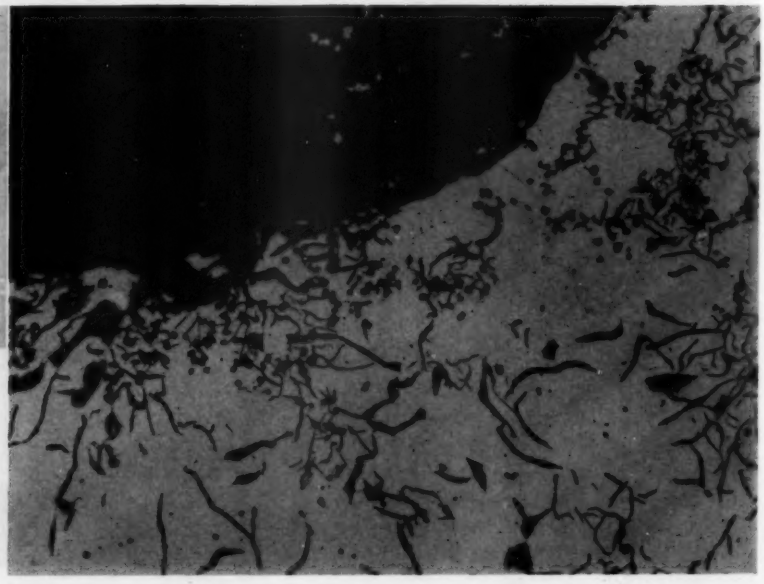


Fig. 8 (Above) — Casting F — Bracket; overall dimensions: 6 in. x 3½ in. thick; Fig. 8A (Right) — Microstructure at 300X, unetched.



Fig. 9 (Above) — Casting G — Bearing Housing; Casting sectioned to reveal defect; Fig. 9A (Right) — Microstructure at 100X, unetched.

In some instances the manganese sulphide segregation extends into the slag-like material.

As the common feature of these defects was manganese sulphide segregation, experiments were carried out to study the influence of variations of manganese and sulphur contents, together with the effect of pouring temperature on the incidence of sub-surface blowholes.

Experimental Work

The test casting and gating arrangement used are shown in Fig. 10. The dimensions of the test casting were chosen to minimize the weight of metal required per melt, and to facilitate the subsequent machining operation. An oil-sand core was used to form the center cod in order to standardize mold conditions beneath the thick section of the test casting.

The metal was melted in an oil-fired crucible furnace, using a plumbago melting pot. The maximum temperature attained in the furnace did not exceed 1400 C (2550 F). Plumbago ladles and conventional skimming practice were employed when pouring the test castings.

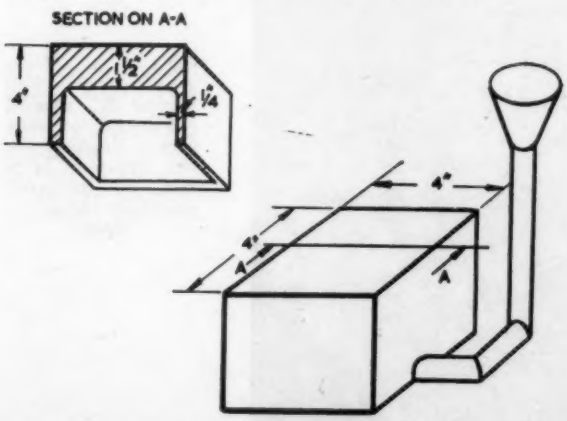


Fig. 10 — Sketch of the test casting gating system. The center of the casting is blocked out by on oil sand core.

Each melt was carried out at a different sulphur content, ranging from 0.009 per cent to 0.150 per cent. The manganese content was progressively raised in five casts from each melt. From each cast three test castings were poured at three different temperatures. The full details of the range of composition employed in the test castings is shown in Table 2. A separate sample for analysis was taken from the ladle at approximately 1350 C (2460 F), immediately prior to pouring the first of each manganese series of test castings in each particular melt. There was some variation in the initial manganese content in each melt, and this was due to the availability of material to give the sulphur content required.

The test castings were poured in green-sand molds using naturally-bonded Bromsgrove red as a base sand. This sand was used as a unit sand, and the average physical properties were: Moisture, 5 per cent; Green Strength, 9 psi; Dry Strength, 130 psi; Permeability No. 27; and Shatter Index, 70.

TABLE 2 — COMPOSITION OF TEST CASTINGS

T.C., %	Si, %	P, %	S, %	Mn, %				
3.26	2.20	1.06	0.150	0.43	0.58	0.80	0.94	1.00
3.12	2.19	1.14	0.145	0.41	0.63	0.79	0.98	1.17
3.10	2.25	1.13	0.140	0.46	0.62	0.83	0.98	1.13
3.01	2.05	1.04	0.130	0.25	0.45	0.68	0.91	1.12
3.04	2.30	1.14	0.110	0.54	0.70	0.93	1.08	1.22
3.10	2.58	0.91	0.091	0.35	0.55	0.76	1.03	1.24
3.07	2.36	0.80	0.088	0.48	0.70	0.95	1.13	1.33
2.73	2.15	1.04	0.065	0.29	0.46	0.64	0.82	0.97
3.09	2.44	0.83	0.058	0.30	0.60	0.68	0.90	1.04
2.80	2.40	1.11	0.057	0.25	0.43	0.60	0.81	0.97
3.15	1.85	0.60	0.009	0.14	0.28	0.54	0.73	0.84

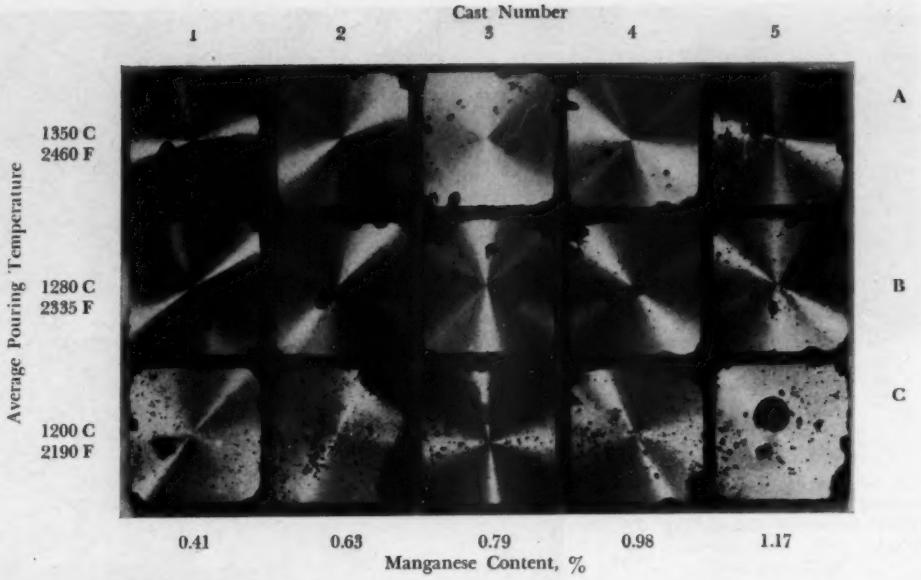
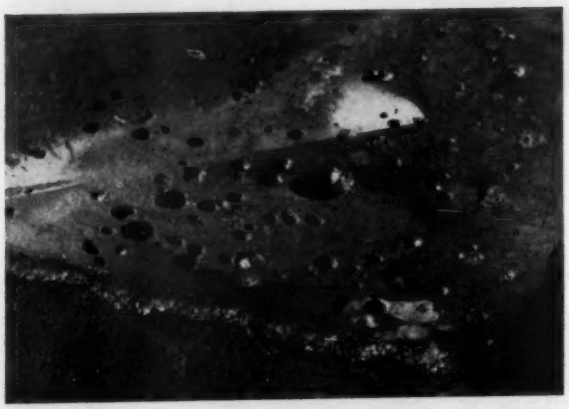


Fig. 11 (Right) — Specimens from Melt A of following composition:

T.C. — 3.12 %
Si — 2.19 %
S — 0.145%
P — 1.14 %

Fig. 11A (Below) — Close up of test casting 4C.



The cores were produced from a Congleton silica sand, using 1 per cent linseed oil and 2 per cent starch as binders. The cores were baked for 1½ hours at 450 F. The average physical properties of the core sand were:

Green Strength, psi — 0.8
Tensile Strength, psi — 240
Dry Permeability No. — 100
Gas Content at N.T.P. — 14 cc/g.

After shotblasting, the top surface of each casting was machined either by planing or turning, the minimum cut being taken which would reveal the blowhole defects, usually about 0.040 in. to 0.100 in.

Of the melts carried out, four representative melts

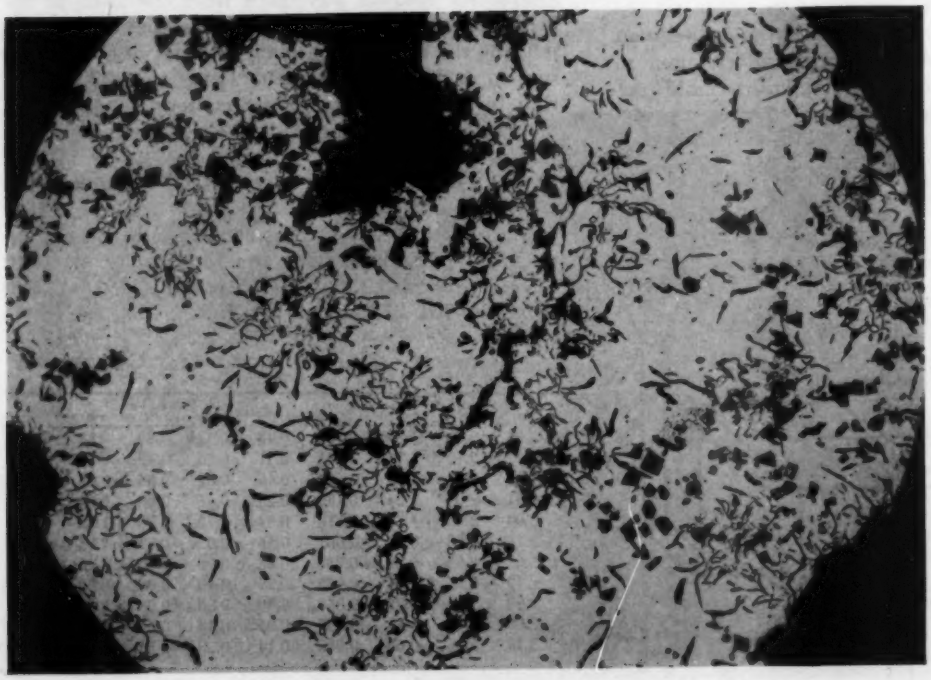


Fig. 12 — Micrograph from Test Casting 5C, showing segregation of manganese sulphide and the inclusion of a slag-like material; unetched; Mag. — 100X.

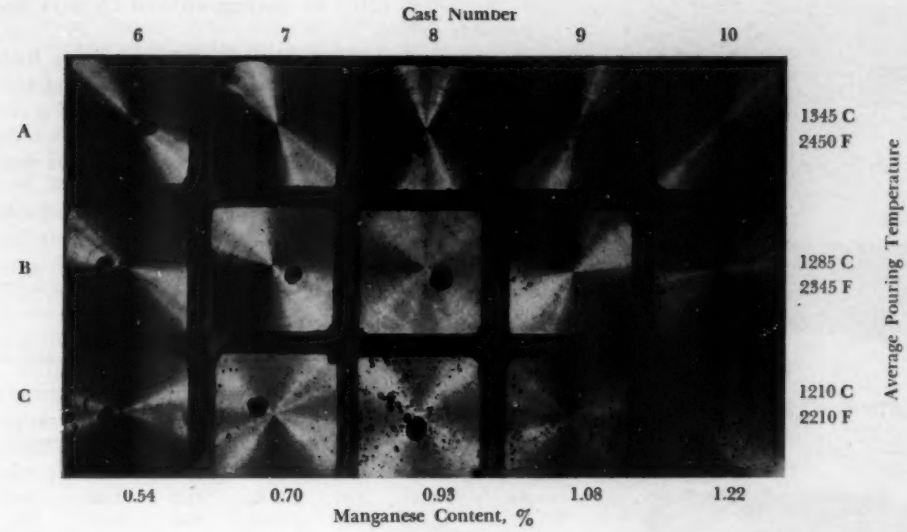


Fig. 13 (Left) — Specimens from Melt B of following composition:

T.C. — 3.04
Si — 2.30
S — 0.11
P — 1.14

Fig. 13A (Below) — Close up of test casting 9C.



are recorded in detail. Data obtained from the other melts were used as additional information to compile the records of sound and defective castings.

Melt A

The larger holes shown on the surface of each test casting were due to shrinkage cavities (Fig. 11). They are clearly distinguishable from the blowholes and they occurred to some extent in all test castings. The photographs clearly show that the incidence of blowholes was increased as the pouring temperature was lowered and the manganese content raised. Figure 11A shows a close-up of the type of blowhole obtained in test casting 4C. All the test castings poured at 1200 C (2190 F) showed the blowhole defect; at 1280 C (2330 F) casting 1B was free from blowholes,

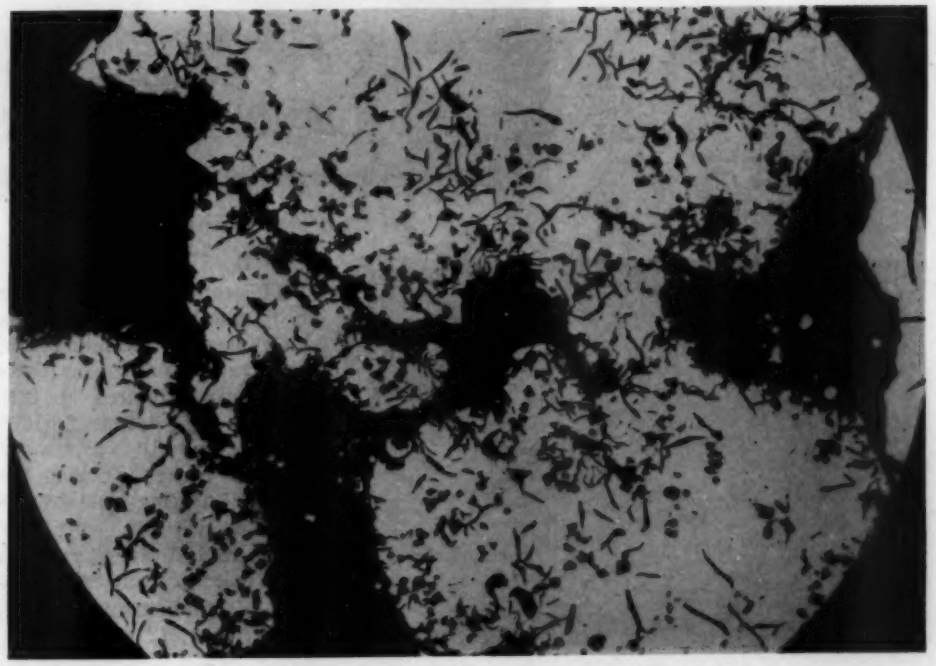


Fig. 14 — Micrograph from Test Casting 10C, showing segregation of manganese sulphide and the inclusion of a slag-like material; unetched; Mag. — 100X.

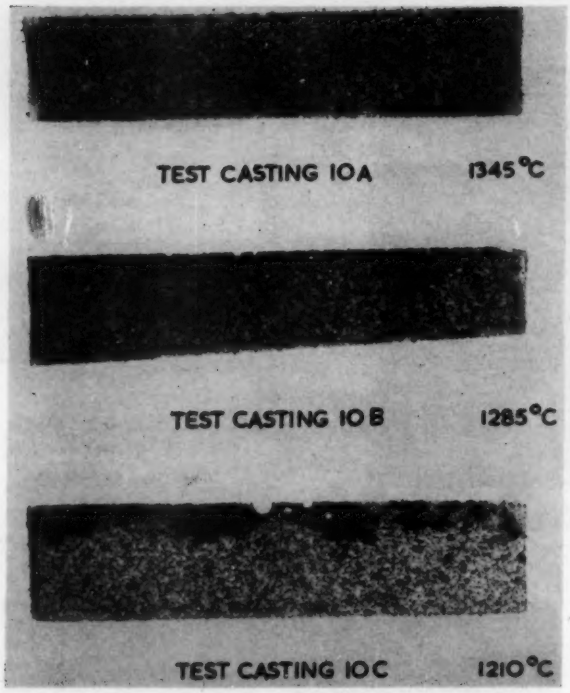


Fig. 15 — Sulphur prints.

and at 1350 C (2460 F) castings 1A and 2A were free from this defect.

A typical photomicrograph, in this case taken from test casting 5C (poured at 1200 C, 2190 F, and containing 1.17 per cent manganese) is shown in Fig. 12. There is a marked segregation of manganese sulphide, and the inclusion of a slag-like material at the base of one of the surface cavities may be seen.

Slag samples taken from the appropriate ladles by means of a copper skimmer, prior to pouring the castings, were analyzed and gave the following results:

TABLE 3

	Metal			Ladle Slag		
Test Casting	S, %	Mn, %	Ladle Temp., C	Mn, %	S, %	Si, %
4B	0.145	0.98	1280	10.0	0.085	40.9
4C	0.145	0.98	1200	27.0	0.64	40.9
5B	0.145	1.17	1280	12.3	0.163	36.9
5C	0.145	1.17	1200	25.5	2.20	32.5

The figures indicate that as the metal is cooling in the ladle, the slag collecting on the surface becomes progressively richer in manganese and sulphur as the metal temperature falls. It should be borne in mind that the sulphur content of the metal for each melt was taken from a separate sample cast at 2460 F.

Fig. 16 (Right) — Specimens from Melt C of following composition:

T.C. — 3.07 %
Si — 2.36 %
S — 0.088%
P — 0.80 %

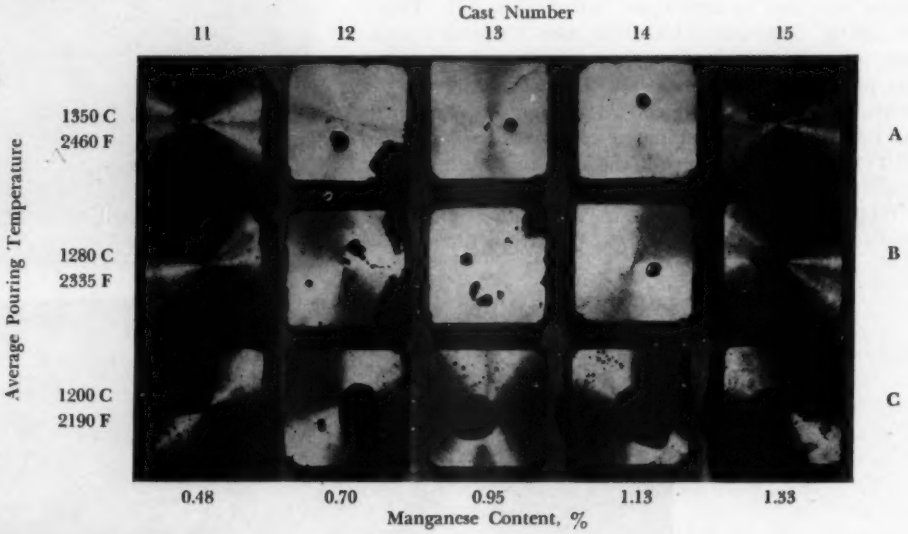


Fig. 16A (Below) — Close up of test casting 15C.



Melt B

The trend of blowhole defects illustrated in Fig. 13 is similar to that observed in Melt A—that is, blowholes are promoted by lowering the pouring temperature and raising the manganese content. All the castings poured at 1210 C (2210 F) exhibit the blowhole defect. At 1285 C, (2345 F), castings 6B, 7B and 8B are free from the trouble, while at the higher pouring temperature of 1345 C (2455 F) castings 6A, 7A, 8A and 9A are also free from defect.

Figure 13A shows a close-up of the blowhole defects in test casting 9C.

A photomicrograph taken from test casting 10C (poured at 1210 C, 2210 F, manganese content



Fig. 17 — Micrograph from Test Casting 15C, showing segregation of manganese sulphide and the inclusion of a slag-like material; unetched; Mag. — 100X.

1.22 per cent) is shown in Fig. 14, and again there is a segregation of manganese sulphide and the inclusion of a slag-like material. It will be observed that in this photomicrograph the presence of manganese sulphide extends into the slag-like material.

Sulphur prints taken from test castings 10A, 10B and 10C are reproduced in Fig. 15. This illustrates the increasing extent of segregation as pouring temperature is decreased.

Melt C

In the castings produced in Melt C, the blowhole trouble was generally confined to those castings poured at the lower temperature of 1200 C, (2190 F) as shown in Fig. 16. There was a surface blemish in casting 15B, and this was shown to be associated with a manganese sulphide segregation. The surface blemishes in test castings 11B to 14B are not associated with a segregation of manganese sulphide.

Figure 16A shows a close-up of the blowholes in test casting 15C.

The photomicrograph in Fig. 17 again illustrates manganese sulphide segregation and the inclusion of a slag-like material. However, it would appear that the segregation of manganese sulphide is not so pronounced as in the higher sulphur melts, A and B.

Melt D

Castings produced in test melt D show a somewhat different form of shrinkage cavity from that

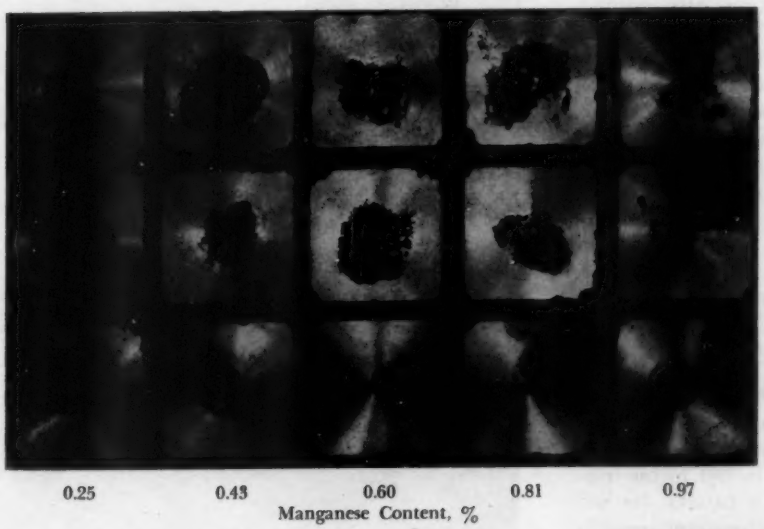


Fig. 18 (Left) — Specimens from Melt D of following composition:

T.C. — 2.8 %
Si — 2.4 %
S — 0.057%
P — 1.11 %

Fig. 18A (Below) — close up of test casting 20C.



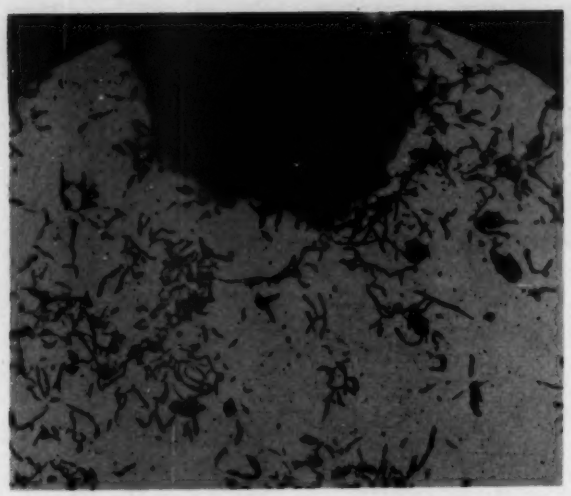


Fig. 19 — Micrograph from Test Casting 19C. There is no evidence of manganese sulphide segregation in the defective area. Unetched. Mag. — 100X.

observed in the test melts previously recorded. This shrinkage cavity is generally in the form of a surface depression. In addition to this surface shrinkage, as shown by Fig. 18, all the castings poured at the lowest temperature of 1220 C (2225 F) showed a surface blemish. It was also observed that there was a surface blemish in test casting 20B, poured at 1280 C (2330 F). No small surface blemishes were observed in castings poured at the higher temperature of 1350 C (2460 F). These blemishes, as may be seen in Fig. 18A, do not possess the marked characteristics of the sub-surface blowholes observed in melts A, B and C.

A photomicrograph taken from test casting 19C is illustrated in Fig. 19. In this particular photomicrograph there is no indication of a segregation of manganese sulphide, and there is no inclusion of a slag-like material. Furthermore, sulphur prints which were taken on test castings 20A, 20B and 20C did not reveal sulphide segregation.

Formation of Slag on Metal Surface

In carrying out these test melts it was observed that variations occurred in the so-called "break" on the surface of the metal in the ladle. At metal temperatures below 1330 C (2425 F) a pattern formed on the surface of the metal. This pattern grew in size as the temperature dropped, until at below 1200 C (2190 F) the pattern was replaced by a continuous film. As the manganese content of a particular melt increased, the pattern on the surface of the ladle appeared to be more marked.

Figures 20 to 22 show the surface of the same ladle of metal at three different temperatures. In this case the composition of the metal in the ladle was as follows:

Total Carbon, %	— 3.10
Silicon, %	— 2.58
Manganese, %	— 1.24
Sulphur, %	— 0.09
Phosphorus, %	— 0.91

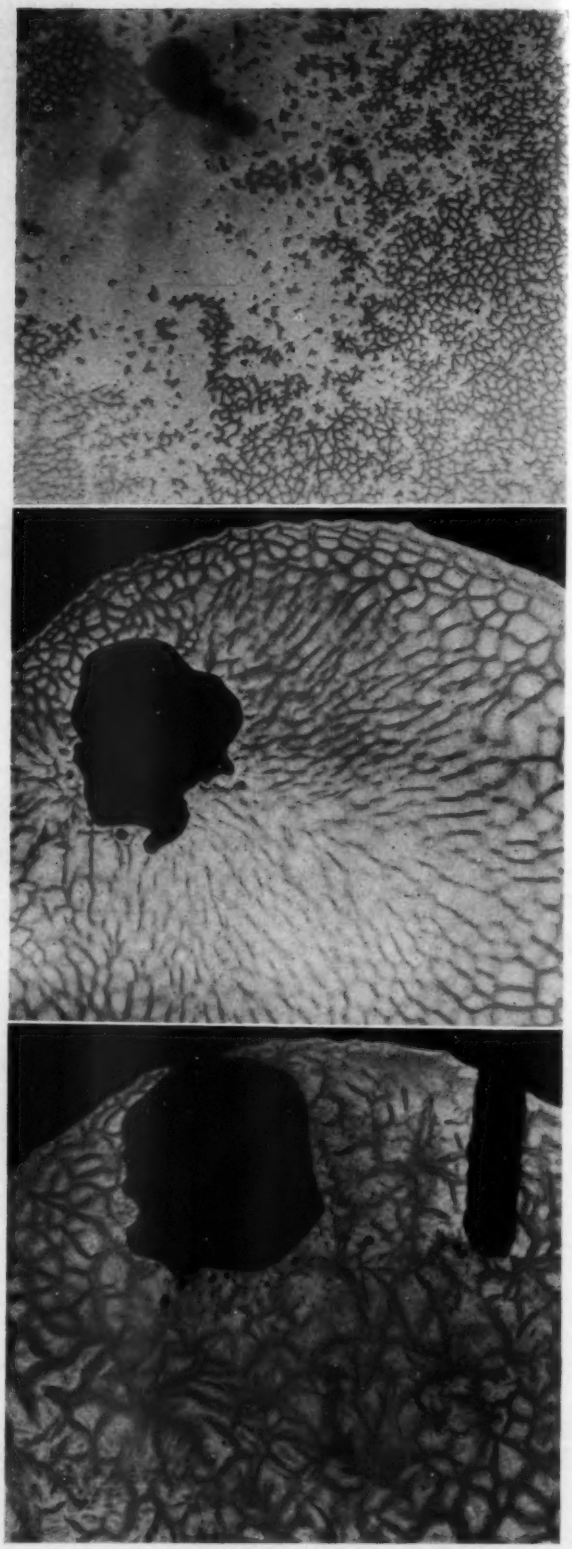


Fig. 20 (Top) — Formation of slag on metal surface; ladle temperature 1280 C (2335 F); Fig. 21 (Center) — Ladle temperature 1250 C (2280 F); Fig. 22 (Bottom) — Ladle temperature 1220 C (2230 F).

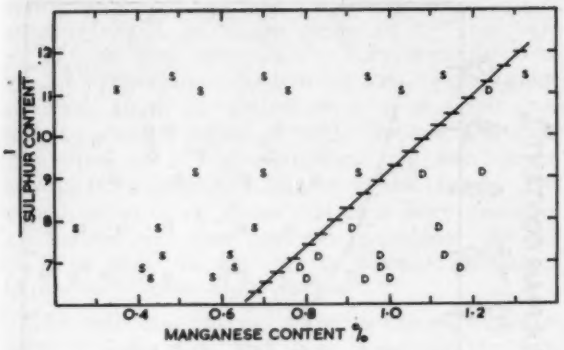


Fig. 23 — Curve shows relationship of sulphur and manganese content to blowholes in castings poured at 1280 C (2335 F).

It will be seen from Figs. 20 to 22 that the pattern becomes larger as the metal temperature falls.

It was observed that a small particle of slag appeared in the center of each polygon form, and then moved across the surface of the metal toward a central slag pool This movement is recorded in Fig. 21, and apart from the central pool, small particles of slag may be seen on the surface of the metal in Fig. 22.

Discussion

The metal composition used in the production of test castings (detailed in Table 2) was kept broadly within the range likely to be met in practice. The blowhole defects produced and shown in the details of Melts A, B and C were similar to those observed in the castings shown in Figs. 3 to 9. The common features of manganese sulphide segregation and the presence of a slag-like material also occurred in the test castings. As would be expected, in both practical examples and test castings the extent of the manganese sulphide segregation increased with increasing manganese and or sulphur content. It would appear that low pouring temperatures favored the occurrence of this type of blowhole, and at a given sulphur content blowholes more readily occurred as the manganese content was raised.

In the interpretation and explanation of the test results obtained, reference may be made to well-established concepts of the behavior of sulphur and manganese in cast iron.[4]

The reaction: $Mn + FeS \rightleftharpoons MnS + Fe$, moves to the right at lower temperatures and gives a precipitation of manganese sulphide. The tendency to go to the right at any one temperature and form manganese sulphide increases with increased manganese in the metal. Therefore, the temperature at which manganese sulphide precipitates from the liquid increases with an increasing manganese content.

By consideration of the above reaction, it has been shown[5] that the product:

(Mn %) x (S %) = an approximate constant K at any one temperature.

If the product is greater than this constant at any one temperature, then the separation of manganese sulphide should occur at this temperature. On the other hand, if the product is less than this constant, then manganese sulphide will not separate until some lower temperature is reached.

Joseph and Holbrook[6] have shown that, in fact, the product of manganese % x sulphur % at which manganese sulphide separates at any one temperature is an approximate constant, irrespective of the actual manganese and sulphur percentages.

The results of the experimental melts described earlier show that the sub-surface blowhole defects arise when the casting temperature is low enough to permit some segregation of manganese sulphide before and during pouring; that is, when the product of manganese % x sulphur % exceeds the constant value K. It would be expected, therefore, that the incidence of the defects would be related to this product at any one temperature.

In order to test the view that at a given pouring temperature some critical value of the product, (manganese content) x (sulphur content), must be exceeded before the segregation of manganese sulphide can occur, it is desirable to arrange the experimental results in a convenient manner. This may be done as follows:

At constant temperature let k = the critical value of the manganese and sulphur content of the metal at which blowholes occur:

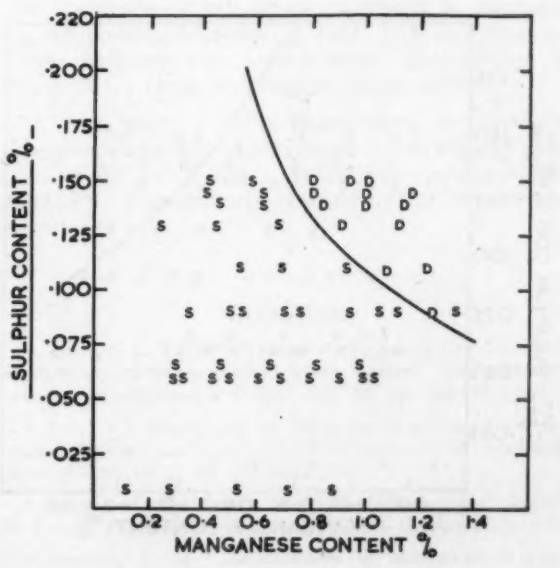


Fig. 24 — Curve shows relationship of sulphur and manganese content to blowholes in castings poured at 1280 C (2335 F).

$$\text{Therefore } (Mn) \times (S) = k$$

$$(Mn) = k \times \frac{1}{(S)}$$

It follows, therefore, that if the incidence of blowholes could be related to the product, manganese content x sulphur content, a plot of (Mn) against $\frac{1}{(S)}$ for each casting at constant pouring temperature would give a diagram in which the zones corresponding to sound and defective castings would be separated by a straight line.

In the graph shown in Fig. 23, the results from test melts A, B and C, together with the results from four other test melts which were carried out, are recorded at a constant pouring temperature of 1280 C (2335 F). Only the test castings in which the defects were associated with a segregation of manganese sulphide were plotted as defective. It is immediately apparent from the graph that sound and defective castings can be separated by a straight line. This indicates that at a constant pouring temperature of 1280 C (2335 F) there is a critical value for the product of the manganese content x sulphur content above which blowholes are likely to occur.

The results obtained from the test melts were plotted on the graph shown in Fig. 24. A value for the product of manganese content of metal x sulphur content of metal was determined from Fig. 23 and was used to draw the curve which, as shown in Fig. 24, segregates sound and defective castings. In a similar manner, graphs showing the segregation of sound and defective castings at constant pouring temperatures of 1350 C (2460 F) and 1200 C (2190 F) against variations of manganese and sulphur, are shown in Figs. 25 and 26.

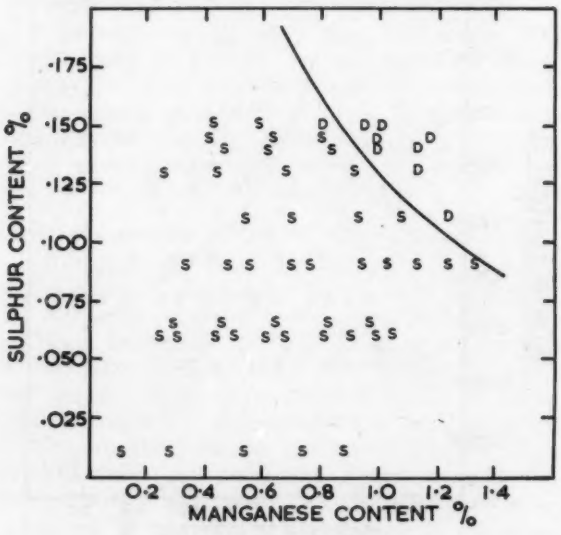


Fig. 25 — Curve shows relationship of sulphur and manganese content to blowholes in castings poured at 1350 C (2460 F).

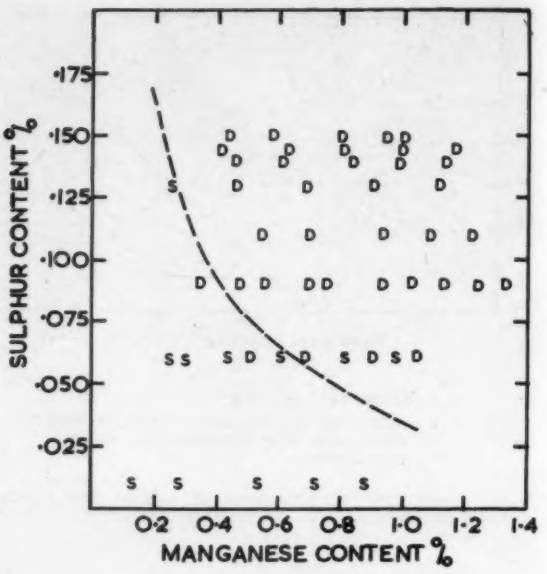


Fig. 26 — Curve shows relationship of sulphur and manganese content to blowholes in castings poured at 1200 C (2190 F).

It will be observed that by increasing the pouring temperature, the value of the product manganese content x sulphur content, above which blowholes are likely to occur, was raised. The influence of pouring temperature was, of course, indicated by observation of the grouped test castings shown in Figs. 11, 13 and 16. The curve separating sound and defective castings at the lower experimental pouring temperature of 1200 C (2190 F) (Fig. 26) has been shown as a broken line, since it was felt that insufficient data were available at such levels of manganese and sulphur content to define clearly the value of the product of manganese content x sulphur content, above which blowholes occurred. However, the visual evidence of Figs. 11, 13 and 16 shows that decrease in the pouring temperature reduces this critical value, since lowering the pouring temperature increased the incidence of blowholes. It follows, therefore, that as far as foundry practice is concerned, pouring temperature will have a considerable controlling influence on the occurrence of sub-surface blowholes.

In untreated cupola-melted iron, the sulphur content is not generally lower than 0.08 per cent. A working level of sulphur content is more likely to be of the order of 0.10 per cent to 0.14 per cent, dependent upon coke quality and with sulphur contents at this level the normally accepted manganese

content required to offset the deleterious effects of sulphur would be of the order of 0.47 per cent to 0.54 per cent manganese.[4] Very often the total available manganese content of the cupola charge does not attain the desired level, so that the manganese content is raised either by the use of a small proportion of a high manganese pig iron, or by adding ferro-manganese briquet to the charge. The experimental work shows that if a large excess of manganese over that required to balance the sulphur be used, the tendency to produce sub-surface blowholes becomes more marked.

The ladle slag analyses taken in the experimental melt A show that there is a progressive increase in the manganese and sulphur content as the metal temperature falls. It has not been possible as yet to collect slag inclusions from the defective castings in order to compare the analyses of such slag with the slag collected from the surface of the metal in the ladle. However, it is believed that the slag-like material present in the defective areas of the castings illustrated largely originates from the slag oxide complex which forms on the surface of the metal in the ladle. This slag may, therefore be termed a ladle slag. It should be noted that in the experimental work, no flux additions were made to the charge, and that the pig iron was shotblasted before charging. In this manner the dangers of contamination by extraneous siliceous materials were minimized as far as possible.

It has been observed in practice that when foundry pouring ladles are in use there is an opportunity for slag to coagulate on the ladle sides as the level of the metal falls. When ladles are refilled, the adhering slag may be liquified and added to the fresh slag in the ladle. Conditions of intermittent ladle usage, in which the ladle becomes cold between fillings, favors such ladle contaminations. It is suggested that this may provide an explanation for apparently anomalous conditions in which sub-surface blowholes have been observed in metal compositions with relatively low manganese and sulphur contents poured at normal temperatures. This feature of ladle contamination may also be a pointer in explaining why sub-surface blowholes can made sporadic appearances in the foundry.

It has been shown that under given conditions, as the metal temperature in the ladle falls, the manganese and sulphur contents of the ladle slag are increased. The series of sulphur prints reproduced in Fig. 15 show that in three test castings of similar chemical composition, sulphide segregation becomes more marked as the pouring temperature falls. It would be expected with higher pouring temperatures that the longer period intervening before solidification in the mold would increase the possibility of segregation. The evidence shows clearly that this is not the case, and it seems probable, therefore, that the segregation apparent in the test castings did, in fact, occur before the metal entered the mold cavity.

It is not proposed to discuss here the manner in which sulphide and slag segregations produce sub-surface blowholes of this type. It is felt that further investigation of slag/metal reactions at critical temperatures is necessary before valid comment can be made. Carbon and silicon could possibly have some bearing on the occurrence of the defect. The work carried out, however, suggests that if metal is poured at relatively low temperatures there is less opportunity for any gas evolved to escape completely from the metal.

As far as the foundryman is concerned, it would appear to be easier to exercise control over pouring temperature and possibly manganese content, rather than over the sulphur content. The virtues of high cupola melting temperatures are lost by tapping into cold ladles and by slow metal distribution. It is also suggested that close attention should be paid to ladle cleanliness, since ladles contamination may contribute to the formation of sub-surface blowhole defects.

Conclusions

The following conclusions can be drawn from this work, which has studied sub-surface blowholes and their association with manganese sulphide segregation in gray iron castings.

1. It has been possible, by experimental work, to reproduce the sub-surface blowholes which were exhibited in the examples of defective castings taken from different foundries. The features of manganese sulphide segregation and the inclusion of a slag-like material were also observed to be present in the defective areas in the test castings.

2. For a given manganese and sulphur content, the incidence of sub-surface blowholes is increased as the pouring temperature falls. The experimental work indicates that low pouring temperatures is the primary factor in promoting such defects.

3. At a given pouring temperature, the product, manganese content of the metal x the sulphur content of the metal, has a critical value which, when exceeded, promotes the occurrence of sub-surface blowholes.

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Acknowledgment

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DISCUSSION

Chairman: H. W. LOWNIE, JR., Battelle Memorial Institute, Columbus, Ohio.

Co-Chairman: G. A. TIMMONS, Climax Molybdenum Co., Detroit, Mich.

M. H. HORTON[1] *(Written Discussion):* The author made an excellent presentation of a common defect that is often visible only in the machine shop.

This type of defect has been seen repeatedly in machine shops and has been associated with poor gating technique, poor slagging and fluxing of the cupola, low temperature pouring, slag contaminated bull and pouring ladles and operating the mixing ladles at too low a level.

The paper clearly illustrates a defect the average foundryman will consider as a form of blow or blows coming from wet sand, low permeability, underbaked cores or wet ladles. He will seldom consider slag or manganese sulphide as the source of the defect.

The slag or cinderlike inclusion is usually present in this type of defect but is often removed upon machining and the defect is then considered a blow. We have often broken scrapped castings in the foundry and pinpointed the cause.

It has been found that this defect can be cured by properly gating the casting. The author associates this defect with low pouring temperatures and has data to prove this point but with poor gating practice one would expect this to increase with cold pouring temperatures if the defect is associated with slag.

It will be found that this type of defect can be produced with hot pouring if the gating system does not trap slag inclusions before reaching the casting.

The writer has often felt the defect has been directly caused by slag due to the extremely thin bluish material lining the cavity with a cinderlike deposit recessed somewhere in that cavity indicating space was occupied by some material. It could be possible a gas condition could exist and be entrapped in the slag globule.

The true formation of this defect is not clearly understood by the writer but having eliminated it by proper gating is led to believe slag is the culprit.

All cases cited by the writer had manganese content of 0.50 through 0.80 with normal sulphur ranges of 0.10 to 0.16.

The pouring temperatures used in the paper are considered low for our practice but may be suitable for large work.

It will be interesting to have the author's analysis of the ladle slag and the material which is found in the defect.

C. F. WALTON:[2] It is not necessary to force a chemical explanation between the occurrence of slag inclusions and associated gas bubbles. Their co-existence can be due to the nucleation effect of the slag in the formation of the gas bubbles from a saturated solution.

R. D. AHLES:[3] Much the same problem of finding subsurface holes associated with small nonmetallic particles is found in steel castings. If we visualize these nonmetallics as forming on the surface of the metal rising in the mold, how is it that they are found 1/8 to 1/4 in. below the cope surface? Can it be that these high manganese and sulphur inclusions react with the hydrogen contained in the iron to produce H_2S gas which causes the observed holes?

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1. Deere & Co., Moline, Ill.
2. Gray Iron Founders Society, Cleveland.
3. General Electric Co., Schenectady, N. Y.

TEMPERATURE DROP IN POURING LADLES

By

Victor Paschkis*

Introduction

Previous work has shown the great significance of the pouring temperature of the metal for the solidification behavior of castings.[1,2] The term "pouring temperature" was used to describe the temperature of the metal as it enters the mold cavity. If that temperature has been selected, the required temperature at which to tap the melting furnace depends on the temperature drop which the metal suffers in transit. The present study deals with what is believed to be the greatest single contributing factor to such temperature drop, the ladle.

The other factors are the heat loss from the metal stream issuing from the furnace when filling the ladle, and the stream issuing from the ladle when filling the casting, and finally the heat loss during passage through sprues, runners and gates. Although there is a definite loss due to the latter items (stream ladle-mold; flow through gate, etc.), this drop is disregarded in the present study by calling the temperature of the metal at the moment of entering the ladle the "tapping temperature" and that on leaving the ladle the "pouring temperature."

It is a well-established practice to reduce the temperature drop by preheating the ladle; but in the present paper an effort is made to present such influence in quantitative terms. At tapping temperature—(This term will be used throughout the paper to define the temperature of the metal entering the ladle, assuming the filling time of the ladle to be zero)—the metal has a certain heat content; for convenience of presentation, in Fig. 1, this heat content is taken with the solidus point as reference temperature. In other words, this heat content at tapping temperature is expressed by:

[(tapping temperature—solidus temperature) x specific heat + heat of fusion] weight of metal in ladle.

One can describe this heat content at tapping by arbitrarily assigning to it a given width in the diagram of Fig. 1. In this figure the widths between the various lines indicate heat quantities. During the sojourn of the metal in the ladle, the heat content at tapping is broken down into two major components: the heat content at pouring and the heat losses. In general, it is desired to keep the latter as small as possible since small heat losses permit, for any given pouring temperature, the use of a lower tapping temperature, i.e., to operate the melting units at a lower temperature.

Heat losses can be separated into two components: heat losses through walls of the ladle, comprising sides and bottom, and heat losses through top. The heat flowing from the metal into the walls is used to cover two kinds of losses: part of the heat is stored in the ladle walls (storage), and part is given off by radiation and convection from the outside surface (radiation). The heat lost from the surface is used for the same two types of losses (storage and radiation) if a lid or other covering is used, or only for surface losses if no covering is used. In the latter case, storage is zero.

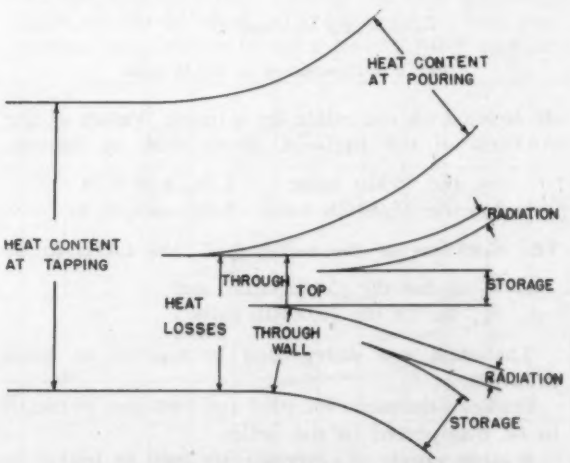


Fig. 1 — Schematic distribution of heat losses.

*Technical Director, Heat & Mass Flow Analyzer Laboratory and Adjunct Associate Professor, Department of Mechanical Engineering, Columbia University, New York.

This report is based on an AFS-sponsored Research Project by the Heat Transfer Committee. Members of the AFS Heat Transfer Committee are as follows: Charles Locke, *Chairman*, W. K. Bock, C. W. Briggs, J. B. Caine, W. J. Childs, W. A. Mader, V. Paschkis, W. S. Pellini, R. C. Shnay, C. E. Sims, E. C. Troy, and L. L. Zinsmeister.

All investigations described in this work were carried out by computations on the Heat and Mass Flow Analyzer.[3,4]

The pouring temperature, in which the foundryman is primarily interested, is, of course, directly related to the heat content of the metal at the moment of pouring. Pouring temperature and the heat content at pouring depend on the following variables:

tapping temperature (Heat Content at Tapping); the greater this value is, the higher can the losses be for a given Heat Content at Pouring
lining thickness
material of lining
thickness of lid, if any
material of covering
size and shape of ladle
properties of metal

The present study comprises two ladle sizes: one with a nominal holding capacity of 150 lb, and one with a holding capacity of 16,000 lb. The shapes of the metal shells are shown in Figs. 2 and 3; these shells

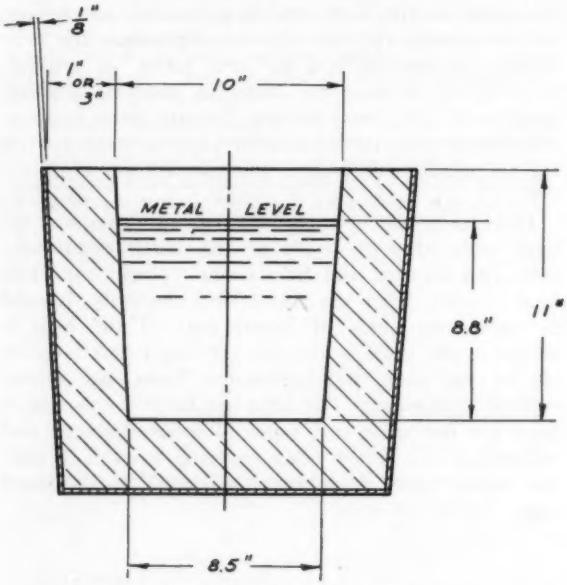


Fig. 2 — Dimensions of 150-lb ladle.

are covered on the inside by a lining. Values of the thickness of the firebrick lining were as follows:

for the 150-lb ladle 1 in. and 3 in.
for the 16,000-lb ladle 6 in. and 10 in.

The thickness of the metal shell was taken to be

1/8 in. for the 150-lb ladle, and
5/16 in. for the 16,000-lb ladle

The shell was disregarded because of its small thickness.

Studies were made for steel and cast iron as metals to be transported in the ladles.

A large variety of coverings are used in foundries. Since it would be impractical to investigate so many different coverings, two limits were chosen for this study: in one case the metal surface was bare, giving off heat directly to the shop; and in the other case the metal was covered by a lid. The lid was assumed to be made of the same material as the lining (firebrick) and to have the same thickness as the lining. It is not customary to preheat the lid; early studies led to the inclusion of the case of preheated lids in the investigation.

A special study was made of the influence of preheating of the ladles and lids. Much of the contradictory experience in ladle practice can probably be attributed to confusion regarding preheating. The inside surface temperature which the ladle reaches is not sufficient to describe the effectiveness of the preheat; the heat stored in the ladle wall prior to tapping is also of importance.

Figure 4 may be helpful in understanding the preheating of the ladle: it represents, schematically, a cross-section through the ladle wall, consisting of the lining and the shell. The several lines represent the temperature distribution at consecutive times. Consider first the condition of the ladle being fully preheated, indicated by the heavy straight line at the extreme left. The "fully preheated condition" is characterized by a constant outside surface temperature; no matter how long the heating on the inside continues, with a constant temperature of the flame used for preheating, the outside temperature will not rise any further. The magnitude of this outside temperature depends on:

the inside wall temperature
inside diameter of lining
lining thickness
material of lining (thermal conductivity)
thickness of metal shell of ladle*
material of shell (thermal conductivity) *
thermal contact resistance between lining and shell*
boundary conductance (surface resistance)

*As is shown in the figure, the temperature drop across the shell is so small that it is permissible to disregard the latter, at least in small ladles.

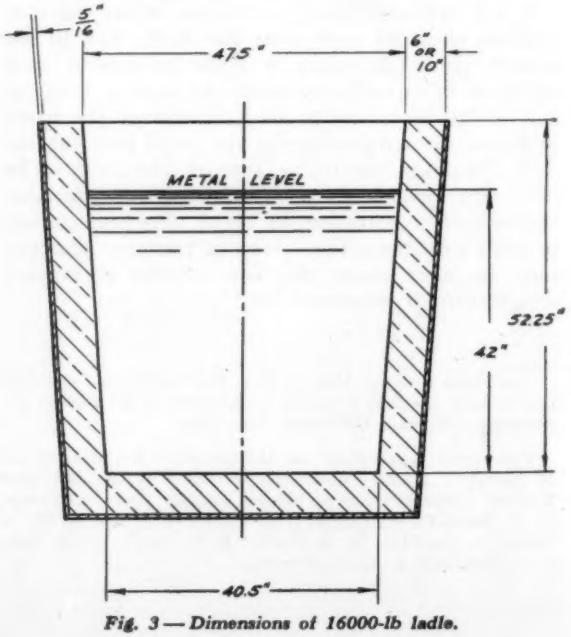


Fig. 3 — Dimensions of 16000-lb ladle.

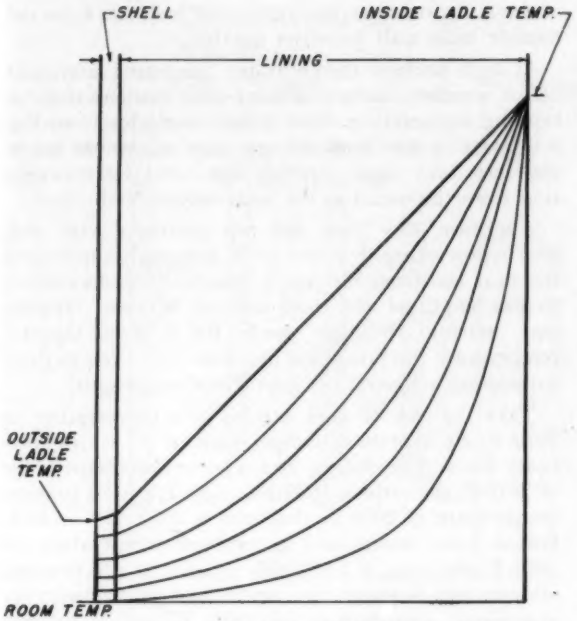


Fig. 4 — Temperature distribution during preheating of ladle.

The course of the lines at intermediate times, i.e., before the final condition, indicated by the heavy line, is reached, and the time required to reach the final condition depend on the characteristics of the burner used in preheating. Therefore, it is not practical to rely on the length of time of preheating to determine whether the ladle is fully preheated. The time is a useful gage in this connection only if other operating conditions are carefully maintained constant. Instead, the operator can observe the outside temperature by means of a surface thermocouple, today obtainable in robust design. If this temperature

remains constant and
has a certain value,

then the ladle is fully preheated to a specified inside temperature. The value of this inside temperature for a certain degree of outside temperature depends on the thickness and conductivity of the lining, and the heat flux across the latter.

The latter is determined by the "boundary conductance" between outside surface and ambient (shop air), which in turn is a function of the outside surface temperature. For the conductivity used in the present study (1.0 Btu/ft,hr,F) the relationship between inside and outside temperature for several values of lining thickness is shown in Fig. 5.

Metal Temperatures

It was stated earlier that it is assumed that both the filling times of the ladle and of the mold are zero; under such conditions one can assume that the temperature of the metal at the moments when ladle or mold, respectively, are filled is uniform. But at the end of the period of holding the metal in the ladle, the temperature will not be uniform. The degree of non-uniformity is controlled by the mixing of the metal in the ladle and this mixing, in turn, depends on the amount of convection currents set up in the metal.

These convection currents are hard to predict, and therefore one can consider two limiting cases: In the first, no mixing occurs, the metal is stagnant, and heat flow within the metal is by conduction only. The second limit is given by so violent a stirring that there are no temperature differences in the metal; the latter is completely mixed. True conditions lie somewhere between those two extremes. In the present report only the first limit—that of no mixing —is considered. This limit results in the prediction of a smaller temperature drop than will actually occur. The assumption of complete mixing (no temperature differences within the ladle) leads to a larger temperature drop (lower temperatures after any given time) than will actually obtain.

150-lb Ladles for Steel

No Lid: Conditions without a lid are investigated first. Figure 6 refers to a ladle with a 1-in. lining and Fig. 7 to one with a 3-in. lining.

In these figures holding times are plotted on the vertical axis, and pouring temperatures are plotted on the horizontal one. Each of the two figures contains six curves: three solid ones, which hold for a preheat temperature of the ladle of 2070 F and three

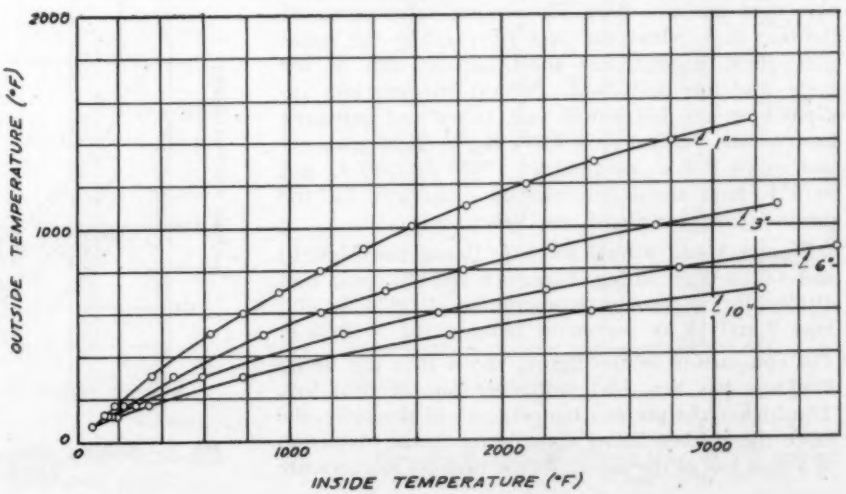


Fig. 5 — Outside vs inside temperature of preheated lining.

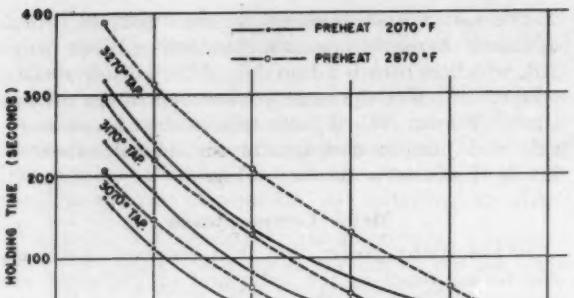


Fig. 6 — Holding time vs pouring temperature; 1-in. lining 150-lb ladle, steel, no lid.

broken ones, holding for a preheat temperature of 2870 F. The temperatures on the curves indicate the tapping temperatures: there are two curves (one solid, one broken) for a tapping temperature of 3270 F, two for a tapping temperature of 3170 F, and two for a tapping temperature of 3070 F.

If the pouring temperature equals the tapping temperature, the permissible holding time is obviously zero; as the pouring temperature goes down, the permissible holding time increases. For example, if metal is tapped from the melting furnace at 3070 F and should be poured into the mold at 2870 F, and if the ladle is preheated to 2070 F, the emptying of the ladle must be completed in 112 sec (1-in. lining).

A preheat temperature of 2870 F, which is shown in the figures, may at first seem impractical. These curves were introduced, however, to indicate the permissible holding times after the ladle was filled several times. In such repeated use the preheating of the ladle will increase and may, practically, end at the value of 2870 F.

By way of example, the influence of preheating is shown for a tapping temperature of 3070 F in Figs. 8 to 11. These figures show both the times required to drop from the tapping temperature to the pouring temperature for several degrees of preheat, and the heat loss of the metal, indicated by the height of the bars over the zero line. The several subdivisions of the bars show where the heat given off by the metal goes (heat storage, loss from outside wall of the ladle, and loss from top). Preheat temperatures are plotted on the horizontal axis, times and extracted heat on the vertical axis. Over the preheat temperature of 70 F (i.e., no preheat), 1070 F, 2070 F, and 2870 F, bars are drawn, shaded differently for the different components of heat losses.

Figures 8 and 9 hold for 1-in. lining and Figs. 10 and 11 for 3-in. lining. Figures 8 and 10 show conditions for a pouring temperature of 2870 F, and Figs. 9 and 11 for a pouring temperature of 2820 F.

A comparison of the figures shows that the lining thickness has very little influence on the heat loss. The higher the preheat temperature of the ladle, the more the thicker lining contributes to the reduction of a heat loss of the metal. As the preheat temperature of the ladle increases, the amount of heat loss from the outside ladle wall increases notably.

At high preheat temperatures the lining may, just before pouring, have a smaller heat content than at tapping temperature. This is indicated clearly in Fig. 9 by starting the "heat storage" part of the bar below the zero line (and showing the total heat extraction from the metal as the total height of the bar).

The heat loss from the top increases with preheat temperature. For any given tapping temperature the heat loss from the top is practically proportional to the length of the time interval between tapping and pouring. In other words, for a given tapping temperature the ratio heat loss from top/time to drop to pouring temperature is practically constant.

Take the case of steel, tapped at a temperature of 3070 F. At a preheat temperature of 70 F (no preheat) for a 1-in. lining and a pouring temperature of 2870 F, the ratio is 1690/66 = 25.6; a for a preheat temperature of 2070 F, the ratio is 2970/117 = 25.4. For a 3-in. lining and a preheat temperature of 1070 F, the ratio is 2440/100 = 24.4. If the pouring temperature is lower, the figure changes slightly: for a pouring temperature of 2820 F and a preheat temperature of 1070 F, the ratio is 3635/150 = 24.2 (1-in. lining), or 3414/150 = 23.4 (3-in. lining). Total heat loss in case of no lid can be shown schematically to be composed of two components, as in Fig. 12A. The heat loss from the top increases linearly with the "holding time" of the metal in the ladle. The losses from the sidewall decrease, including heat storage and losses from the outside surface of the lining.

With Lid: Presentation of results is in the same manner as for the case of a ladle without a lid. Figures 13 and 14 show times plotted against pouring temperatures for 1-in. and 3-in. linings, respectively.

The lid is assumed to be of the same material and same thickness as the lining of the ladle proper. Except for a ladle of small thickness without preheat, the ladle with lid cools appreciably slower than the

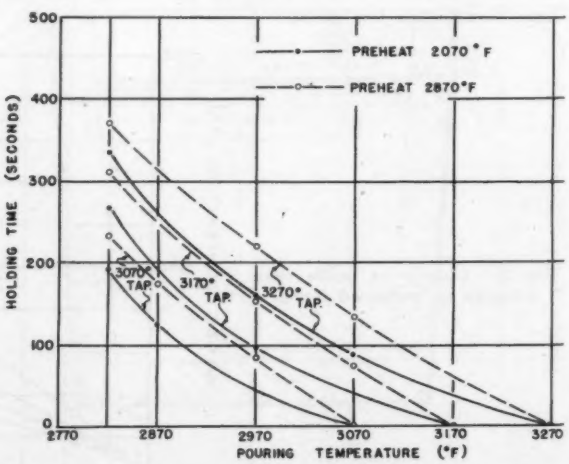


Fig. 7 — Holding time vs pouring temperature; 3-in. lining 150-lb ladle, steel, no lid.

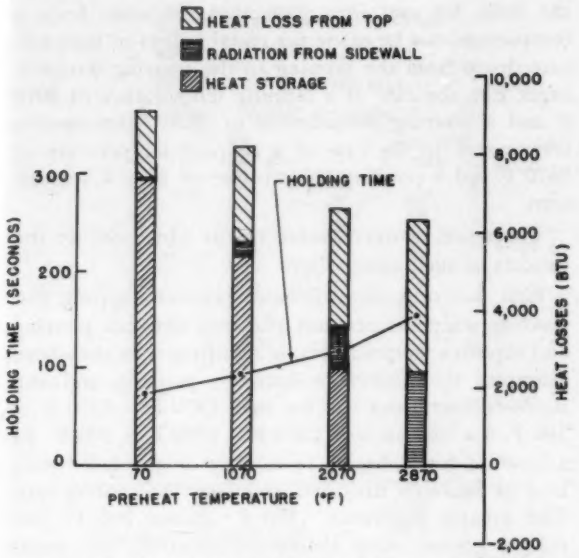


Fig. 8 — Heat distribution and holding time; 2870 F pouring temperature; 1-in. lining, 150-lb ladle, steel, no lid.

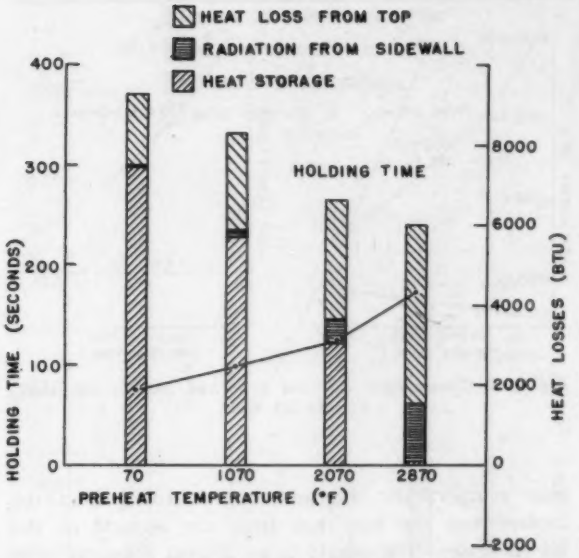


Fig. 10 — Heat distribution and holding time; 2870 F pouring temperature; 3-in. lining, 150-lb ladle, steel, no lid.

ladle without lid. The advantage of the lid is greater where there is a higher preheat and a greater thickness; both imply a greater amount of heat stored.

Similar conclusions can be drawn from Figs. 15 to 18, which correspond to Figs. 8 to 11. However, there are now four components to absorb the heat loss from the metal in the ladle: Heat storage in and heat loss from the ladle walls, and storage in and losses from the lid.

Heat losses per unit of inside surface area, both by storage and from the outside, are smaller for the lid than for the side wall, because there is an additional heat transfer resistance between metal and lining in case of the lid, while no such resistance exists between metal and sidewall. Figures 15 to 18 indicate that there is no appreciable change of heat losses from the top with changing length of time or changing degree of preheat. But the pouring temperature exerts a significant influence on the heat losses. Thus, a characteristic picture, as in Fig. 12B, obtains. If the lid is not preheated, it absorbs a relatively large amount of stored heat; at increasing values of pre-

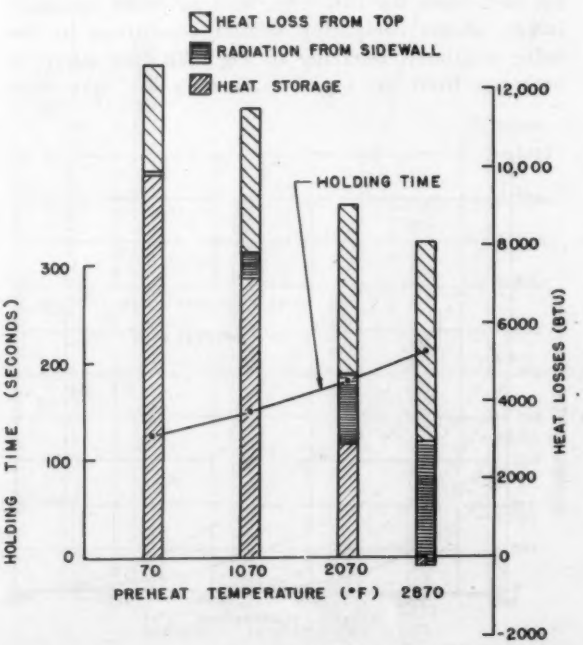


Fig. 9 — Heat distribution and holding time; 2820 F pouring temperature; 1-in. lining, 150-lb ladle, steel, no lid.

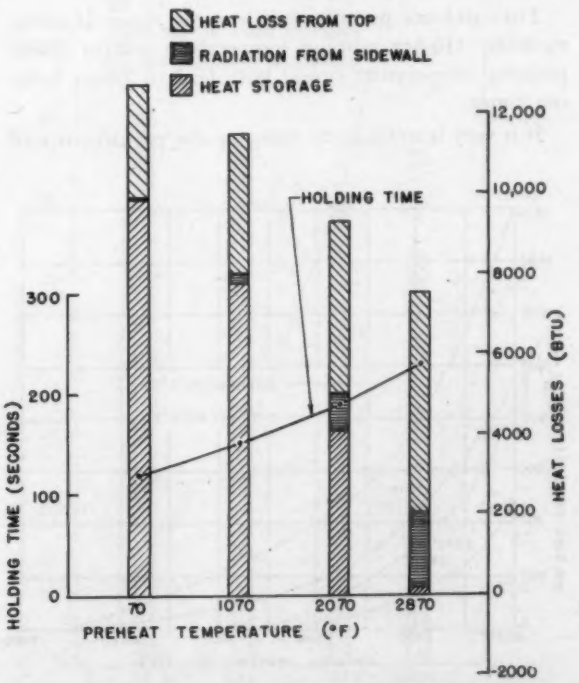


Fig. 11 — Heat distribution and holding time; 2820 F pouring temperature; 3-in. lining, 150-lb ladle, steel, no lid.

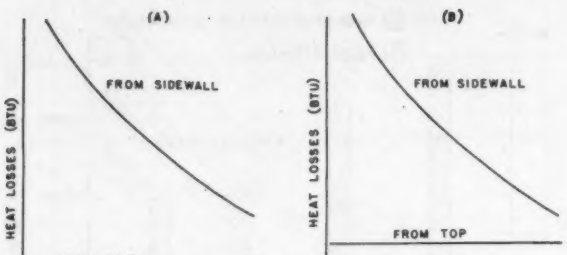


Fig. 12 — Composition of total heat loss. A. No lid, steel. B. with lid, steel.

heat temperature the amount of stored heat decreases, but the heat loss from the outside of the lid increases. The result is an almost constant heat loss from the top (lid), while the character of the curve, sidewall losses vs. holding time, is materially unchanged.

150-lb Ladle for Cast Iron

For the case of cast iron, only one tapping temperature, 2870 F, was investigated. It does not seem necessary to show the distribution of heat for each case. Figure 19 holds for the ladle with 1-in. lining, Fig. 20 for the 3-in. lining. Each figure contains two solid lines (no lid) and two broken ones (with lid). Two curves hold for a pouring temperature of 2670 F and two for a pouring temperature of 2620 F.

Time did not permit extending the range of investigation. Higher tapping temperature and/or lower pouring temperature would both lead to longer holding times.

It is very interesting to compare the performance of the ladle for cast iron with that for steel. Such a comparison can be made for equal values of temperature drops from the tapping to the pouring temperatures, e.g., the case of a tapping temperature of 3070 F and a pouring temperature of 2870 F for steel is comparable to the case of a tapping temperature of 2870 F and a pouring temperature of 2670 F for cast iron.

There are, however, some points which reduce the validity of such comparison:

First, not only the difference between tapping and pouring temperatures, but also that between pouring and liquidus temperatures, is significant. In the above example, the difference between pouring and liquidus temperatures was for steel 2870 F − 2770 F = 100 F, but for cast iron 2670 F − 2280 F = 390 F. As a layer of metal begins to solidify on the ladle wall, heat of fusion is liberated, reducing the cooling rate. The greater difference (390 F against 100 F) permits a greater drop before solidfication; the result is a shorter time to achieve such drop.

Secondly, the total heat loss (Btu) is smaller for lower temperatures, such as prevail in the case of cast iron.

Thirdly, the rate of heat losses (Btu/hr) is lower at the lower temperatures prevailing in pouring cast iron. The time to achieve a given heat loss (expressed in Btu) becomes longer, as the rate of heat loss (expressed in Btu/hr) becomes smaller.

But conversely, the time becomes shorter if a given rate of heat loss applies to a smaller total loss. Since both total loss and rate of heat loss are smaller in the case of cast iron than in the case of steel, the outcome, namely, the cooling time is in some instances smaller for steel than for cast iron, and in other instances larger, always comparing similar conditions in the ladle. Similarly, referring to Fig. 12B, the curve of heat loss from the top, in case of a lid, may show

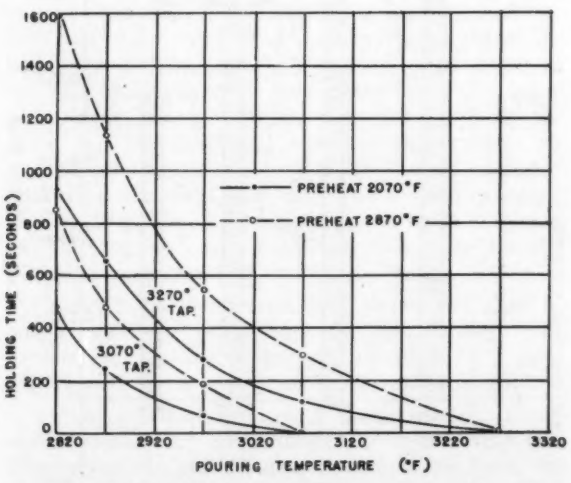


Fig. 13 — Holding time vs pouring temperature; 1-in. lining 150-lb ladle, steel, with lid.

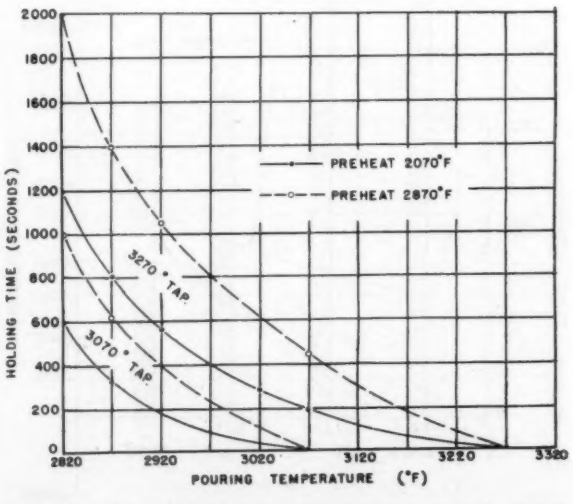


Fig. 14 — Holding time vs pouring temperature; 3-in. lining 150-lb ladle, steel, with lid.

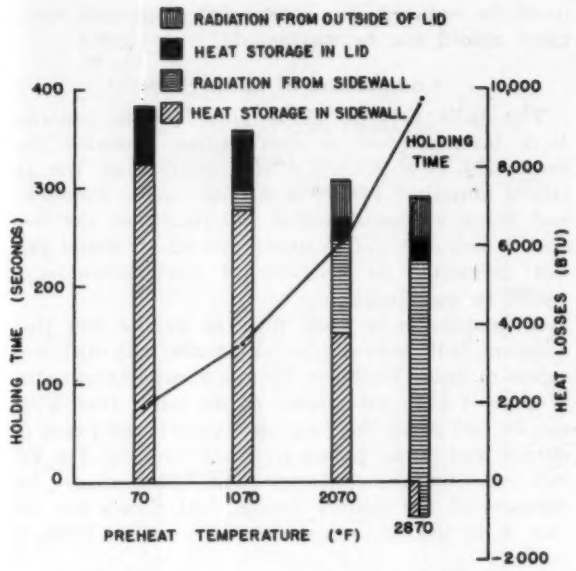


Fig. 15 — Heat distribution and holding time; 2870 F pouring temperature; 1-in. lining 150-lb ladle, steel, with lid.

a somewhat different course than the smooth decline shown for steel.

16000-lb Ladle

The same cases have been studied for this ladle size as for the smaller one. Space does not permit presentation of graphs similar to Figs. 8 to 11 for all

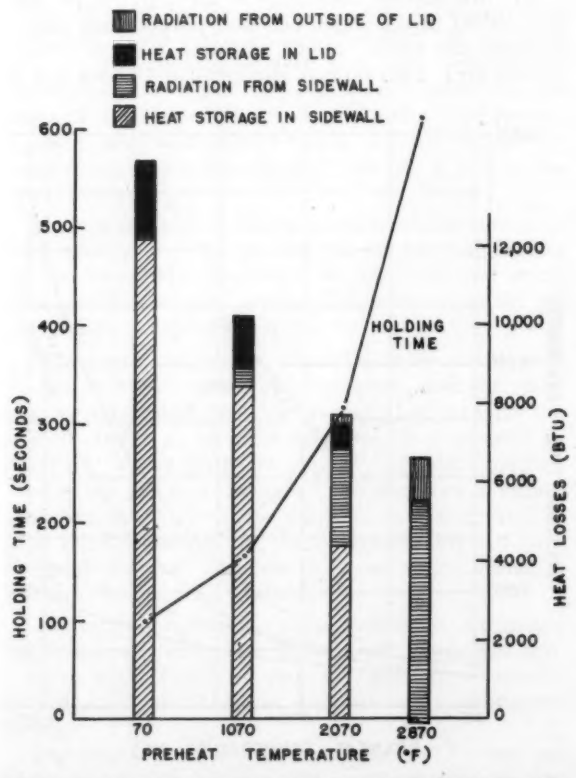


Fig. 17 — Heat distribution and holding time; 2870 F pouring temperature; 3-in. lining 150-lb ladle, steel, with lid.

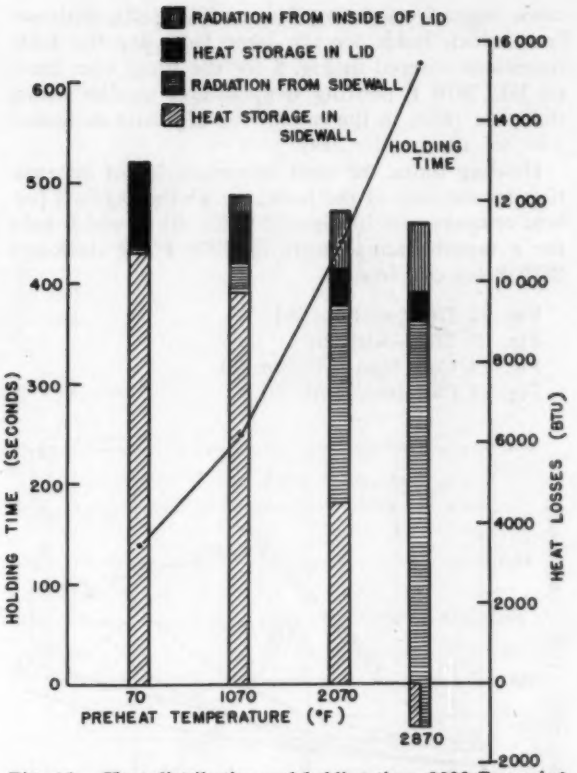


Fig. 16 — Heat distribution and holding time; 2820 F pouring temperature; 1-in. lining 150-lb ladle, steel, with lid.

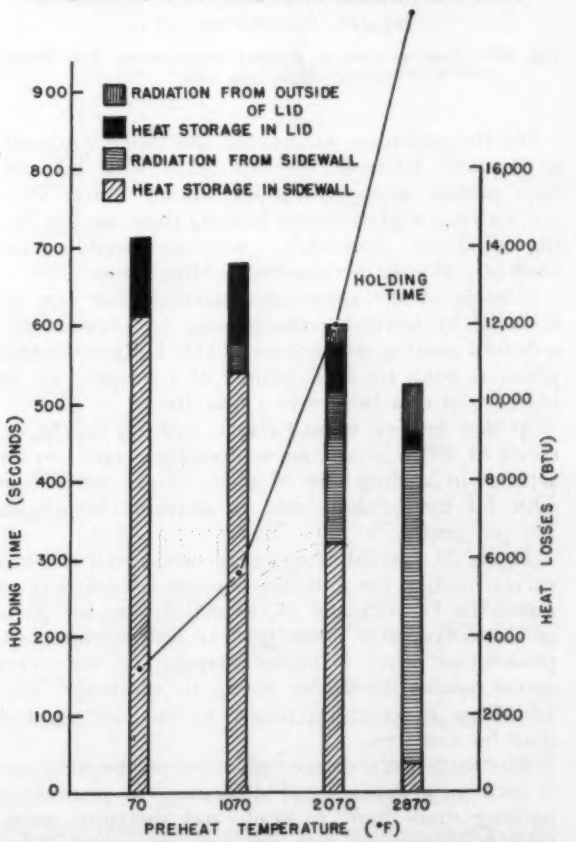


Fig. 18 — Heat distribution and holding time; 2820 F pouring temperature; 3-in. lining 150-lb ladle, steel, with lid.

cases. Instead, only one such graph, Fig. 21, is shown here, which holds for the large ladle for the same conditions covered in Fig. 8 for the small one: steel, no lid; 2870 F pouring temperature; smaller lining thickness (6-in. in the case of the big ladle as against 1-in. for the smaller one).

Holding times, the most important bit of information for the user of the ladle, are plotted against preheat temperatures in Figs. 22 to 25, all of which hold for a tapping temperature of 3070 F for steel and 2870 F for cast iron:

Fig. 22 Steel—without lid
Fig. 23 Steel—with lid
Fig. 24 Cast Iron—without lid
Fig. 25 Cast Iron—with lid

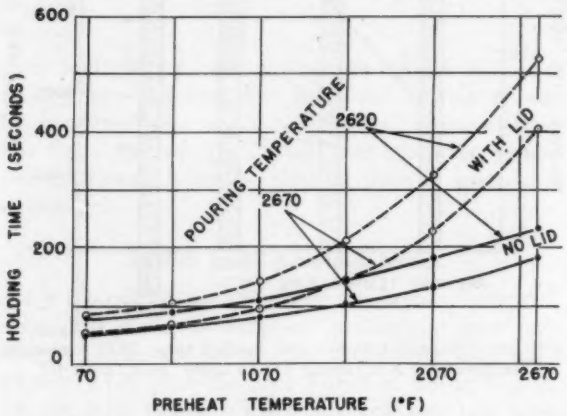


Fig. 19 — Holding time vs preheat temperature; 1-in. lining 150-lb ladle, cast iron.

For the operation without lid the lining thickness is of small influence; for operation with lid and high preheat temperatures the thicker lining obviously gives a slightly longer holding time; smaller lining thickness than 6-in., however, would show markedly shorter permissible holding times.

Increase of the permissible holding time can be achieved by increasing the tapping temperature for a desired pouring temperature. Table 1 illustrates this point. It holds for steel, poured at a temperature of 2820 F, out of a ladle with a 6-in. lining.

As may be seen from Table 1, with no lid the increase of 200 F in tapping temperature results in an increase in holding time of about 70 per cent, while with lid the holding time is increased by almost 100 per cent.

Figure 24 (no lid) shows only two curves, the two curves holding for different pouring temperature as noted. In Figs. 23 and 25 (with lid) the left parts of the curves (for lower preheat temperatures) approach each other. At higher temperatures the curves spread, giving the thicker lining an advantage. This advantage is somewhat bigger in the case of steel than for cast iron.

The nature of the curves obtained on the computer is such as to make small differences in permissible holding times hard to read, and therefore uncertainties of perhaps ± 5 per cent may exist. They could be reduced, but it was felt that such extra effort would not be warranted.

Comparison of Ladle Sizes

The ladle sizes of 150-lb and 16000-lb capacity have been selected as representing probably the limits of sizes of interest to the foundryman. But an almost unlimited variety of sizes is used in foundries and hence a comparison of the results of the two sizes is desirable; also, a correlation which would permit estimation of behavior of intermediate sizes would be significant.

A comparison is made difficult by the fact that different ladle sizes are usually made with different values of lining thickness. From a viewpoint of radiation losses from the outside of the ladle, two ladles can be said to be lined equally heavy if the ratios of outside and inside surface areas are the same. For the two ladles under consideration this holds, almost, for the case of the thinner linings, but poorly for the case of the thicker linings, as may be seen in Table 2.

TABLE 2 — SURFACE AND VALUES/SURFACE RATIOS

Ladle Size, lbs	Lining Thickness, in.	Surface Ratio	Volume/ Surface Ratio ft
150	1	1.4	0.11
16000	6	1.49	0.72
150	3	2.37	0.46
16000	10	1.88	1.36

But this surface area ratio is meaningless for the heat stored in the walls. For that purpose one might compare the ratio: volume of lining material/inside surface area. This ratio is also tabulated above but is

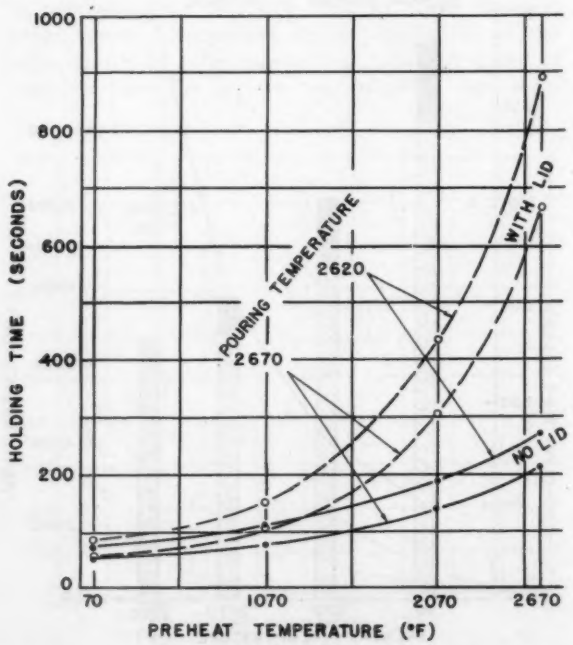


Fig. 20 — Holding time vs preheat temperature; 3-in. lining 150-lb ladle, cast iron.

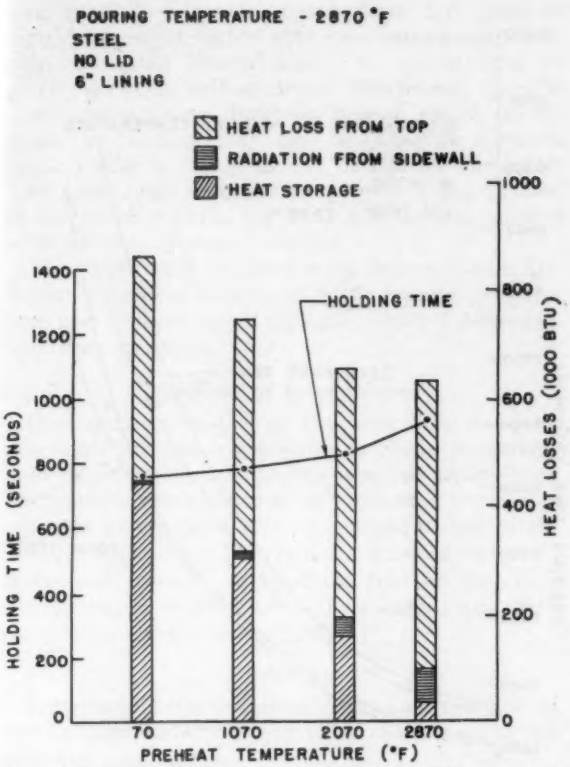


Fig. 21 — Heat distribution and holding time; 2870 F pouring temperature, 6-in. lining, 16000-lb ladle, steel, no lid.

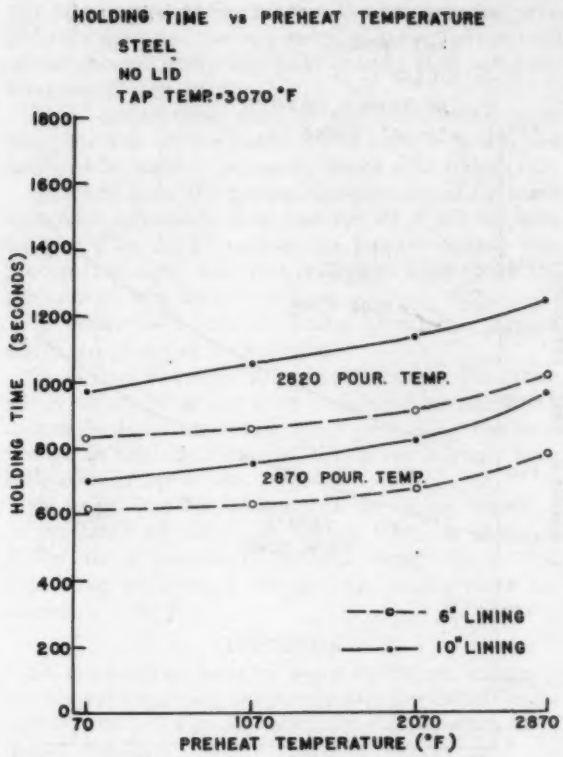


Fig. 22 — Holding time vs preheat temperature; steel, no lid, 3070 F tapping temperature.

quite different for the small and large ladle. A second difficulty arises from the fact that part of the losses are through the top surface and part through the sidewalls. If one wanted to compare ladles on the basis of a volume (of metal) to surface (of metal) ratio, a large uncertainty would prevail: should the top surface, losing heat through the lid, if any, be included or not?

Because of these difficulties, comparisons and general conclusions are drawn, using the four ladles which have been considered so far (big and small ladle, each with two lining thicknesses) without attempting a complete quantitative analysis.

In any given situation the metal to be transported in the ladle, the pouring temperature, and the ladle size are prescribed; ladle covering (lid or no lid), lining thickness, tapping temperature, and preheat of the ladle can be varied to achieve a required holding time of the metal in the ladle. The questions of choice of lining material, of the shape of the ladle, and of different thickness of lid, if any, against sidewall have not been studied. The four first-mentioned variables will now be discussed individually.

The *tapping temperature* has great influence on the holding time. The percentage values given (in connection with Table 1) seem to hold approximately for big and small ladles and for thin and thick linings.

The *preheat temperature* is important for any set of conditions. The impact of the preheat is, under any combination of conditions, larger for the 150-lb ladle

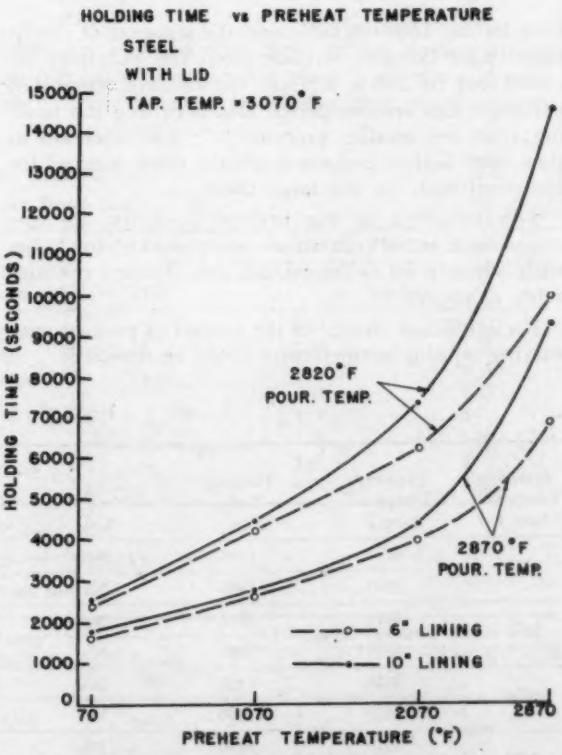


Fig. 23 — Holding time vs preheat temperature; steel, with lid, 3070 F tapping temperature.

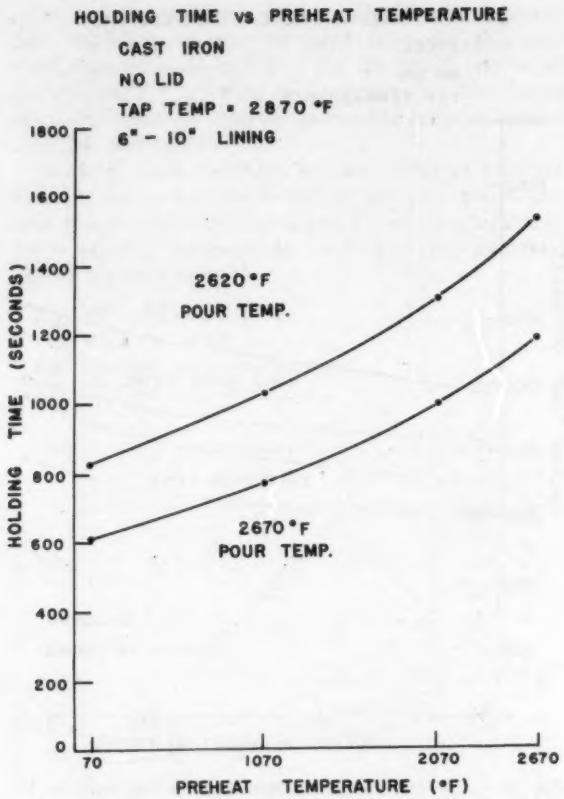


Fig. 24 — Holding time vs preheat temperature; cast iron, no lid, 2870 F tapping temperature.

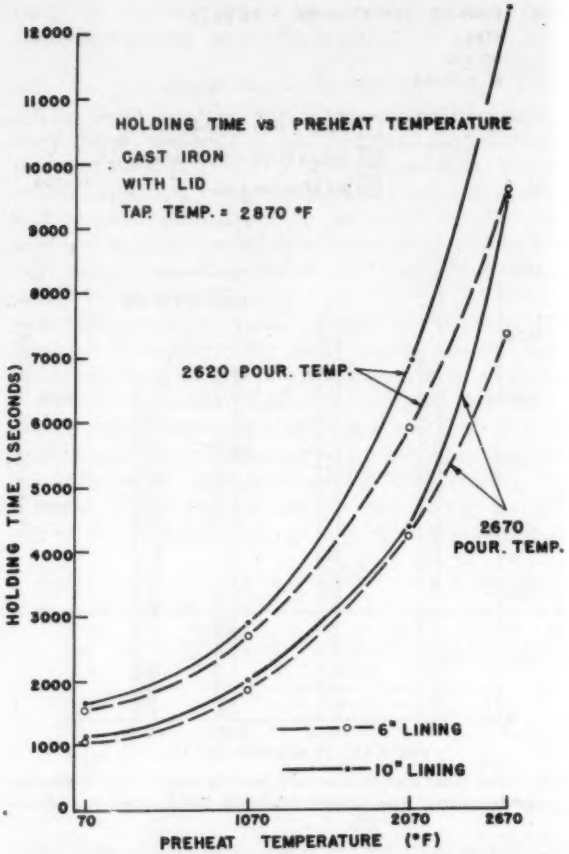


Fig. 25 — Holding time vs preheat temperature; cast iron, with lid, 2870 F tapping temperature.

than for the 16000-lb ladle and the same set of conditions. This becomes obvious from the examples incorporated in Table 3. Here the relative lengths of holding times are compared, always calling the holding time for smaller preheat "1". The increase in time with higher preheat is always more marked for the small than for the large ladle.

The influence of the preheat is—again, for any comparable set of conditions—more marked for ladles with lids than for ladles without lids. Table 4 contains a few examples.

No significant change of the impact of preheat with varying tapping temperatures could be detected.

It is not possible to express in concise terms the relationship of preheat and metal to be cast. Steel and cast iron are tapped and poured at different temperatures. Disregarding these differences a comparison of the increase in holding times for the two metals (each of which is once tapped into a non-preheated ladle and once in a ladle preheated to 2070 F), indicates that for ladles of identical design such preheat is more important for cast iron than for steel.

Lining thickness, in the values which have been

TABLE 1 — INFLUENCE OF TAPPING TEMPERATURE

Preheat Temperature, F	Tapping Temperature, F	Holding Time, sec	Lid	Heat Loss Total, 1000 Btu	Heat Storage Walls 1000 Btu	Heat Storage Lid 1000 Btu	Radiation Walls 1000 Btu	Radiation Top 1000 Btu
2070	3070	1130	No	854	200	—	50	604
	3270	1990	No	1580	316	—	84	1180
2870	3070	1220	No	788	19	—	88	681
	3270	2060	No	1490	43	—	157	1290
2070	3070	6120	Yes	1060	490	135	350	85
	3270	11900	Yes	1840	670	200	790	180
2870	3070	9840	Yes	964	—76	34	827	179
	3270	17300	Yes	1840	—10	45	1460	345

considered, is relatively unimportant, but gains in significance with higher preheat temperature. However, very thin linings would, of course, lead to greatly reduced holding times. Conversely, there is a maximum lining thickness, beyond which no increase in holding time can be achieved, however much lining is being added. In case of the small ladle, this limit is probably beyond 3 in.; in case of the 16000-lb ladle, the limit may be just reached with the 10-in. lining.

The importance of using a *lid* increases with the degree of preheat applied, with the tapping temperature, and with the size of the ladle. Table 5 shows two significant examples.

Prospects of Interpolation

Having extensive data on two ladle sizes, the question arises whether it is possible to derive from them holding times for intermediate sizes of ladles. Such interpolation would have to be based on a correlation per unit weight, unit area, or a combination of the two. No such simple correlation is possible on purely theoretical grounds. Attempts to find an empirical correlation for either times or heat losses so far have not yielded any promising results.

Conclusions

From these thermal considerations some conclusions can be drawn which should be of importance to the practical foundryman:

1) Preheating of ladles is essential. For the small ladle, see Figs. 8 to 11 and Figs. 15 to 18; for the large ladle, compare any of the appropriate lines in Tables 1, 3, and 4.

2) In preheating to a certain temperature the time plays an important role; proper preheating to a given inside temperature can be considered complete only when the lining is fully soaked with heat; this state can be recognized by the fact that the outside temperature does not rise any more. For the linings used in the present study, the final outside ladle temperatures are shown in Fig. 5.

3) A considerable part of the heat losses occur from the top of the ladle. Some kind of protection against heat losses is necessary; use of a lid is desirable.

4) The higher the preheat temperature of the ladle the more important does the use of a lid become; without a lid the losses from the top are directly proportional to time, but with a lid the losses from the top change very little with time.

5) There is a critical lining thickness, beyond which no gain can be expected.

6) It does not seem important whether the lid itself is preheated or not since preheating of the lid increases its radiation losses while decreasing the heat storage in the lid. However, this point has not been fully cleared up as yet.

7) Increasing the thickness of the lining beyond 1 in. for ladles of 150 lb, or beyond 6 in. for ladles of 16,000 lb is practically without significance. This holds true particularly for preheat temperatures not exceeding 2100 F.

APPENDIX

All calculations must be based on certain values of the several variables; these variables are listed below in the same order as the materials, etc., are mentioned in the text.

The properties of the lining were introduced into the calculations with the following values:

thermal conductivity	1.0	Btu/ft,hr,F
specific heat	0.266	Btu/lb,F
density	170	lb/cu ft

It was assumed that these values do not change with temperature and that there is no thermal resistance (and, therefore, no temperature drop) between

TABLE 3 — INFLUENCE OF PREHEAT
Comparison of Large and Small Ladle

Preheat Temperature, F	Tapping Temperature, F	Metal	Pouring Temperature, F	Lining Thickness	Lid	Relative Holding Times, Small Ladle	Relative Holding Times, Large Ladle
70 2870	3070	Steel	2870	Thin	No	1 2.35	1 1.23
70 2870	3070	Steel	2870	Thin	Yes	1 5.11	1 4.25
2070 2870	3270	Steel	2820	Thin	No	1 1.29	1 1.09
70 2670	2870	Cast Iron	2670	Thick	No	1 4.20	1 1.95

TABLE 4 — INFLUENCE OF PREHEAT
With and Without Lid

Preheat Temperature, F	Tapping Temperature, F	Metal	Pouring Temperature, F	Lining Thickness	Ladle Size	Relative Holding Times, Without Lid	Relative Holding Times, With Lid
70 2870	3070	Steel	2870	Thick	Small	1 2.35	1 6.10
70 2670	2870	Cast Iron	2620	Thin	Large	1 1.81	1 6.18

TABLE 5 — LID AND LADLE SIZES

Lid	Lining	Tapping Temperature, F	Preheat Temperature, F	Pouring Temperature, F	Material	Relative Holding Time, Small Ladle	Relative Holding Time, Large Ladle
No	Thick	3070	2070	2870	Steel	1	1
Yes						1.91	2.92
No	Thin	2870	2070	2670	Cast Iron	1	1
Yes						1.99	3.28

the metal in the ladle and the inside surface of the lining and between the outside surface of the lining and the metal shell.

The boundary conductance between the outside of lining and the shop was taken as a function of surface temperature, varying from 1.5 Btu/sq ft, hr, F at 100 F to 14.2 Btu/sq ft, hr, F at 150 F.

The properties of the metal as used in the calculations were as shown in Table 6.

The thermal properties used in the computations of the heat loss from the top were as follows:

bare metal : emissivity 0.85; (It is recognized that the emissivity would change with the degree of oxidation of the metal, but lacking more specific information the value of 0.85 was used throughout the study); the boundary conductance changes of course with the temperature of the metal surface.

lid : boundary conductance between metal and lid 177 Btu/sq ft, hr, F; thermal properties of lid and boundary conductance between lid and shop were taken to be the same as in the lining.

TABLE 6

		Steel		Cast Iron	
thermal conductivity,	liquid	9.2	Btu/ft, hr, F	9.2	Btu/ft, hr, F
thermal conductivity,	solid	18.4	Btu/ft, hr, F	18.4	Btu/ft, hr, F
specific heat	liquid	0.200	Btu/lb, F	0.200	Btu/lb, F
specific heat	solid	0.165	Btu/lb, F	0.165	Btu/lb, F
density	liquid	450	lb/cu ft	450	lb/cu ft
density	solid	468	lb/cu ft	468	lb/cu ft
heat of fusion		126	Btu/lb	126	Btu/lb
solidification range		2770-2670 F		2280-2080 F	

Acknowledgment

The author is indebted to the members of the AFS Heat Transfer Committee for valuable suggestions and comments made while the study was in progress and upon reviewing the reports. Thanks are due, also, to J. W. Hlinka and P. Calcaterra, coworkers in the Heat and Mass Flow Analyzer Laboratory, who carried out the details of the work and discussed many aspects of the paper.

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PRINCIPLES OF CORE BLOWING

By

A. M. Clark*

Too often we stress the needs of the high production foundries without giving enough thought to the smaller shops. In this paper the author hopes to cover core blowing for both the small and large foundry, along with discussing new developments, both in the Ford foundries and the rest of the industry, in which a wide variety of techniques are used for blowing cores.

A review of the basic principles of core blowing is needed for those who are currently not too close to the core room. To start, let's take a look at some early history, and a few types of core blowing machines.

The casting of cats in bronze (Fig. 1) was accomplished during the Seventh Century B.C. When this

*Ford Motor Co., Dearborn, Mich.



Fig. 1 — Asphalt core revealed in ancient bronze casting.



Fig. 3 — Large lion (16 ft long and 20 ft high) cast in China about 953 A.D. still stands in a monastery yard.



Fig. 2 — Oldest known iron stove required use of cores.

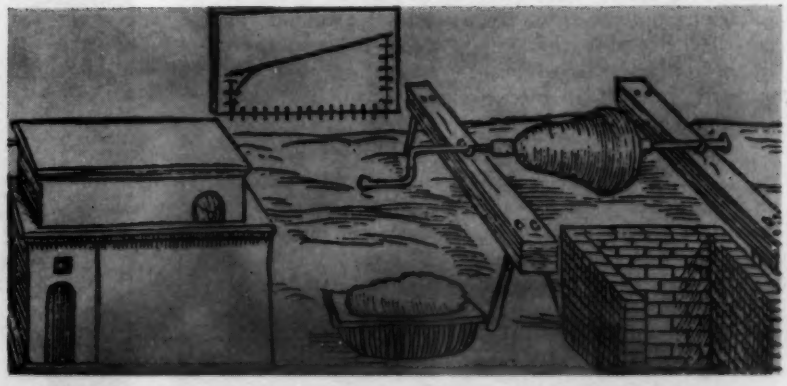


Fig. 4 — This crude equipment was used to form cores in church bells.



Fig. 5 — This view shows a typical core room in a shop about 1880.

casting was broken open, a permanent asphalt core was revealed. Apparently foundrymen had their shake-out problems in the early days too.

Figure 2 is a view of the oldest known cast iron stove made around 206 B.C. Thus, we see that the earliest castings were made by using cores. The author doubts if the tolerance of this core was held to plus or minus 0.010 in.

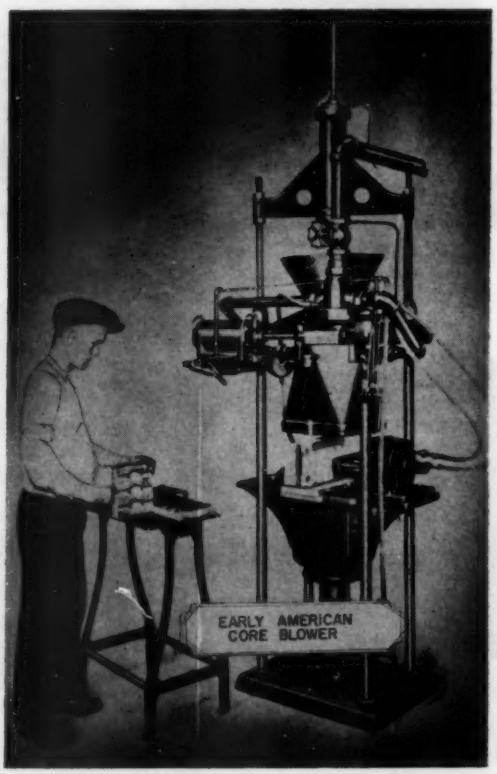
Figure 3 shows a large lion cast in China about 953 A.D. It still stands in the yard of one of their monasteries. The lion measures 16 feet in length and 20 feet in height. A casting of this size was a great accomplishment in the early ages.

Figure 4 shows the crude method used to form an early bell core. Casting of bells exerted quite an influence on the foundry industry. Demand for church bells of various sizes raised the casting of metal to a practical art, and so it was the church that gave the foundry an incentive that encouraged early developments.

It is interesting to note that the "Lost Wax Process," which is investment or destructible pattern molding, was used by the early Egyptians. This principle has retained its importance for more than 3500 years. In fact, this process, as used today, has contributed to the superior performance of our jet aircraft.

It wasn't until the Nineteenth Century that core making became standardized and equipment was made available for mass production in American shops. Figure 5 depicts a common core room scene around 1880. The cores were dried in kilns and

Fig. 6 — Early core blower.



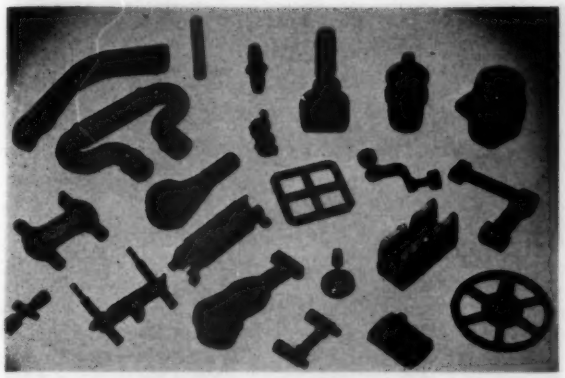


Fig. 7 — Bench blowers produce a variety of intricate cores.

then placed on racks to be carried to the foundry.

Core blowing machines were introduced into the core room about 45 years ago. Figure 6 shows an American blower developed around the turn of the century.

Early progress was slow and kept highly secretive. A coremaker was a craftsman and to protect his trade would not reveal any new developments or operating techniques.

Core making practice in this country has improved rapidly in recent years, in response to the demands on the foundry for quality, lighter casting weights, and closer casting dimensions. The material suppliers, manufacturers, technicians, and foundrymen alike have contributed to this progress. However, there is still a great need for improvement. Metallurgical advancements have made it imperative that we further reduce the dimensional tolerances of cores. Modern casting designs demand more intricate and improved coring, to reduce the end cost for today's competitive market.

An example of the savings that can be realized by using cores is the number of gun and tank parts that were cast during World War II for the United States Army Ordnance Department. Prior to the war, many of these particular parts were machined from solid blocks of steel. By changing the metal specifications, and using cores, these parts were cast at about half the previous weight and the machining time was reduced by more than 50 per cent. This is not an extraordinary example. The foundry industry is constantly striving to accomplish similar improvements.

For the purpose of discussing core blowers, we have arbitrarily classified the machines into two basic groups. They are the small bench blowers and the large floor models. In this discussion, we will consider small machines as bench blowers, even though they may be mounted on bases and could be considered floor models.

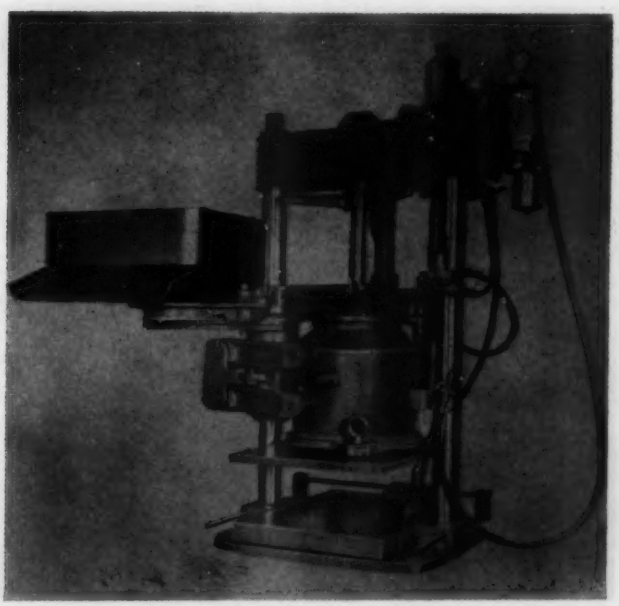
Core rooms having low production jobs will find small machines more efficient and economical. Because of high initial investment, it is not practical to have a costly, large machine idle during numerous job changes.

Figure 7 shows a few examples of cores blown by a small bench blower today. It is reported that the bench blower was introduced in the early 1940's. Because of the simplicity of design, core box rigging and high comparative productivity, it was readly accepted by many of the more progressive foundries with production schedules that warranted the use of these machines.



Fig. 8 (Left) — Cartridge type blower.

Fig. 9 (Below) — Sand magazine swings from blowing to filling point.



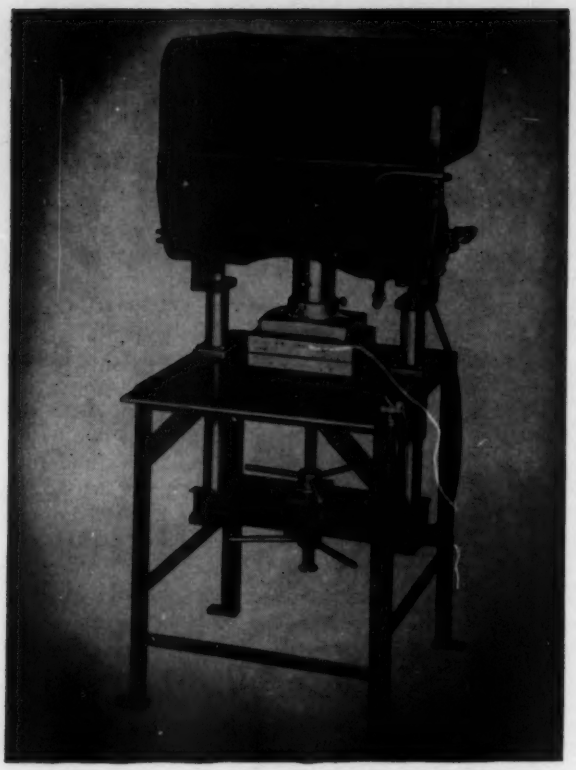


Fig. 10 — Sand magazine is stationary on bench blower.

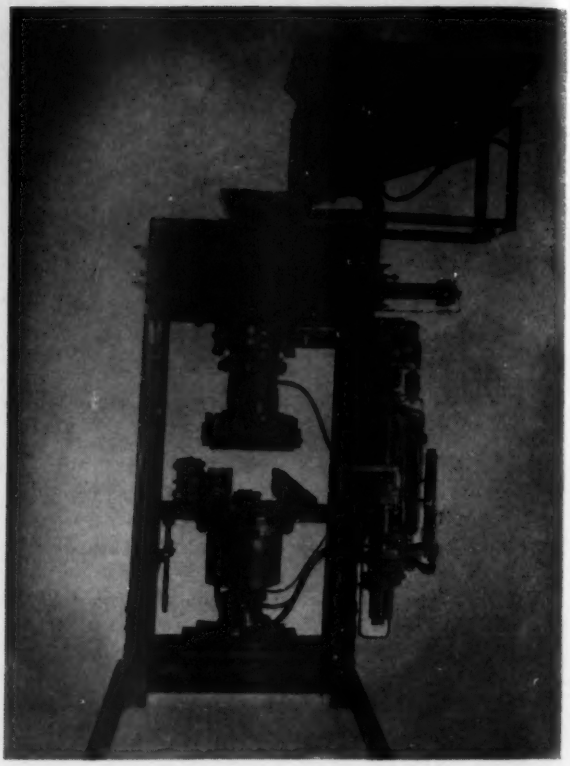


Fig. 11 — Floor model has stationary sand magazine.

Figure 8 illustrates a cartridge-type bench blower. The cartridge, commonly referred to as a sand magazine, is considered a part of the core box equipment. However, one cartridge may be used for several boxes of approximately the same size. The tubular part of the cartridge must be high enough to hold sufficient sand to fill the core box. The cartridge is filled by hand. Then the core box and cartridge are placed in the machine for blowing. The right handle clamps the box. The left handle blows the core.

Figure 9 illustrates a swing-type bench blower. The sand magazine swings from the blowing to filling position, either mechanically or manually, according to the manufacturer's design. On some models, the sand magazine swings in the direction of the operator, while others swing to the rear.

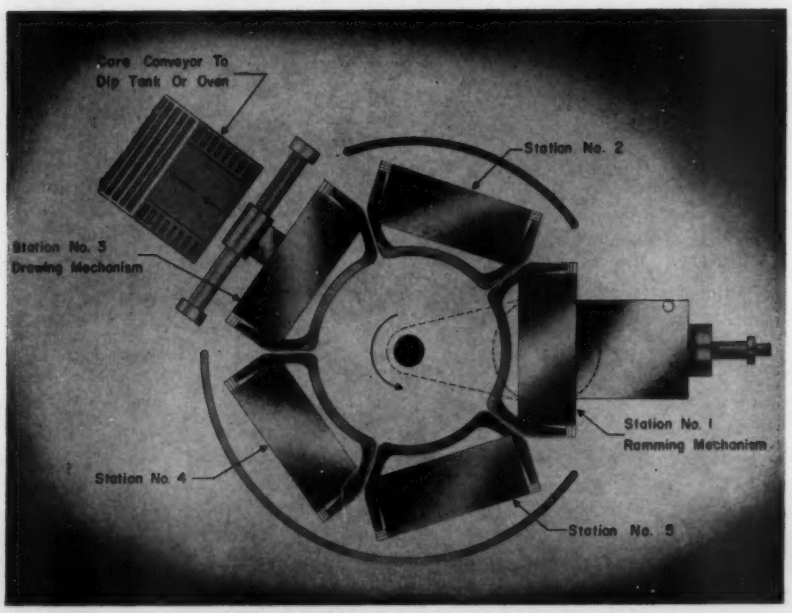


Fig. 12 — This machine has 5 stations for core production.

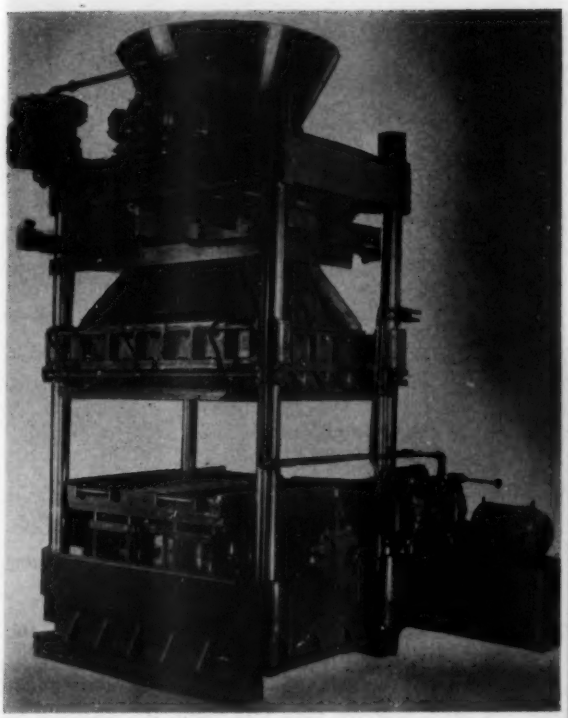


Fig. 13 — World's largest core blower.

Figure 10 shows a bench blower which has a stationary sand magazine. This eliminates the time and effort of moving the magazine from the filling to the blowing position. The table is adjustable for different height core boxes. Actuating one handle puts the machine through a complete blowing cycle.

Having very briefly reviewed several types of bench blowers, let's take a look at large *floor model blowers,* which have the advantage of lending themselves more readily to automation.

Figure 11 illustrates what is considered an average size machine, with a stationary sand magazine, automatic controls and sand feeder. The advantage of this type machine is that the sand magazine does not reciprocate. When the top half of the box is mounted on the blow plate, the line up of the upper and lower halves of the core box can be kept very accurate, thus assuring successful drawing of intricate cores.

Some modern machines can be run continuously when equipped with individual stations for blowing, rollover and draw mechanisms. It is necessary to stop the machine only to change boxes.

Figure 12 is an example of a high production operation. If all stations on the machine are rigged for the same core, only one upper core box half is needed. This is mounted on the blow plate while the lower halves are mounted on each of the stations. If the machine is rigged for similar cores, then both upper and lower core boxes are placed at *each* individual station. After each phase of the operation, the table advances one position. At station one, the core is automatically blown and the box unclamped as vibrators operate against it to free the core. The lower half of the box and core are lowered onto the indexing table. Next, the table moves up to station two where a dryer is placed on the core. When the table indexes to station three, the dryer is clamped to the box, both are rotated 180 degrees and then vibrated as the core is drawn from the box and left resting on the dryer. The box is then rotated back to its original position. Stations four and five are considered make ready stations, where loose sand is automatically blown from the box and a parting

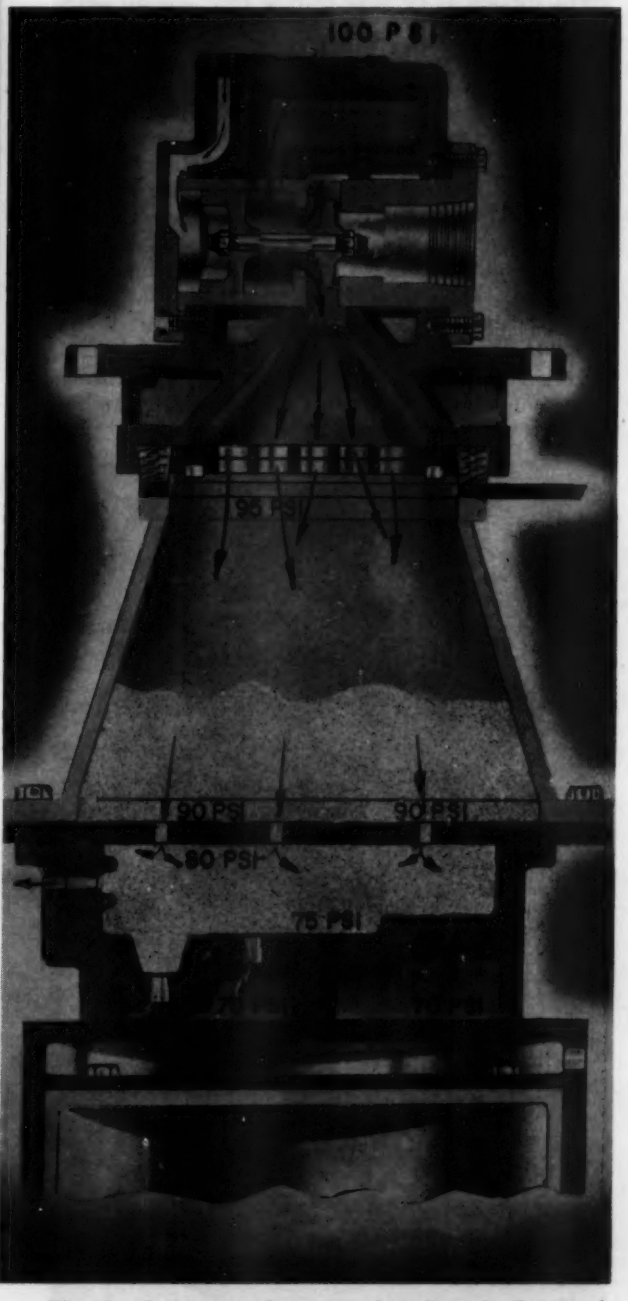


Fig. 14 — Satisfactory air pressure drop shown in diagram of blower.

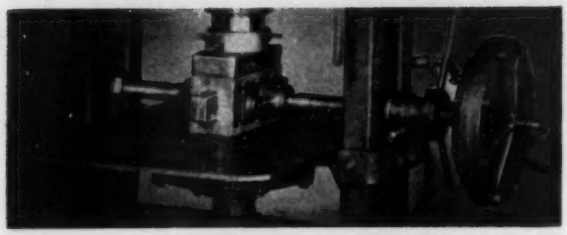


Fig. 15 — This automatic clamping device also positions the core box and lines up the blow holes.

agent applied along with placing core support wires if needed.

Figure 13 shows a machine which is believed to be the world's largest core blower. The sand magazine holds approximately 3000 lb of sand. There are two diaphragm type blow valves, each having a 4-in. diameter opening and the airline feeding the machine is 6 in. in diameter. The hydraulic clamping pressure is close to 75 tons. Two of these machines were used to blow cores at the magnesium foundry of the Ford aircraft plant in Chicago. The blower is designed to blow cores up to a maximum weight of 1500 lb. In the Ford plant, however, the largest core blown weighed 375 lb.

Having covered several types and sizes of machines as a complete unit, we want to discuss some of the important components of the core blower and the role they play in blowing cores.



Fig. 17 — Mechanical clamp raises core box vertically, sealing blow head against magazine and box.

The basic principle of core blowing consists of floating sand into the core box using the medium of compressed air. In our present day machines, the velocity of the air must be kept high, to obtain the desired velocity of the sand particles, thus insuring their deposit in the remote corners of the core box. It follows, therefore, that the air pressure within the box must be relieved by adequate venting.

Figure 14 shows the pressure drop from 100 lb at the inlet to 95 lb at the top of the sand magazine and 90 lb at the bottom. The air pressure in the core box shows a well vented condition.

Most core blowing machines are served by an air-line carrying 95 to 125 lb of air at the machine. Some cores can be blown at 80 lb pressure, but as a general practice, it is recommended that a pressure of 90 lb be used.

A sufficiently large air inlet to the machine should be provided so that when more and larger blow holes are used, the pressure drop from the blow valve to the core box is held to a minimum. A 2½-in. air inlet is recommended for the smaller size blower and a 4-in. inlet on the larger size.

Air must be introduced as quickly as possible to minimize the effects of channeling. This requires a

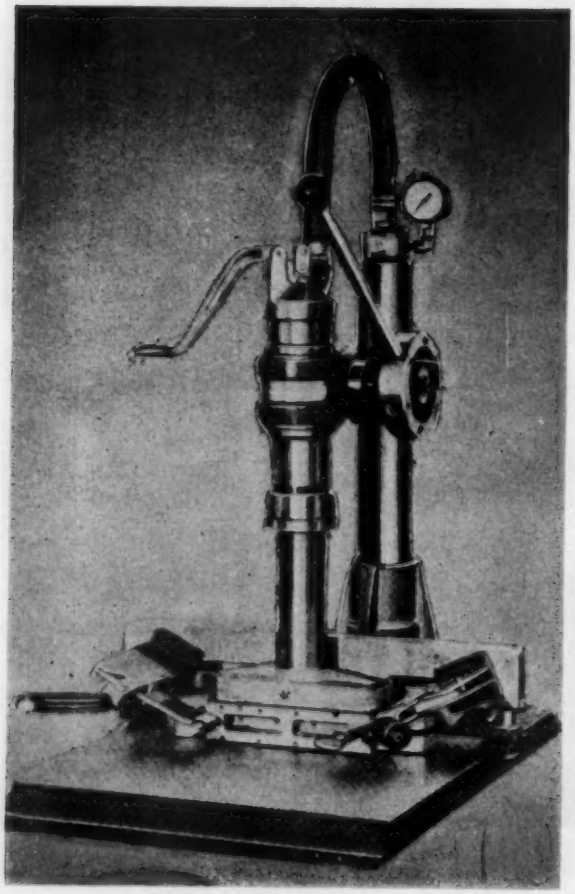


Fig. 16 (Left) — Manually operated clamps are used for low production jobs.

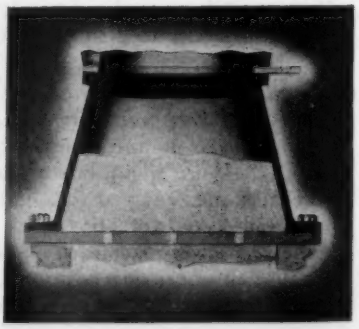


Fig. 18 (Right) — Typical magazine for sand.

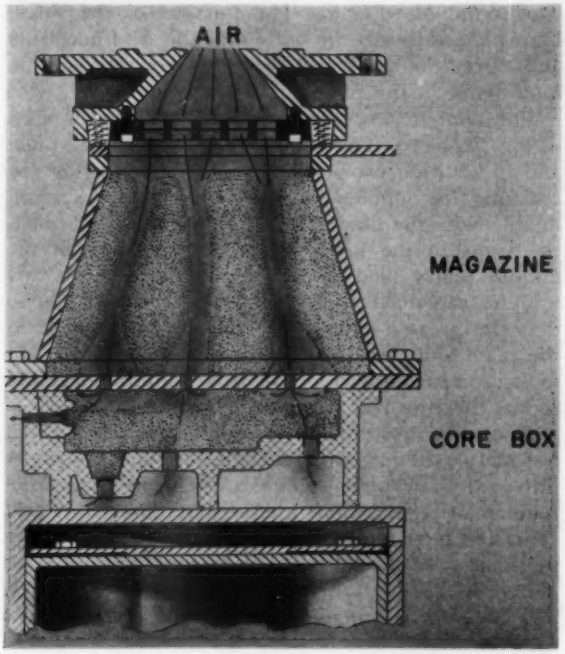


Fig. 19 — Diagram shows how air channels through sand above blow holes preventing good coreblowing.

fast acting inlet valve and an air receiver for the machine, or if there are several machines on a common airline, the engineer must be careful to use an air header of sufficient size to prevent line pressure drops.

Air is usually introduced into the machine by a poppet valve, slide type valve, or diaphragm.

To resist the high pressure of air required to blow a core, we find it necessary to hold the box in the

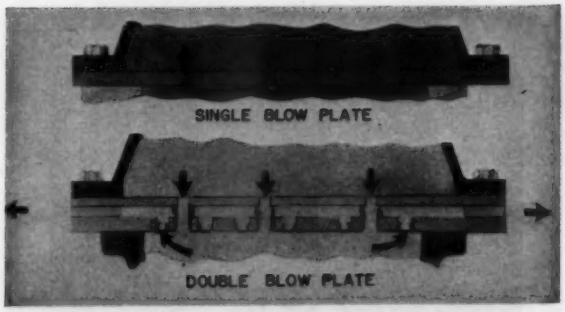


Fig. 22 — The single blow plate seals the sand magazine and directs the sand into core box. The double plate provides more venting.

machine by either an automatic or manual clamp. The clamp shown in Fig. 15 is used for automatically positioning the core box horizontally. It keeps vertically-split core boxes closed during the blowing sequence and lines up the blow holes between the blow plate and the core box.

The manually operated clamps shown in Fig. 16 are used for low production jobs only. Another common clamp is the use of table-mounted guides which position the box so the blow holes are lined up properly. Guides of this type can be utilized when the boxes of relatively the same size and shape are continually run on the same machine.

The mechanical clamp shown in Fig. 17 raises the core box in a vertical squeeze, thus forming a seal between the blow head of the machine and the magazine, also between the blow plate and the core box itself. It is important that the clamping pressure be greater than the blowing pressure, in order to avoid blowing out at the parting.

The next component that serves a very necessary purpose is the sand magazine. Figure 18 represents

Fig. 20 — Introducing air through screened inner liner reduces channeling by agitation in magazine.

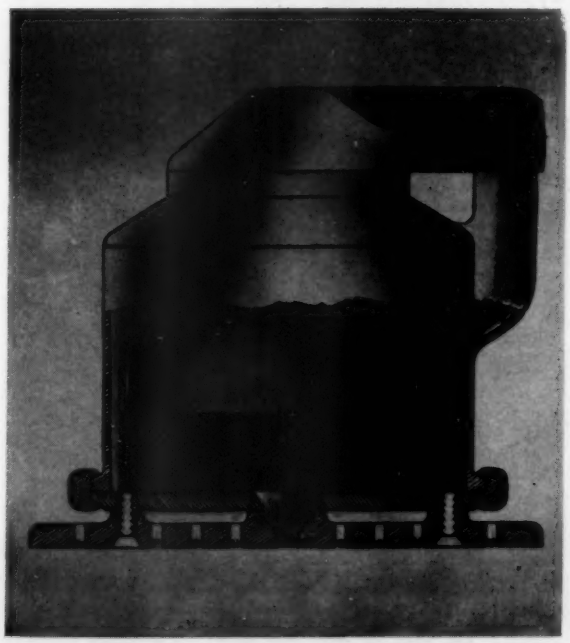


Fig. 21 — Continuously rotating paddles may be used to break up channels and keep blow plate covered.

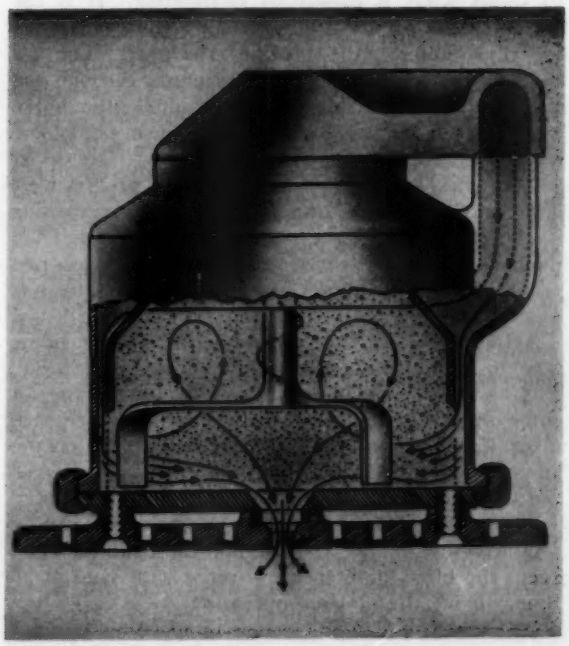




Fig. 23 — Universal plate permits versatile arrangement of tubes, vents.

a plain sand chamber which manufacturers can supply in various sizes and shapes, to accommodate a variety of applications. In our opinion, a magazine should be selected which has a bottom opening equal to, or slightly larger than the core box to be filled. We highly recommend that the manufacturers' specifications be adhered to in determining the magazine size and shape. When trouble arises in blowing a new job, the magazine capacity is frequently overlooked as a source of the trouble.

Under certain conditions, sand adheres to the inside walls of a magazine. When this happens, it is necessary to rap on the chamber to move the sand. Plating or an application of plastic has been used with some success to eliminate this problem.

One of the major difficulties in core blowing has been the channeling of sand in the magazine. Figure 19 shows what is meant by channeling. It is the author's opinion that tunnels start to form above the blow holes while blowing the core. The air entering the top of the magazine passes through the sand, and as the air approaches the small blow holes, its velocity is increased sufficiently to tear sand loose from the mass and carry it into the box. The cavity left in the sand magazine becomes progressively larger until it fails to support the sand above, causing it to drop and lower the sand in the chamber. Then the channeling starts over again. This condition prevents blowing good cores and causes excess core box wear.

Agitation within the sand magazine may be necessary to prevent channeling in some applications. The agitation may be performed by having the air enter the magazine through a slotted or screened inner liner, shown in Fig. 20. This method creates an aerating air circuit that conditions and carries the sand through the blow holes into the core box.

As shown in Fig. 21, continuous agitation can be produced by rotary type mechanical paddles. This keeps the blow plate evenly covered with sand at all times during the blowing operation.

The next item in the review of machine components is the blow plate. The purpose of the single blow plate is to seal the bottom of the sand magazine and direct the sand into the core box through the blow holes in the plate.

The double blow plate, shown in the lower view of Fig. 22, serves the same purpose as the single plate and in addition provides additional venting, thereby reducing the number of core box vents to a minimum. Note the flow of the air passing out through the vents in the lower plate, on out between the double plates held apart by spacers. Almost any foundry can make their own vented blow plates, so that the greatest amount of venting takes place in the plate and not in the body of the core box. On high production machines, it is common to have the top of the core box mounted to the blow plate.

Figure 23 illustrates a universal blow plate which is widely used in small foundries and job shops. A universal blow plate can take the place of a large number of regular blow plates. The blow tubes and vents are interchangeable. Thus, by merely moving the blow tubes to the positions needed, the universal plate can be used with many different core boxes. The holes in the plate, which do not require blow tubes, are filled with vents.

A very important feature in the construction of good core blowing equipment is the size and placing of the blow holes. There are several schools of thought on this subject. Some authorities claim that many small holes produce better results than a few large holes. Others claim the holes should be as large as possible without dumping the sand. Those favoring the small holes believe the sand is sprayed at a greater velocity, getting into difficult corners and packing much better. As shown in Fig. 24, it is not necessary to use a round blow hole. Slotted holes have been used with success in many applications.

The cross section in Fig. 25 shows the blow tube. The use of a tube is important because it can be replaced when worn out. Without the tube, the hole in the blow plate would be worn by the sand blasting effect of each blow, and the plate would have to be replaced. It is obviously more economical to replace

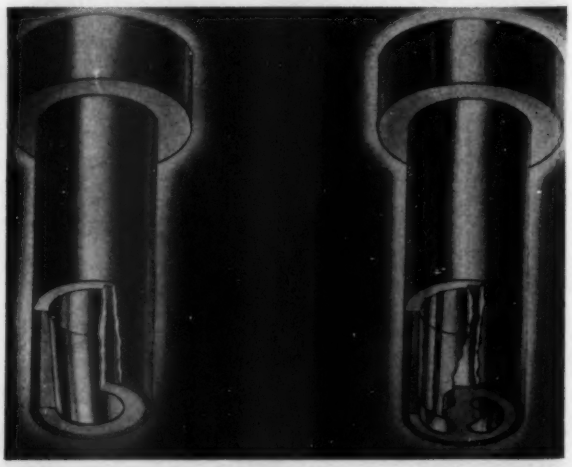


Fig. 24 — Blow holes may be round or slotted. They may be as large as possible without dumping the sand or small so as to spray sand.

tubes than blow plates. Some bushings are lined with rubber or plastic or equipped with an adaptable sleeve. Others use a simple short piece of steel or brass tubing. When properly installed, they also serve the purpose of providing an air-tight connection between the blow plate or vent plate and the core box. The bushing should break the sand cleanly and evenly at the core line to save extra cleaning labor. This is done by having the bottom of the blow tube taper off to the core line by means of a rubber or steel insert as shown in these cut-away sections. Another method is the use of a straight tube with the bottom section tapped. The threads then help to break the sand cleanly.

One manufacturer of core blowers recommends a ratio of six core box vents for each blow hole of equal size. However, it is seldom possible to find room for that many vents in the core box. Most cores are blown successfully with a minimum ratio of three vents for each blow hole of equal size. In very small cores, the ventilation at the parting is sometimes sufficient to permit the blowing of the core without additional venting. The same manufacturer has found that at approximately 3/8-in. diameter the comparative effectiveness of screen-type and slotted vents is approximately equal. Below 3/8-in. diameter the slotted vents are most effective. Above 3/8-in diameter, the screen-type vents become more effective.

It is interesting to note the 150-mesh screen vent shown on the lower right of Fig. 26. This screen was developed for use with extremely fine sands used in the "D" process. Foundrymen have found this screen stays exceptionally clean during several hours' use.

Another disadvantage in the use of screen vents is the tendency of the screen to blow down and leave a slight lump on the core. The slotted vents shown on the left hold their shape better and last longer, but are more difficult to keep clean. Good practice suggests the use of screen vents on print sections of the core. However, they may be used in other areas where the lump will not affect the quality of the core or casting.

Having covered some of the important components of the core blower, the author will mention a major disadvantage of conventional core blowing.

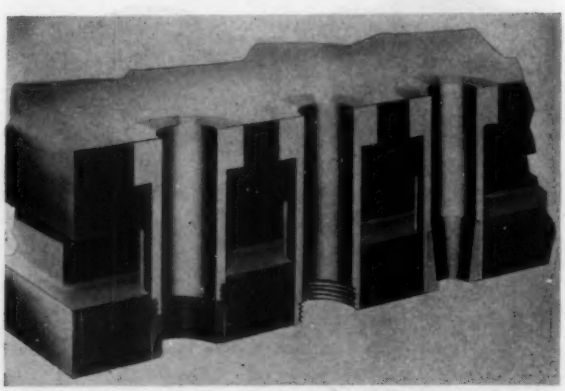


Fig. 25 — Cross section showing three different blow tubes designed to break and clean at core line.

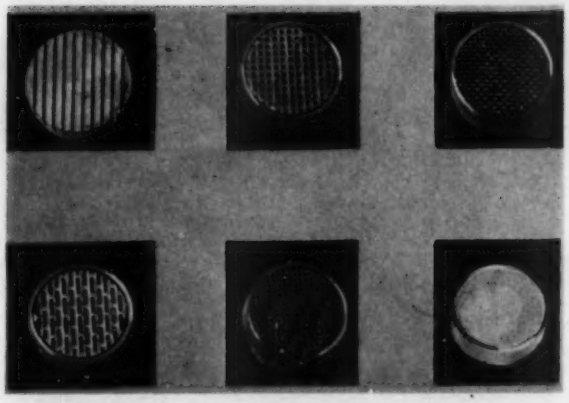


Fig. 26 — Vents may be designed in a variety of ways. Note the 150 mesh screen vent in lower right.

Our present core blowers propel, or blast the sand into the box. Satisfactory results can be obtained when each grain of sand is completely surrounded by air. A close examination of this principle concludes that it is similar to sand blasting. The core boxes are consequently subjected to severe wear. The very nature of this operation dictates the use of an abrasive resistant core box, costly to build and maintain.

Core Shooting Principle Explained

To minimize this condition, new principles of core blowing are being developed. The shooting principle that was recently introduced in Europe may be an answer. This operation uses only a small, measured amount of compressed air in the sand magazine. Due to the large size blow valve the air is applied instantaneously, and the sand is extruded into the box. In other words, the sand is sufficiently accelerated in a body so that it fills the box completely and uniformly. Theoretically, this should result in even strength values throughout the core and render uniform permeability.

There has been considerable interest in this principle of core blowing. In fact a few American manufacturers have incorporated some of the shoot principles in their newer machines.

The new principle will be reviewed, but the following discussion is not to be construed as a recommendation, favorable to any particular machine.

The reported advantages of the shoot machine, as compared to the conventional core blowers can be summarized as follows:

Since the air supply valve (Fig. 27) is closed during the blowing cycle, only the air from the reservoir expands on the column of sand. This principle requires an extremely fast and very large air opening, so that in a fraction of a second, the entire force of the air from the reservoir works on the sand to accelerate it very rapidly into the core box. Thus, the consumption of compressed air is at a minimum. A reduction in the amount of air required to blow cores is very important to reduce core room costs. The sand magazine is provided with a slotted cylinder sleeve, thus the air from the shoot valve acts through the holes in the sleeve reducing the friction of sand flowing down the cylinder.

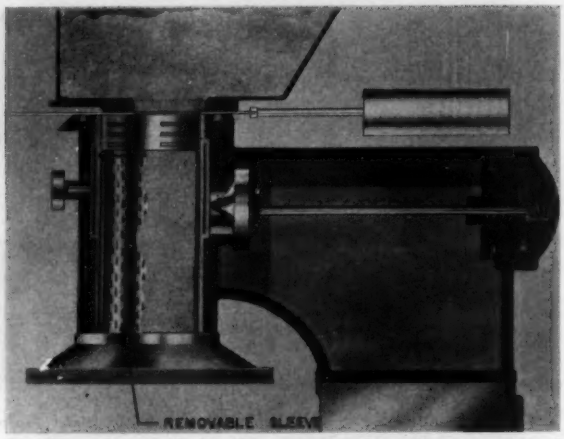


Fig. 27 — A new design for core shooting offers promise of many advantages over conventional ways. The shoot machine uses less air, core box wear is hardly measurable, venting problems are less.

A reported wide range of core sand mixtures can be used.

Core box wear is claimed to be hardly measurable. In fact, it is reported that wood core boxes are used in a number of applications.

Further comments are that the core box is not under airline pressure. Thus, venting, which can cause serious difficulty in conventional blowing, is considered very simple when using the shoot principle. In fact, only the atmospheric air present in the core box must be vented. This is often accomplished by venting the blow plate, or drilling a hole in the trouble spots of the core box to allow the air to escape.

It is the author's opinion that these principles are fundamentally sound and should be incorporated in our future core blowers.

While discussing new developments, the author wants to mention some of the relatively new techniques that are being used at the Ford foundries. We have reached some degree of success in blowing several large cores for the cylinder block, cylinder head and intake manifold by using resin-bonded sand. See Fig. 28.

The use of resin-bonded sand for blowing shell cores and shell molds certainly has its place in the foundry industry. In the author's opinion, further cost reduction of resins is necessary to accentuate the competitive advantage of shell cores and molds.

Based on the work we have done with the CO_2 process, we feel it constitutes an additional challenge to blowing resin-bonded sand. Sodium silicate bonding of sands is not new by any means, having been used in the manufacturing of grinding wheels, lost wax process, and as a clay deflocculant.

Phenolic resin and sodium silicate bonding of sand are just two of the recent developments in the foundry industry. In order to maintain our competitive standing in modern industry, we need definite and specific objectives. In our opinion, these are the principle objectives of core blowing engineering.

We must work toward a *reduction* in:

Skilled Core Makers
Physical Effort
Initial Investment
And Maintenance Costs.

We must work toward an *improvement* in:

Dimensional Control
And flexibility of Equipment

And finally, we must work toward an *increase* in production per man hour.

In order to make longer and more rapid strides toward these objectives, more help must be forthcoming from our machine manufacturers. They will have to develop machines and techniques that will make possible the blowing of more intricate cores, and the blowing of cores in one piece that are now being assembled.

Considerable emphasis has been placed on building shell making machines, while comparatively little has been done in developing the field of blowing shell cores or blowing shell molds. In addition, much more could be done toward expanding the CO_2 process.

In too many cases, the research in developing new machines has been left up to the foundries. There is a very lucrative field for the manufacturer who will develop the core blowing machines needed to further our objectives. The rapidity of these advancements will hasten the reduction in casting costs. As foundrymen, we are responsible for producing good castings at the lowest possible cost, not engineering machines. Foundrymen, however, must be ready and willing to accept the new ideas. New designs will not help if they are not accepted in a spirit of cooperation.

A further word of caution to foundrymen, the more technical and complicated the manufacturers make core blowing machines in order to gain the foregoing objectives, the more skilled the foundry maintenance people will have to be.

Technological Improvements Needed in Coremaking

The machine manufacturers have the know-how to develop new techniques. Without limiting themselves exclusively to the following fields, the author would like to see them direct their major efforts toward the realization of these technological improvements:

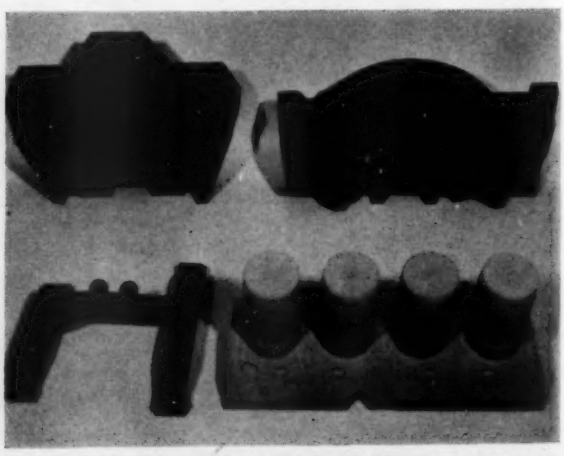


Fig. 28 — Cores for cylinder blocks, cylinder heads and manifolds were blown with resin-bonded sand.

Investigate the diaphragm principle of core making, and come up with positive recommendations as to its applications, or the practicability of eliminating completely the use of air.

Look into the possibility of exerting a negative pressure on the core box during the blow cycle. Conceivably, a better core would result. In any event, constant improvement of venting techniques should be strived for.

Put increased efforts into the development of the techniques of blowing hollow cores.

Further the CO_2, or any other process, to eliminate baking, making it possible to position the core machines closer to the molding line, thereby eliminating costly core handling systems.

Eliminate the noise problem of exhaust valves and vibrators connected with the core blowers.

Extend every effort to develop new ways and means of minimizing manual labor and fatigue.

Obviously, any new developments must encompass ease of maintenance and good safety practice.

Summary

In this paper we have reviewed some of the early history of core making, illustrating the efforts of our predecessors to make better, lighter and more accurate castings. We have also taken a look at the many sizes and types of present day core blowers. We have reviewed the methods of core blowing and discussed some of the machine components. We have mentioned the pros and cons of conventional blowing and some of the steps that are being taken by the manufacturers to eliminate present day problems. Most important of all, however, we have reviewed the principle objectives of the foundries, the machine manufacturers, and the engineers, in searching the new fields to be developed. It is the author's personal opinion that if sand castings are to maintain their competitive position in modern industry, much greater emphasis must be placed on research and technological improvement.

The seriousness of this situation is brought out by the fact that permanent-mold and die-casting processes dominated the light metal foundry operations last year. Aluminum die casting shipments increased 45 per cent while the gain in sand casting was a little over 8 per cent.

Acknowledgment

In gathering information on the types of present day core blowers, and the various methods of blowing, the AFS Sand Division had several meetings with various manufacturers' representatives. We received some very valuable assistance from them at meetings, from their letters, sketches, and drawings. To each and every manufacturer who contributed toward making this presentation, we give wholehearted thanks. It was certainly gratifying to see the major core machine manufacturers engaging in an exchange of information, even though there were many controversial opinions expressed.

NOISE REDUCTION IN THE MANUFACTURE OF MARINE PROPELLERS THROUGH THE USE OF VIBRATION DAMPING

By

James H. Botsford*

A noise problem existed in a shop where large marine propellers are manufactured and repaired. In the manufacturing process, it is necessary to remove metal from the castings in cleaning, shaping and balancing operations. Most of the metal is removed by chipping. In the case of a large propeller, several thousand pounds of metal must be removed altogether which requires several hundred hours of chipping. Many of the propellers are made of an alloy similar to bell bronze and produce a loud ringing sound during chipping.

A view of the propeller shop is shown in Fig. 1. The range of sizes and designs of propellers manufactured is indicated by the assortment of propellers visible. The shop is necessarily an open area as shown, to permit the propellers to be moved about freely. In consequence of this the noise produced by chipping on any one of the propellers spreads throughout the shop.

Several plans to reduce the noise at propeller chipping operations were tried. Wooden blocks and props were arranged to support the blades as firmly as possible. Steel weights and sand bags were used to weight down the blades. Each of these measures reduced the vibration of the blades somewhat and the noise, which was generated by blade vibration, was reduced proportionately. Small sheds for housing a single propeller during chipping were constructed to isolate the noise; some were lined with sound absorbing material. A row of the chipping sheds is shown in Fig. 2.

The propellers to be chipped are moved out of the propeller shop on small rail cars and transferred by crane to similar cars which carry them into the chipping sheds. A barrier is then placed in front of the shed to confine the chipping noise to the shed. These approaches will not be discussed further except to say that no combination of them gave satisfactory noise reduction and each introduced new problems. The discussion will be limited to the vibration dampers which were developed and with which it is practical to reduce the ring of the propeller during chipping well below the noise of the chipping tool alone.

To many, the term vibration damping means the prevention of transmission of vibration from one place to another, as when a vibrating machine is placed on resilient springs or rubber mounts to reduce the vibration transmitted to the foundation.

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(Left)
Fig. 1 — View of propellor shop.



(Right)
Fig. 2 — View of chipping sheds.



(Left)
Fig. 3 — Vibration damping device.



(Right)
Fig. 4 — Vibration damping clamps on propeller blade.

However, the vibration of the machine itself is not reduced when placed on resilient supports, but is actually greater than would be the case if the machine were attached firmly to the foundation.

When the term damping is used properly, it refers to the reduction of vibration through the removal or absorption of the motional energy which is accomplished by the introduction of friction into the vibrating system in some way. With the vibration opposed by friction, the motional energy is quickly converted to heat and the vibration dissipated. A familiar example is the use of undercoating in automobiles which serves principally to damp the vibration of the thin metal panels.

The distinction between vibration isolation and vibration damping becomes clear when an automobile suspension system is considered. Springs are used to isolate the vibration of the wheels and axles from the frame and body. If springs only were used, however, the car would bounce incessantly and make driving difficult. To prevent this, shock absorbers are installed to present frictional resistance to the bounding and to damp out these vibrations. In the case of the shock absorber, the frictional resistance is obtained by forcing a viscous liquid to flow through a stricture.

In the propeller shop, production requirements imposed certain restrictions on the vibration damping arrangements which could be used. The damping system had to be easily applied and removed leaving the surface of the propeller clean and unmarred. Moreover, the damping system could not make large areas of the propeller inaccessible to chippers so that much time would be lost in moving the dampers as chipping progressed. It was also desirable that it be possible to attach the system to the propeller in such a way that there would be no danger of it falling off if the propeller were turned over or moved to a new position. These requirements were met by the system shown in Fig. 3.

The damper of Fig. 3 consists of two 6-in. steel disks ½ in. thick, one of which is welded to a bracket containing a screw to form a clamp having large bearing faces. The damping material is affixed to the inner faces of the steel disks and is pressed against the propeller surface when the clamp is attached to the blade. When the blade vibrates, the heavy clamp does not follow the vibration of the blade because of its inertia and the resilience of the damping material, in this case felt carpet padding.

This differential motion of the clamp and blade causes a great deal of flexure and compression in the felt, and the resultant rubbing of the fibers converts the vibrational energy to heat. The use of the clamp assures greater flexure of the damping material than would be obtained if the felt alone were laid on the blade and as a result increases greatly the damping which can be obtained with a given amount of damping material. Also the clamp arrangement fulfills the special requirements for a damping system mentioned above.

Actually, the felt shown in Fig. 3 provided insufficient damping and little noise reduction was obtained. Many other materials were tried and best results were obtained with an asphalt impregnated felt normally used as a damping material in automobile bodies, metal furniture, etc. This material was Type 401-62-lb industrial sheathing felt adhered with Type B 1340 asphalt adhesive.

Fifty dampers were made of an improved design using this material; several of these are shown attached to a blade being chipped in Fig. 4. Two steel plates ½ in. thick and 6 in. square were welded to the jaws of a 6-in. C-clamp. The square plates were simpler to make than the round ones and the use of the C-clamp produced a one-piece structure which is easy to handle.

In order to evaluate the effectiveness of the vibration dampers in reducing the noise produced by the propeller during chipping, a series of noise measurements were made with a chipper working steadily at the tip of one blade. Since the changes in noise produced by the propeller vibration were of interest, the microphone was placed within a few inches of the propeller surface at parts of the propeller well

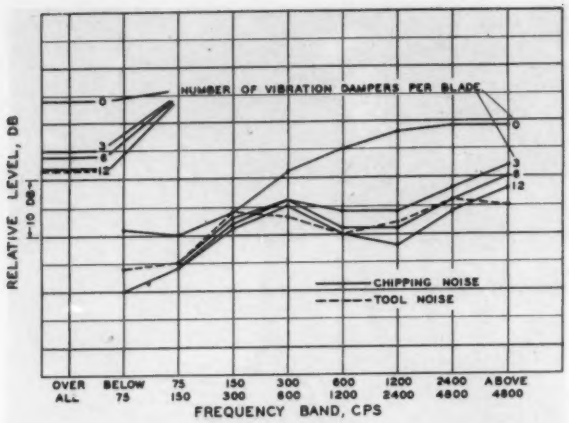


Fig. 5 — Noise levels at destroyer propeller.

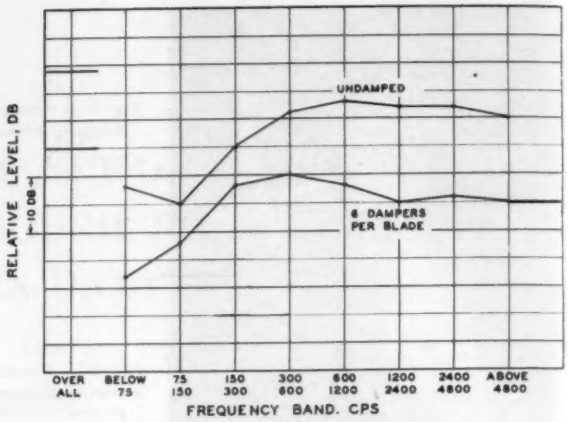


Fig. 6 — Noise levels at tanker propeller.

away from the chipper where the noise produced by the vibrating propeller was picked up preferentially to the noise generated by the chipping tool itself.

The curves of Fig. 5 show the octave band levels with one man chipping on a four-blade bronze destroyer propeller weighing approximately 25,000 lb, under various conditions. They were obtained by averaging the measurements at three positions. The upper curve shows the noise levels produced by chipping on the undamped propeller and the lower solid curves show the levels with three, six and twelve dampers attached to each blade.

The noise levels produced when the chipping tool was operated without touching the blade are shown by the dotted curve. The tool noise exceeded the chipping noise with six to twelve dampers per blade in some octave bands because the tool made more noise when running free than when used to remove metal. Note that the chipping noise decreased as more dampers were used and was about the same as tool noise with six dampers per blade.

Equally good results were obtained on other types of propellers. Results of noise measurements at a five-blade nickel-aluminum-bronze super tanker propeller weighing 53,000 lb are shown in Fig. 6. Again the curves are the averages of measurements at three positions. The upper curve shows the noise levels produced by chipping on the undamped propeller; the lower curve shows the levels with six dampers attached to each blade. The noise reduction shown is significant but greater reduction could be obtained through further damping of this propeller since the level of tool noise was not reached.

Through the use of vibration dampers, it was possible to reduce the noise at propeller chipping to the point where the chipping tool itself is responsible for the noise which remains. Admittedly, the propellers described are unusual castings; but, theoretically, there is no reason why vibration damping cannot provide equivalent results on any casting or assembly which produces considerable noise during chipping. A simple test will reveal those structures which can be quieted through vibration damping. If striking the structure which is ready to be chipped produces a ringing sound afterward, then the noise produced by chipping on that casting can be reduced by vibration damping.

The longer a casting rings after being struck, the more noise it will produce during chipping and the greater the noise reduction which can be obtained through damping. An efficient damping system is necessary to reduce the ring during chipping below tool noise and considerable experimentation may be required to find the best method of obtaining such high damping in each case. When sufficient damping has been obtained, striking the casting will elicit only a dull thud.

There are probably many operations which can be quieted by vibration damping. Paradoxically, the very noisiest are likely to be the easiest to quiet. And what is more appealing, they may be discovered and the best means of damping developed with instrumentation limited only to a hammer.

EFFECT OF TITANIUM DEOXIDATION ON THE MECHANICAL PROPERTIES AND MICROSTRUCTURE OF CAST Cr-Mo-V STEEL

By

R. D. Ahles,* C. H. Weight,* and R. S. Zeno**

ABSTRACT

A study was made to determine the effect of titanium deoxidation on the mechanical properties, inclusion type and microstructure of cast Cr-Mo-V as a function of sulfur content. A range of titanium additions from 0 to 2 lb/ton and sulfur contents of 0.010 to 0.028 per cent was studied.

The base composition was the same for all castings.

The following conclusions are made:

1. The tensile ductility (R.A.) and quality factor (T.S. x R.A.) of cast Cr-Mo-V steel decreases with increasing sulfur and/or titanium content.

2. Cast Cr-Mo-V, titanium deoxidized with 2 lb/ton exhibited a marked resistance to tempering.

3. Cast Cr-Mo-V with a maximum sulfur content of 0.017 per cent, titanium deoxidized with up to 1 lb/ton resulted in good tensile properties (minimum quality factor of 5×10^6).

4. Localized areas of fine Type II sulfide inclusions were observed even in the lowest sulfur content titanium deoxidized material. These inclusions, however, were not present in sufficient quantity to materially affect tensile ductility.

Introduction

The authors have, as have other steel casting producers, been concerned with pinholing which commonly occurs in silicon-deoxidized low-alloy steel castings. It has long been known in the foundry industry that the use of stronger deoxidizers e.g. Al and Ti can eliminate or minimize the pinholing problem.

Of all the deoxidants available for turbine grade steels, titanium appeared to be one of the most promising since it is a strong carbide former and was reported by Houdremont et al[1] and observed by Zeno to increase the resistance to tempering.

After a few titanium-deoxidized basic electric heats of Cr-Mo-V steel were made it was found that the room temperature tensile ductility of the heat-treated steel did not consistently meet the specification requirements because of the formation of fine Type II sulfide inclusions. The sulfur level of the steels produced during this period ranged from 0.018 to 0.027 per cent.

This investigation was made to study the effect of titanium deoxidation on the mechanical properties, inclusion type and microstructure of cast Cr-Mo-V as a function of sulfur content. A range of titanium additions from 0 to 2 lb/ton and sulfur contents of 0.010 to 0.028 per cent was studied.

Procedure

Material

A 50,000-pound heat of 1 Cr, 1 Mo, ¼ V steel produced by the basic electric (two slag) process was used in this study. The charge consisted of half foundry returns and half steel scrap. After melt down the heat contained 0.28 per cent carbon and was ore boiled to the 0.17 per cent carbon level at which time it was blocked, the slag changed and final additions made. This heat was recarburized to 0.20 per cent after the block (one hour before tapping) with high carbon ferromanganese.

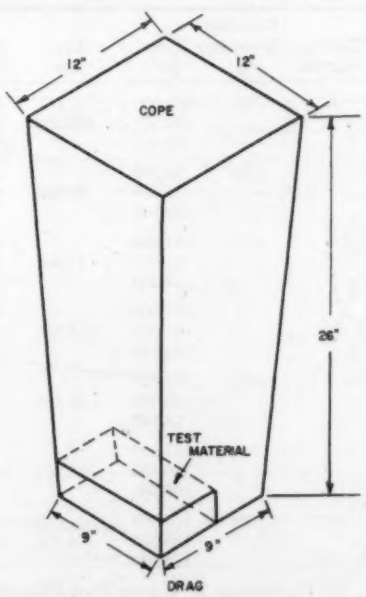


Fig. 1 — Test Casting

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TABLE 1 — SULFUR AND TITANIUM CONTENTS OF CASTINGS

Casting Number	Titanium Added %	Lb/Ton	Total Titanium, Wet*	Total Oxygen***	Sulfur**
6	Nil	0	.002	.0056	.011
7	.018	⅓	.016	.0065	.010
8	.037	⅔	.026		.011
9	.055	1	.019	.0051	.010
10	.110	2	.082	.0051	.010
16	Nil	0	.002		.013
17	.018	⅓	.014	.0046	.014
18	.037	⅔	.026		.014
19	.055	1	.038		.013
20	.110	2	.061		.013
26	Nil	0	.002		.017
27	.018	⅓	.016	.0054	.017
28	.037	⅔	.025		.018
29	.055	1	.035		.018
30	.110	2	.061	.0054	.017
36	Nil	0	.002	.0081	.024
37	.018	⅓	.015		.024
38	.037	⅔	.037		.024
39	.055	1	.038	.0053	.024
40	.110	2	.066		.024
46	Nil	0	.002		.027
47	.018	⅓	.010		.029
48	.037	⅔	.041		.028
49	.055	1	.056		.027
50	.110	2	.093		.027

*Titanium analysis double checked.

**Sulfur analysis triple checked.

***Total oxygen from vacuum fusion analysis.

To check for possible segregation, samples were taken from the top and center of each ingot for titanium and sulfur analyses. The titanium and sulfur content was uniform throughout each ingot.

The ladle analysis of this heat in weight per cent was as follows: C, 0.20; Si, 0.38; Mn, 0.67; S, 0.011; P, 0.011; Cr, 1.05; Ni, 0.15; Mo, 1.10; V, 0.17; and N, 0.008.

Twenty-five dry-sand mold castings (ingots) were poured from this heat each weighing approximately 900 pounds (Fig. 1). The remainder of the heat was used for a similar investigation of aluminum deoxidized Cr-Mo-V. Sulfur (FeS) additions to produce a range of 0.010 to 0.028 weight per cent were made in the ladle. Titanium additions ranging from 0 to 2 lb/ton were made in the mold. The titanium was added as 10 per cent iron - 45 per cent silicon - 45 per cent titanium.

Following the sulfur additions in the ladle, the metal was stirred with a steel rod to assure thorough mixing and uniformity of composition. The titanium and sulfur contents of the various ingots are summarized in Table 1.

Heat Treatment

All castings were initially given a heat treatment together consisting of a 1050 C (1922 F) austenitize for 10 hours, air cool, temper at 690 C (1274 F) for 20 hours followed by a furnace cool. Because of differences in the response to heat treatment it was necessary to redraw or retreat certain castings (Tables 2-6), in order to evaluate the material at a strength level of approximately 100,000 psi.

Tests

The following tests and analyses were made on sections taken from the same location in each casting (Fig. 1): Room temperature tensile, metallographic examination, total oxygen determinations, and sulfur and titanium determinations.

TABLE 2 — SUMMARY OF MECHANICAL PROPERTIES FOR GROUP I CASTINGS (0.010% SULFUR)

Casting Number	Residual Titanium, %	T.S., psi	Average	Y.S., .02% O.S., psi	Average	Elong. % in 2 in.	Average	R.A., %	Average*	Q.F.**	Average	Additional Heat Treatment
6	.002	107,500		81,250		23.0		60.0		6.46		
		106,500	106,500	83,000	81,900	21.0	21.3	60.5	60.3	6.45	6.44	
		105,500		81,500		20.0		60.5		6.40		
7	.010	99,500		74,500		21.0		58.7		5.85		
		99,000	99,300	73,750	74,000	19.5	20.0	53.8	56.2	5.33	5.56	
		99,500		73,750		19.5		56.1		5.59		
8	.026	94,500		68,000		22.5		64.0		6.05		
		94,000	94,000	67,500	67,300	23.0	22.3	58.7	60.5	5.52	5.69	
		93,500		66,250		21.5		58.8		5.50		
		104,500		83,750		19.5		54.0		5.62		10 hr at 1050 C air quench
		105,000	104,700	78,750	81,300	19.0	20.0	52.5	54.4	5.52	5.70	20 hr 690 C furnace cool
		104,500		81,250		21.5		56.8		5.95		
9	.019	101,500		76,250		21.5		63.4		6.45		
		99,000	99,700	73,750	74,200	21.0	21.2	61.0	61.5	6.05	6.14	
		98,500		72,500		21.0		60.1		5.93		
		107,500		80,000		17.5		49.4		5.32		
		106,500	106,800	81,250	80,800	12.0	16.2	45.5	49.8	4.85	5.32	
		106,500		81,250		19.0		54.4		5.80		
10	.082	103,500		73,250		17.0		47.6		4.94		10 hr at 1050 C air quench
		105,000	104,000	78,000	76,700	16.5	16.5	50.8	48.2	5.34	5.03	
		103,500		78,750		16.0		46.3		4.30		

* The average R. A. after final heat treatment was used in the plot of R. A. vs Residual Titanium (Fig. 2).

** Q.F. = Tensile Strength × Reduction of Area × 10^{-6}, the average of all results were used in the plot of Q.F. vs Residual Titanium (Fig. 3).

TABLE 3 — SUMMARY OF MECHANICAL PROPERTIES FOR GROUP II CASTINGS (0.013% SULFUR)

Casting Number	Residual Titanium, %	T.S., psi	Average	Y.S., .02% O.S., psi	Average	Elong. % in 2 in.	Average	R.A., %	Average*	Q.F.**	Average	Additional Heat Treatment
16	.002	105,500		81,250		17.0		59.4		6.23		
		103,500	104,800	78,750	80,000	20.0	19.0	60.7	60.3	6.28	6.31	
		105,500		80,000		20.0		60.8		6.42		
		101,500		76,500		20.5		60.8		6.18		12 hr at 690 C furnace cool
		101,500	101,500	77,500	76,800	21.0	20.7	59.6	58.3	6.06	5.92	
		101,500		76,250		20.5		54.4		5.53		
17	.014	103,500		78,750		20.5		63.4		6.56		
		101,000	102,000	76,250	76,500	19.0	19.8	57.6	59.9	5.82	6.11	
		101,500		74,500		20.0		58.6		5.95		
		99,500		73,750		18.0		60.6		6.03		12 hr at 690 C furnace cool
		98,500	99,000	72,500	73,300	18.5	19.0	60.1	60.5	5.93	5.99	
		99,000		73,750		20.5		60.8		6.02		
18	.026	99,000		72,500		23.5		61.1		6.05		
		98,500	98,300	72,500	72,100	22.5	22.3	62.2	60.5	6.13	5.95	
		97,500		71,250		21.0		58.2		5.67		
19	.038	111,000		83,750		21.0		58.1		6.45		
		105,500	107,000	81,250	81,500	19.5	19.8	57.1	56.9	6.05	6.10	
		104,500		79,500		19.0		55.4		5.80		
		100,000		73,750		15.5		37.0		3.70		10 hr at 690 C furnace cool
		99,500	100,000	75,000	74,600	19.0	18.0	51.9	48.3	5.15	4.83	
		100,500		75,000		19.5		56.0		5.64		
20	.061	110,000		87,500		18.0		52.2		5.75		
		107,500	109,300	79,500	84,400	18.0	17.8	51.8	51.0	5.58	5.58	
		110,500		86,250		17.5		48.9		5.41		
		109,500		85,500		17.5		46.9		5.14		10 hr at 690 C furnace cool
		109,500	109,500	85,000	85,200	16.0	17.2	45.8	47.0	5.03	5.16	
		109,500		85,000		18.0		48.4		5.30		

* The average R. A. after final heat treatment was used in the plot of R. A. vs Residual Titanium (Fig. 2).
** Q.F. = Tensile Strength × Reduction of Area × 10^{-6}, the average of all results were used in the plot of Q.F. vs Residual Titanium (Fig. 3).

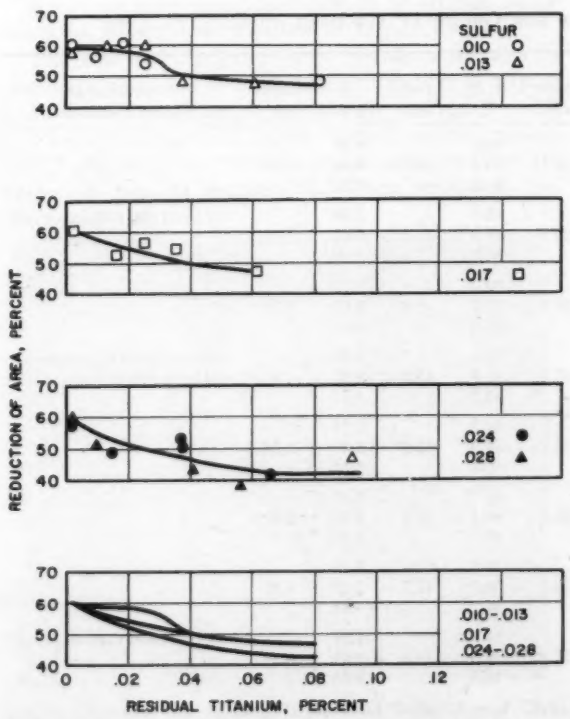


Fig. 2 — Effect of residual titanium and sulphur content on tensile ductility of Cr-Mo-V steel.

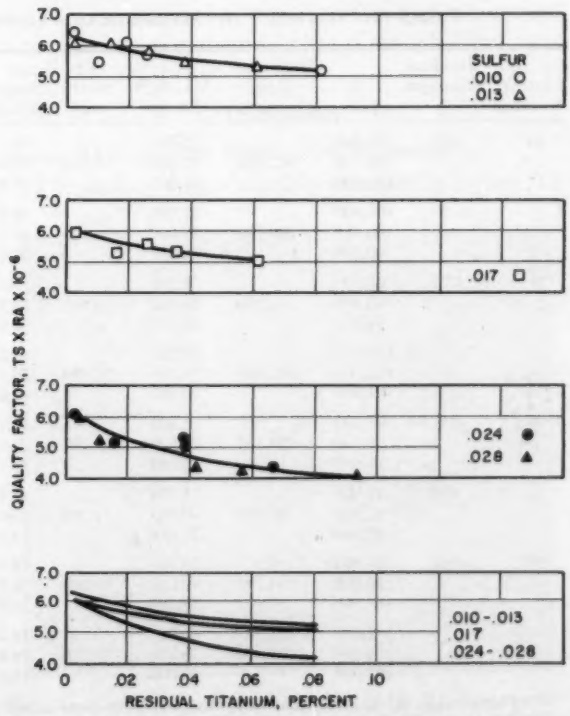


Fig. 3 — Effect of residual titanium and sulphur content on quality factor of Cr-Mo-V steels.

TABLE 4 — SUMMARY OF MECHANICAL PROPERTIES FOR GROUP III CASTINGS (0.017% SULFUR)

Casting Number	Residual Titanium, %	T. S., psi	Average	Y.S., .02% O.S., psi	Average	Elong. % in 2 in.	Average	R. A., %	Average*	Q. F.**	Average	Additional Heat Treatment
26	.002	98,500		73,750		21.0		61.7		6.09		
		98,000	97,800	73,750	73,700	22.0	21.3	63.3	61.2	6.20	6.00	
		97,000		73,500		21.0		58.7		5.70		
27	.016	103,500		76,500		21.5		57.5		5.96		
		101,000	103,300	77,500	77,200	21.0	20.8	56.5	56.5	5.70	5.50	
		105,500		77,500		20.0		55.5		5.85		
		99,000		76,000		20.5		57.2		5.66		12 hr at 690 C furnace cool
		99,500	99,500	76,000	76,200	20.5	20.8	49.2	53.3	4.87	5.31	
		100,500		76,500		21.5		53.6		5.40		
28	.025	101,000		76,250		21.0		62.8		6.35		
		100,000		77,000		20.0		57.0		5.70		
		99,000	99,400	72,500	74,800	23.0	20.7	59.1	56.6	5.85	5.63	
		98,000		72,500		18.0		44.5		4.36		
		99,000		76,000		21.5		59.6		5.90		
29	.035	103,500		80,500		20.0		54.0		5.60		
		103,000	102,800	80,000	79,500	19.5	19.5	51.0	52.7	5.25	5.42	
		102,000		78,000		19.0		53.0		5.41		
		98,500		75,000		22.0		60.1		5.92		12 hr at 690 C furnace cool
		97,500	98,200	73,000	73,500	20.5	20.8	50.0	54.6	4.88	5.37	
		98,500		72,500		20.0		53.8		5.30		
30	.061	108,500		85,000		19.5		51.9		5.64		
		109,500	108,800	87,000	85,700	17.0	18.2	44.6	47.6	4.88	5.18	
		108,500		85,000		18.0		46.3		5.02		
		107,500		85,000		19.5		51.9		5.58		10 hr at 690 C furnace cool
		107,500		85,000		17.0		45.1		4.85		
		107,000	107,250	85,000	84,400	17.0	17.5	44.3	46.6	4.74	4.99	
		107,000		82,500		16.5		44.9		4.80		

* The average R. A. after final heat treatment was used in the plot of R. A. vs Residual Titanium (Fig. 2).
** Q.F. = Tensile Strength × Reduction of Area × 10^{-6}, the average of all results were used in the plot of Q.F. vs Residual Titanium (Fig. 3).

TABLE 5 — SUMMARY OF MECHANICAL PROPERTIES FOR GROUP IV CASTINGS (0.024% SULFUR)

Casting Number	Residual Titanium, %	T. S., psi	Average	Y.S., .02% O.S., psi	Average	Elong. % in 2 in.	Average	R. A., %	Average*	Q. F.**	Average	Additional Heat Treatment
36	.002	101,000		75,000		19.0		62.5		6.36		
		102,750	102,100	77,500	75,800	22.0	20.7	59.1	60.7	6.08	6.20	
		102,500		75,000		21.0		60.6		6.21		
		101,000		77,500		20.0		55.1		5.56		12 hr at 690 C furnace cool
		100,500	100,700	77,000	76,900	20.5	20.2	59.8	57.8	6.01	5.82	
		100,500		76,250		20.0		58.6		5.90		
37	.015	95,500		71,250		21.0		60.6		5.78		
		95,500	95,200	65,000	68,300	21.5	21.5	52.9	55.8	5.05	5.30	
		94,500		68,752		22.0		53.8		5.08		
		102,000		76,250		16.5		53.8		5.48		10 hr at 1050 C air quench
		103,000	102,000	78,000	76,900	16.5	15.3	47.8	48.9	4.92	4.98	20 hr at 690 C furnace cool
		101,000		76,500		13.0		45.2		4.57		
38	.037	98,500		73,500		19.5		54.2		5.34		
		98,000	98,000	73,500	73,200	20.0	19.7	54.9	53.9	5.38	5.28	
		97,500		72,500		19.5		52.7		5.14		
39	.038	99,500		73,750		21.5		56.4		5.61		
		97,000	98,000	70,000	71,700	20.0	20.2	49.4	51.0	4.79	5.00	
		97,500		71,250		19.0		47.1		4.60		
40	.066	104,000		79,500		18.0		49.3		5.13		
		106,000	104,700	80,000	79,000	16.0	16.7	36.7	42.1	3.89	4.41	
		104,000		77,500		16.0		40.2		4.18		
		104,000		80,000		18.0		41.9		4.36		12 hr at 690 C furnace cool
		103,500	103,800	78,750	79,600	18.0	17.2	44.6	41.8	4.61	4.34	
		104,000		80,000		15.5		38.8		4.03		

* The average R. A. after final heat treatment was used in the plot of R. A. vs Residual Titanium (Fig. 2).
** Q.F. = Tensile Strength × Reduction of Area × 10^{-6}, the average of all results were used in the plot of Q.F. vs Residual Titanium (Fig. 3).

TABLE 6 — SUMMARY OF MECHANICAL PROPERTIES FOR GROUP V CASTINGS (0.028% SULFUR)

Casting Number	Residual Titanium, %	T.S., psi	Average	Y.S., .02% O.S., psi	Average	Elong. % in 2 in.	Average	R.A., %	Average*	Q.F.**	Average	Additional Heat Treatment
46	.002	98,000		73,750		20.0		62.3		6.11		
		97,500	97,600	72,500	72,900	21.0	20.0	61.7	61.1	6.02	5.96	
		97,250		72,500		19.0		59.2		5.76		
47	.010	95,000		70,000		22.0		53.2		5.05		
		96,000	95,000	71,250	69,600	20.0	21.2	51.2	53.5	4.91	5.08	
		94,000		67,500		21.5		56.0		5.26		
		105,000		78,750		18.0		51.6		5.41		10 hr at 1050 C air quench
		105,500	105,000	78,750	78,750	18.0	17.7	53.0	51.0	5.60	5.35	20 hr at 690 C furnace cool
		104,500		78,750		17.0		48.4		5.05		
48	.041	100,000		73,750		19.0		45.7		4.57		
		97,000	98,300	72,500	72,900	18.0	17.7	43.4	43.5	4.21	4.28	
		98,000		72,500		16.0		41.5		4.06		
49	.056	103,500		81,250		15.0		45.5		4.71		
		104,500	104,000	80,000	80,400	17.0	16.3	39.2	42.1	4.10	4.38	
		104,000		80,000		17.0		41.6		4.33		
		103,000		80,000		16.0		38.2		3.93		12 hr at 690 C furnace cool
		102,000	102,200	80,000	80,000	15.0	15.3	37.6	38.7	3.83	3.96	
		101,500		80,000		15.0		40.4		4.10		
50	.093	102,000		76,250		12.0		28.9		2.95		
		101,000	101,500	75,000	75,400	14.0	11.0	37.6	33.3	3.80	3.38	
		100,000		72,500		17.0		51.7		5.17		12 hr at 690 C furnace cool
		100,500	100,200	74,500	73,600	15.5	16.7	44.6	46.9	4.49	4.71	
		100,000		73,750		17.5		44.3		4.43		

* The average R. A. after final heat treatment was used in the plot of R. A. vs Residual Titanium (Fig. 2).
** Q.F. = Tensile Strength × Reduction of Area × 10^{-6}, the average of all results were used in the plot of Q.F. vs Residual Titanium (Fig. 3).

Results

The mechanical properties as a function of titanium and sulfur content are summarized in Tables 2 through 6 and Figs. 2 and 3.

Representative photomicrographs of sulfide inclusions are shown and described in Figs. 4 and 5.

The size and distribution of titanium nitrides in the medium sulfur (0.017 per cent) castings as a function of titanium content are shown in Figs. 6 – 10.

The microstructures of the medium-sulfur castings (0.017 per cent) as a function of residual titanium are shown in Figs. 11 through 15. These structures and grain sizes were representative of all castings over the range of sulfur contents investigated. All samples are etched with a mixture of ferric chloride and zephiran chloride to bring out the grain boundaries.

The total oxygen and titanium contents of the castings are summarized in Table 1. The high recovery of titanium was not surprising since the additions were made to metal free of slag.

Discussion

Room Temperature Mechanical Properties

The plots of tensile ductility and quality factor (TS

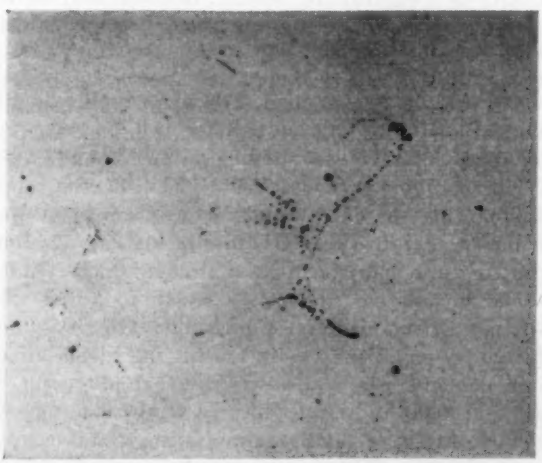


Fig. 4 — Casting No. 28, 0.017% S, 0.025% Ti, typical Type II sulphide inclusions. Note sharply outlined titanium nitrides. Mag. — 100X.

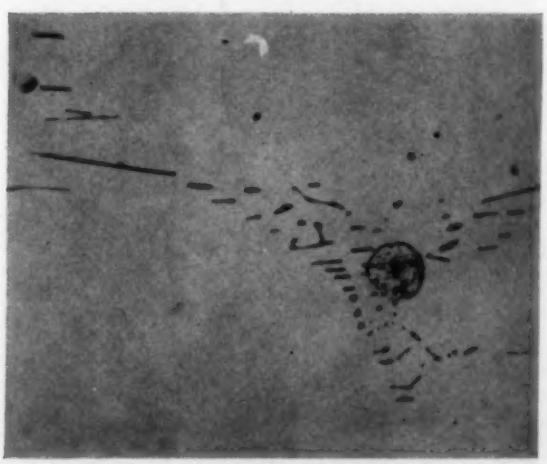


Fig. 5 — Casting No. 50, 0.027% S, 0.093% Ti, fine needle-like pink phase, believed to be some form of Type II sulphide inclusion. Mag. — 500X.

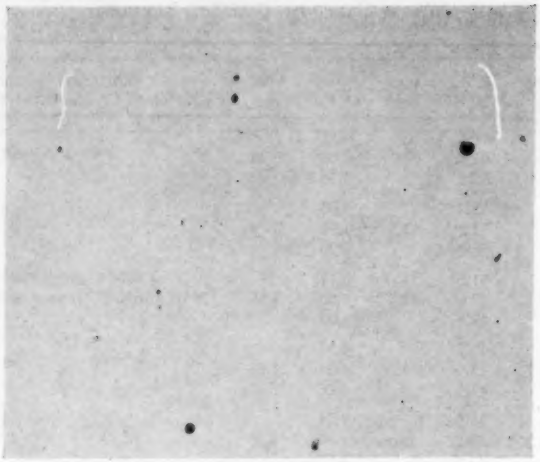


Fig. 6 — Casting No. 26, 0.017% S, 0.002% Ti, fine Type I sulphide inclusions, no titanium nitrides observed. Mag. — 250X.

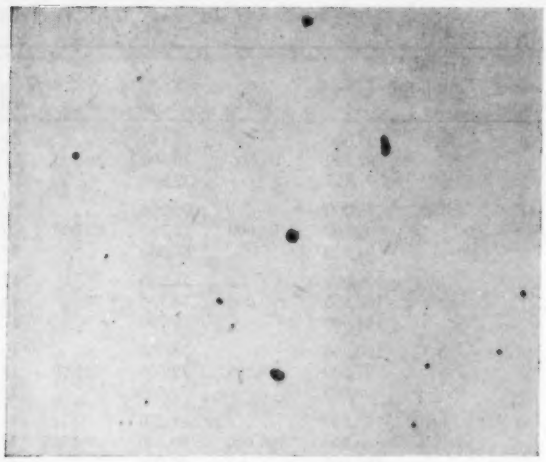


Fig. 7 — Casting No. 27, 0.017% S, 0.016% Ti, fine Type I sulphide inclusions, a few random titanium nitrides were observed. Note nitrides are adjacent to sulphides. Mag. — 250X.

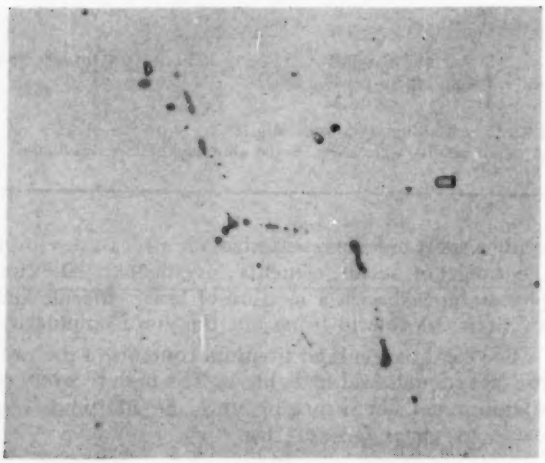


Fig. 8 — Casting No. 28, 0.017% S, 0.025% Ti, Type I and II sulphide inclusions, larger and more numerous randomly distributed titanium nitrides than in Casting No. 27 (Fig. 7). Mag. — 250X.

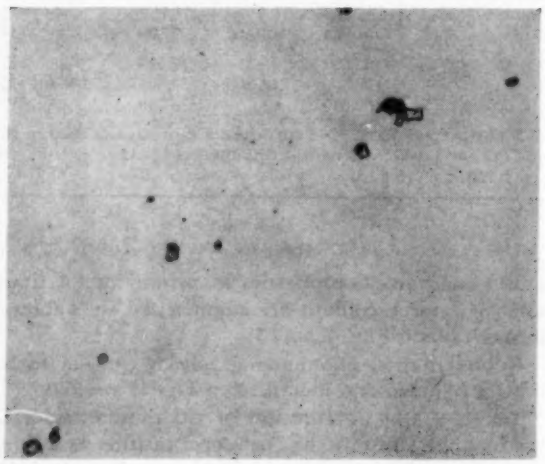


Fig. 9 — Casting No. 29, 0.017% S, 0.035% Ti, Type I sulphide inclusions, larger and more numerous randomly distributed titanium nitrides than in Casting No. 28 (Fig. 8). Mag. — 250X.

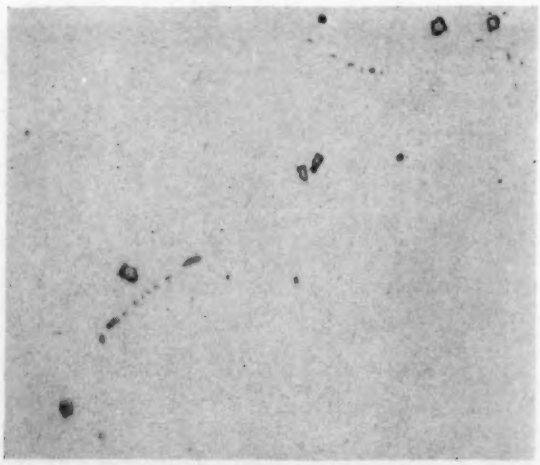


Fig. 10 — Casting No. 30, 0.017% S, 0.061% Ti, Type I and II sulphide inclusions, randomly distributed titanium nitrides, the same size but more numerous than in Casting No. 29 (Fig. 9). Mag. — 250X.

x RA) as a function of residual titanium and sulfur contents show that optimum ductility is obtained with no titanium and low sulfur contents. For a given sulfur level, increasing the residual titanium content results in a gradual decrease in tensile ductility and quality factor. Sims[2] has shown for plain carbon steel that residual titanium over 0.024 per cent results in a sharp decrease in tensile ductility.

Following the first normalize and temper it was observed that the ingots with the highest titanium contents also had the highest strength levels. Additional tempering treatments failed to lower the strength of this material appreciably. This resistance to tempering of the Cr-Mo-V with the higher titanium contents was also observed by Houdremont et al[1] and Zeno. The resistance to tempering suggests that Cr-Mo-V or similar alloys with titanium contents of 0.06 to 0.10 per cent may have improved high temperatuure properties as compared to silicon deoxidized steels.



Fig. 11 — Casting No. 26, 0.017% S, 0.002% Ti, upper bainite. Mag. — 100X.



Fig. 12 — Casting No. 27, 0.017% S, 0.016% Ti, upper bainite, with exception of few large grains, average grain size is finer than in Casting No. 26 (Fig. 11). Mag. — 100X.



Fig. 13 — Casting No. 28, 0.017% S, 0.025% Ti, upper bainite, with exception of few large grains, average grain size is finer than in Castings No. 26 and 27 (Figs. 11 and 12). Mag. — 100X.

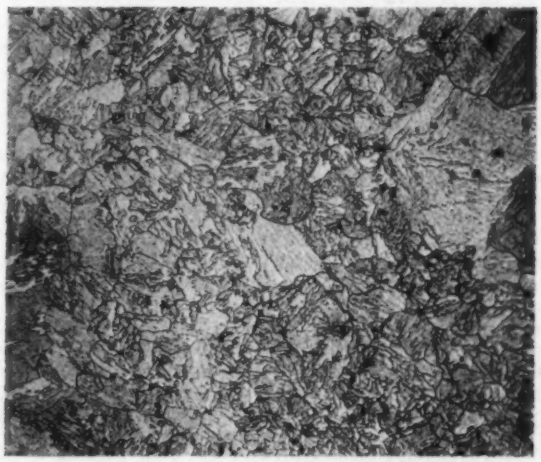


Fig. 14 — Casting No. 29, 0.017% S, 0.035% Ti, upper bainite, average grain size is comparable to Casting No. 28 (Fig. 13). Mag. — 100X.

Metallographic Results

The Cr-Mo-V with a sulfur level of 0.010 - 0.013 per cent was generally free of Type II sulfide inclusions for all titanium contents, with the exception of isolated areas containing a line-up of fine sulfides. At the 0.017 per cent sulfur level a few Type II sulfide inclusions were observed. These inclusions first appeared at the 0.025 per cent titanium level and increased in population with increasing titanium content.

Figure 4 shows the nature of these sulfides for the 0.025 per cent titanium content. As the sulfur content increased to 0.024 - 0.028 per cent the number of Type II inclusions observed increased considerably. Again the number of inclusions formed also increased with increasing titanium content. It is pointed out that the predominant sulfide inclusion was Type I for all sulfur and titanium levels investigated with the exception of 0.24 per cent sulfur and 0.066 per cent titanium and 0.028 per cent sulfur and

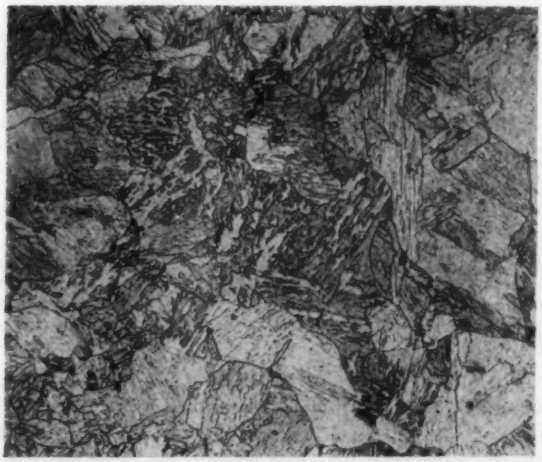


Fig. 15 — Casting No. 30, 0.017% S, 0.061% Ti, upper bainite, average grain size is appreciably coarser than in Casting No. 26. Mag. — 100X.

0.056 per cent and 0.093 per cent titanium where the major sulfide inclusion was Type II. In general, the physical appearance of the Type II sulfide inclusions observed was quite similar. An exception was found in Sample 50 (0.027 per cent S and 0.093 per cent Ti). In addition to the presence of many angular pink and yellow carbonitrides and rounded gray sulfides, a large amount of very fine needle-like pink phase was observed (Fig. 5).

This was concluded to be a product of the high titanium and sulfur content since material containing 0.010 per cent S and 0.082 per cent Ti did not contain these inclusions. Angular yellow titanium - nitrides were observed in all the titanium-bearing material (Figs. 7 to 10). Only a moderate amount of nitrides was formed even at the highest titanium contents because of the low nitrogen level (0.008 per cent) of this material. In general, it was observed that the inclusion count of the steel increased in proportion to the sulfur and titanium content.

An anomalous condition was observed in the grain size of the Cr-Mo-V. The material titanium deoxidized with 2 lb/ton (0.061 per cent residual) resulted in the largest grain size following heat treatment (Fig. 15). In fact this material had a larger grain size than the material free of titanium (Fig. 11). The smaller titanium contents (0.016 and 0.025 per cent) on the other hand resulted in somewhat finer grain sizes compared to the titanium-free material (Figs. 12 and 13). Since the grain size of the Cr-Mo-V was not affected by the sulfur content, the 0.017 per cent sulfur level material was selected as representative of the grain sizes obtained with several titanium contents (Figs. 11 to 15).

An examination of free energy of formation data at steel-making temperatures obtained by Chipman,[3] Holladay[4] and Quill[5] and some calculations using these data may be helpful in understanding the reasons for the grain refining and then coarsening action of titanium in Cr-Mo-V steel.

		Free Energy of Formation ΔF at 1600 C (2912 F) in Cal/Mol*
Most Stable	TiO_2	−61,120
	TiN	−32,640
	TiS	−25,730
	TiC	−18,270
	VN	+ 1,730
	Mn_5N_2	+10,200
	Si_3N_4	+16,610
	Fe_4N	+19,000
Least Stable	Mo_2N	+22,700

*The order of the reaction of the compounds given above check reasonably well with other data. The calculations were made for a temperature of 1600 C (2912 F). As the liquid steel cools to the solid state, the free energy of the compounds will change, and therefore the order of the compound stability may also change.

Note that following the oxide, the nitride of titanium is the most stable compound that forms in Cr-Mo-V. The TiN is also appreciably more stable (forms more readily) than any of the other nitrides. Sufficient titanium to completely deoxidize the steel apparently results in some grain refinement, due perhaps to the formation of titanium oxide and extremely fine titanium carbonitrides which act as grain growth inhibiting agents.

It was observed that at the lower titanium levels (0.016 to 0.025 per cent) the size of the TiN compound formed was extremely fine and quite randomly distributed (Figs. 7 and 8). As the titanium content increased (0.035 to 0.061 per cent) the size of the TiN compound was comparable, the quantity formed increased and was randomly distributed without regard to grain boundaries. Since the nitrogen level of the steel was the same for all castings, it is reasonable to deduce that on a micro scale the areas adjacent to a large titanium carbonitride will be titanium and nitrogen poor.

Then conversely if one expects randomly distributed fine nitrides to inhibit grain growth, nitrogen-free areas should result in coarse grain sizes. It is believed that the coarser and mixed grain sizes of the Cr-Mo-V were caused in part by the variations in nitrogen and titanium content on a micro scale.

Another consideration is the effect of nitrogen on the nucleation and grain growth of upper bainite, the structure generally obtained in normalized Cr-Mo-V (Figs. 11 to 15). Data obtained by the authors indicated that increasing nitrogen content will result in grain refinement in Cr-Mo-V steels.

Summary of Total Oxygen Content

The results of the total oxygen determinations (Table 1) are in accord with oxygen - titanium solubility data determined by Lyaudis and Samarin[6] and Comstock et al.[7] The metal temperature at the time of our titanium additions was approximately 2850 F (1566 C) based on our normal foundry practice.

A great deal of work has been done in recent years on the mechanism of inclusion formation in cast steels. It is generally accepted that the final form of metal sulfides is dependent upon the oxygen level of the molten steel. Hilty and Crafts[8,9] have shown for iron alloy systems that the oxygen - sulfur ratio is important in controlling the type of sulfide inclusion formed. The results of our investigation also show, as might be reasoned, that the quantity of sulfur present is as important as the oxygen level, if not more so. After all, sulfide inclusions of any type, could not form if the basic constituent, (sulfur), were eliminated.

This point is emphasized by examining the following data:

Sample No.	Sulfur, %	Oxygen, %	Titanium, %	Quality Factor (T.S. x R.A. x 10^{-6})
9	0.010	0.0051	0.019	6.14
39	0.024	0.0053	0.038	5.00
36	0.024	0.0081	0.002	6.01

Note that excellent mechanical properties were obtained with combinations of low sulfur and oxygen and high sulfur and oxygen, however, the high sulfur-low oxygen combination did not fare so well. Since the solubility of sulfur in molten steel increases with decreasing oxygen content these results were not surprising.

The formation of isolated Type II sulfide inclusions in the lowest sulfur content, titanium deoxi-

dized material may be explained on the basis of micro variations in the oxygen content. It was observed that spectrographic determinations of titanium in Cr-Mo-V were extremely unreliable because of the titanium concentration caused by the formation of carbonitrides. We found variation in titanium content from 0.030 - 0.085 per cent areas only 1/8 in. apart. These areas of titanium-rich material will no doubt have low oxygen levels on a micro scale.

Isolated Type II sulfides are often observed in these areas of concentrated titanium nitrides (Figs. 4, 8 and 10).

Since the spectrographic method was found to be unreliable, all titanium analyses were made by the wet colorimetric technique.

From the data presented in this study there appears to be a critical level of both sulfur and titanium which determines the quantity and the nature of the sulfide inclusions formed. In general, it was observed that the Type II sulfides formed in the titanium deoxidized material were appreciably finer than Type II sulfides observed in silicon or aluminum deoxidized Cr-Mo-V. Since TiS has a different crystallographic structure than AlS or FeS this is not surprising. It is pointed out, however, that even though the Type II inclusions observed were very fine, they were quite effective in reducing tensile ductility if present in appreciable quantity.

Conclusions

Several castings (ingots) of cast Cr-Mo-V with sulfur contents of 0.010 to 0.028 per cent and titanium additions of 0 to 2 lb/ton were examined. The base composition was the same for all castings. The mechanical properties, inclusion types and distribution and microstructure of these castings were determined. The following conclusions are made:

1. The tensile ductility (R. A.) and quality factor (T. S. x R. A.) of cast Cr-Mo-V steel decreases with increasing sulfur and/or titanium content.

2. Cast Cr-Mo-V, titanium deoxidized with 2 lb/ton exhibited a marked resistance to tempering.

3. Cast Cr-Mo-V with a maximum sulfur content of 0.017 per cent, titanium deoxidized with up to 1 lb per ton resulted in good tensile properties (minimum quality factor of 5×10^6).

4. Localized areas of fine Type II sulfide inclusions were observed even in the lowest sulfur content titanium deoxidized material. These inclusions however, were not present in sufficient quantity to materially affect tensile ductility.

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DISCUSSION

Chairman: C. K. Donoho, American Cast Iron Pipe Co., Birmingham, Ala.

Co-Chairman: W. O. Igelman, National Malleable & Steel Castings Co., Melrose Park, Ill.

Secretary: D. N. Rosenblatt, American Foundry & Machine Co., Salt Lake City, Utah.

C. E. Sims[1] *(Written Discussion):* This work appears to have been carefully done and the papers should be a useful contribution to our knowledge of the behavior of aluminum and titanium in cast steel. This is particularly true in the present case because the steel described has a lower carbon content than those commonly used for steel castings, and the behavior of deoxidizers is quantitatively different in such low carbon steels as compared to medium-carbon steels.

No serious problem of ductility should be encountered in these steels, but the data presented show a consistent decrease in ductility when small additions of either aluminum or titanium are made. A characteristic difference in the effect of aluminum and titanium is also shown, namely, whereas increasing additions of aluminum first lowered ductility to a minimum value and then caused a partial recovery, increasing additions of titanium produced a progressive decrease in ductility. Furthermore, there was a good correlation between sulphur content and ductility, the higher sulphur steels being consistently lower in ductility, which agrees with the findings of other investigators. These factors indicate strongly that the nonmetallic inclusions, particularly the type and quantity of sulphides, can account for the variations in ductility noted. When the sulphur content is sufficiently low, neither Al nor Ti will have much effect on ductility when used as deoxidizers.

Aluminum is a noted grain refiner but its action is influenced by the composition of the steel, notably the nitrogen and sulphur contents. Titanium, on the other hand, has a poor record as a grain refiner, except under certain limited conditions. One of these conditions is a medium manganese content. In such steel, a small addition of titanium gives a remarkably fine grain with a high grain coarsening temperature. No consideration seems to have been given to the effects of vanadium in this work. Vanadium gives a fine-grained structure if the heat-treating temperature is kept within a small interval above the Ac3 temperature. Vanadium steels have low grain coarsening temperatures, however, and when they are heated above this temperature, they are apt to be very coarse grained.

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(See an allied paper starting on the following page.)

EFFECT OF ALUMINUM DEOXIDATION ON THE MECHANICAL PROPERTIES AND MICROSTRUCTURE OF CAST Cr-Mo-V STEEL

By

R. S. Zeno,* R. D. Ahles,** and W. R. Nestle*

ABSTRACT

A study was made to determine the effect of aluminum deoxidation on the mechanical properties, inclusion type and microstructure of cast Cr-Mo-V as a function of sulfur content. A range of aluminum additions from 0 to 2 lb/ton and sulfur contents of 0.011 to 0.027 per cent was studied.

The base composition and heat treatment were the same for all castings.

The following conclusions were made:

1. For sulfur contents ranging from 0.011 to 0.018 per cent, the addition of 0 to 2 lb of aluminum per ton caused little loss in tensile ductility and quality factor (T.S. x R.A.). The predominant inclusion was the Type I sulfide.

2. For sulfur contents of 0.024-0.027 per cent, the addition of 0 to 2 lb of aluminum per ton produced an appreciable loss in tensile ductility and quality factor and a strong tendency towards the formation of Type II sulfide inclusions.

3. For these lower sulfur castings (0.011 to 0.018 per cent), the addition of aluminum (½ lb/ton) caused a decrease in the mechanical properties with no change in the grain size or microstructure. This suggests an effect of aluminum per se on the mechanical properties of alloy steels.

4. Type II sulfide inclusion formation in Cr-Mo-V can be minimized for all oxygen levels and many deoxidation practices provided the sulfur contents are below 0.018 per cent.

5. The role of oxygen in sulfide inclusion formation in alloy steels for a given sulfur level and deoxidation practice has not been clearly defined to date.

Introduction

Several investigators[1,2,3,4,5,6] have shown that the aluminum deoxidation of plain low- and medium-carbon cast steels must be carefully controlled to prevent a loss in tensile ductility. The importance of low-sulfur contents has also been shown,[2,5] particularly when aluminum deoxidation is practiced. The quantity of aluminum added and the sulfur content will determine to a great extent the type of sulfide inclusion formed in cast steels. Extreme caution must be exercised to prevent the formation of the undesirable intergranular Type II sulfide inclusion and the attendant loss in tensile ductility.

This investigation was made to study the effect of aluminum deoxidation on the mechanical properties, inclusion type and microstructure of cast Cr-Mo-V as a function of sulfur content. A range of aluminum additions from 0 to 2 lb/ton and sulfur contents of 0.011 to 0.027 per cent was studied.

Procedure

Material

A base heat of Cr-Mo-V produced by the basic electric process was used in this study. The carbon content prior to the block was 0.17 per cent. This heat was recarburized to 0.20 per cent after the block (one hour before tapping) with high-carbon ferromanganese.

The ladle analysis of this heat in weight per cent was as follows: C, 0.20; Si, 0.38; Mn, 0.67; S, 0.011; P, 0.011; Cr, 1.05; Ni, 0.15; Mo, 1.10; V, 0.17; and N, 0.008.

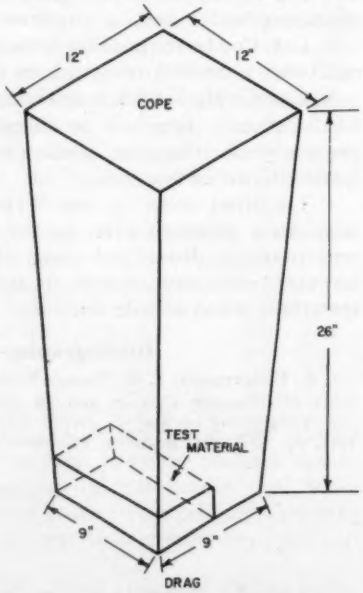


Fig. 1 — Test Casting.

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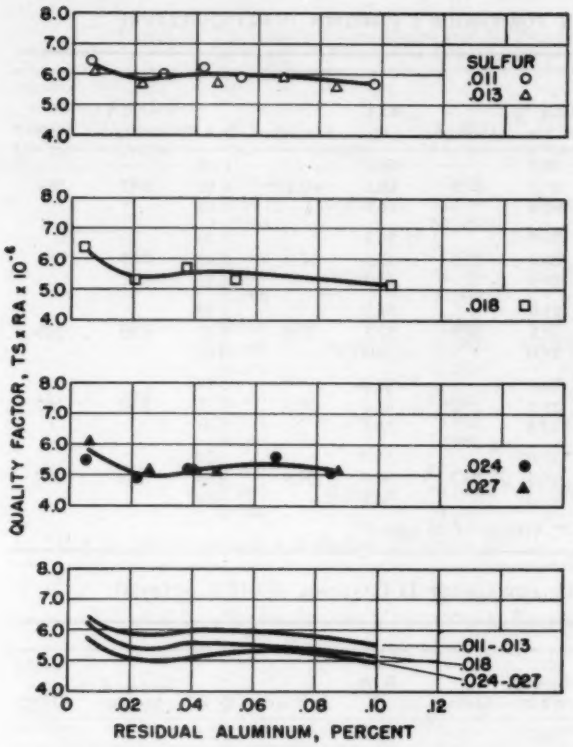


Fig. 2 — Effect of residual aluminum and sulphur content on quality factor of Cr-Mo-V steel.

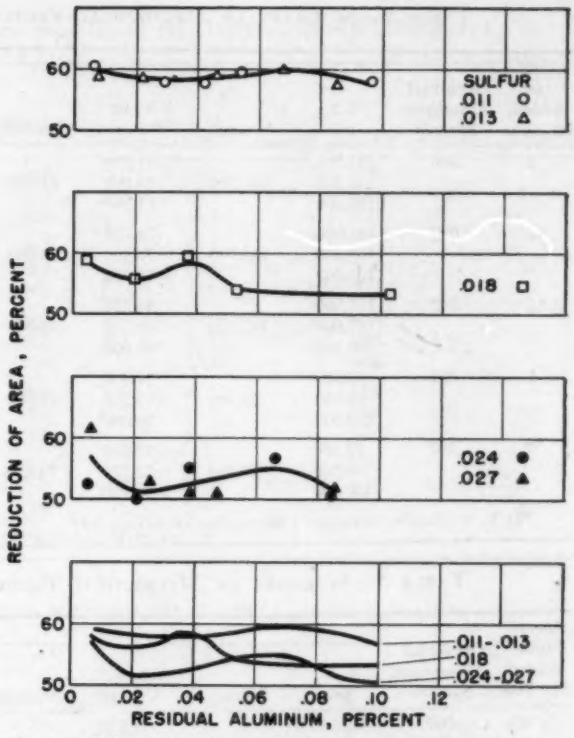


Fig. 3 — Effect of residual aluminum and sulphur content on tensile ductility of Cr-Mo-V steel.

Table 1 — Summary of Sulfur, Oxygen and Aluminum Contents of Cr-Mo-V Castings

Casting Number	Al Added %	Total Al Wet***	Total Al Spectro	Total O_2**	Sulfur*
6	Nil	.004	.006	.0056	.011
2	.0183	.019	.029	.0062	.011
3	.0365	.041	.042	.0047	.011
4	.055	.054	.054	.0060	.011
5	.110	.097	.097	.0061	.011
11	Nil		.007	.0055	.013
12	.0183		.022	.0043	.013
13	.0365		.046	.0045	.013
14	.055		.068	.0040	.013
15	.110		.085	.0047	.013
21	Nil		.004	.0055	.018
22	.0183		.020	.0052	.018
23	.0365		.037	.0048	.018
24	.055		.053	.0045	.018
25	.110		.103	.0037	.018
31	Nil		.005	.0043	.024
32	.0183		.021	.0052	.024
33	.0365		.038	.0055	.024
34	.055		.066	.0042	.024
35	.110		.084	.0040	.024
41	Nil		.006	.0055	.027
42	.0183		.025	.0036	.027
43	.0365		.038	.0036	.027
44	.055		.047	.0053	.027
45	.110		.085	.0027	.027

* Average sulfur content based on drillings from 25 broken tensile bars; range = ± .002% from the averages.

** Total oxygen from vacuum fusion analysis.

*** Total aluminum (wet analysis), these results were used to calibrate standards for subsequent spectrographic determinations of total aluminum.

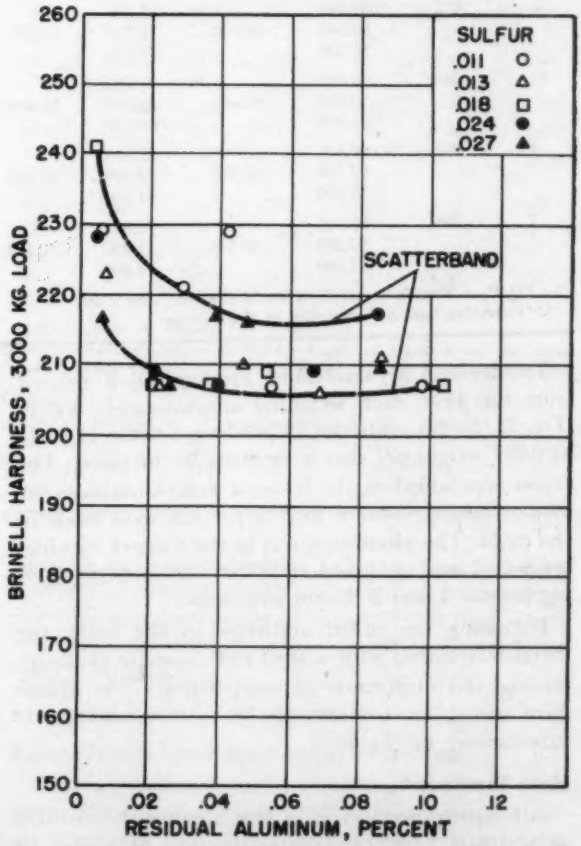


Fig. 4 — Effect of residual aluminum and sulphur content on Brinell hardness of Cr-Mo-V steel.

TABLE 2 — SUMMARY OF MECHANICAL PROPERTIES FOR GROUP I CASTINGS (0.011% SULFUR)

Aluminum Added, Lb/Ton	Residual Aluminum, Spectro	T.S., psi	Average	Y.S., .02% O.S., psi	Average	Elong. % in 2 in.	Average	R.A., %	Average	Q.F.*	Average	BHN**
0	.006	107,500		81,250		23.0		60.0		6.45		
		106,500	106,500	83,000	81,900	21.0	21.3	60.5	60.5	6.45	6.43	229
		105,500		81,500		20.0		60.5		6.38		
1/3	.029	105,000		78,750		20.0		59.1		6.20		
		103,000	103,700	78,300	78,600	20.0	20.0	57.5	57.9	5.92	6.01	221
		103,000		78,750		20.0		57.3		5.91		
2/3	.042	106,500		83,750		21.0		58.9		6.26		
		107,000	107,700	82,500	83,800	19.5	20.5	57.7	57.8	6.17	6.22	229
		109,500		85,000		21.0		56.8		6.22		
1	.054	98,500		74,000		24.0		60.3		5.94		
		98,500	99,000	73,750	74,600	21.5	23.0	58.6	59.3	5.77	5.87	207
		100,000		76,000		23.5		59.1		5.91		
2	.097	97,500		73,750		25.0		58.7		5.73		
		98,700	98,700	73,750	74,600	20.5	22.7	56.0	57.8	5.46	5.69	207
		100,000		76,250		22.5		58.8		5.88		

*Q.F. = Tensile Strength x Reduction of Area x 10^{-6} **Average of 10 values

TABLE 3 — SUMMARY OF MECHANICAL PROPERTIES FOR GROUP II CASTINGS (0.013% SULFUR)

Aluminum Added, Lb/Ton	Residual Aluminum, Spectro	T.S., psi	Average	Y.S., .02% O.S., psi	Average	Elong. % in 2 in.	Average	R.A., %	Average	Q.F.*	Average	BHN**
0	.007	102,500		78,750		18.5		58.8		6.03		
		104,000	103,500	80,000	79,300	21.0	20.3	57.0	59.0	5.93	6.10	223
		104,000		80,000		21.5		61.1		6.35		
1/3	.022	98,500		73,750		15.0		38.8***		—		
		96,500	96,500	71,250	71,300	21.0	21.0	60.2	58.6	5.81	5.65	207
		96,500		71,250		21.0		56.9		5.49		
2/3	.046	97,000		70,000		22.5		60.7		5.89		
		96,000	96,000	72,500	70,400	21.5	21.3	58.3	59.3	5.59	5.69	210
		95,000		68,750		20.0		58.8		5.58		
1	.068	98,500		73,750		22.0		60.1		5.92		
		97,500	97,700	73,000	72,700	23.5	22.7	59.8	60.2	5.84	5.89	206
		97,000		71,250		22.5		60.8		5.90		
2	.085	98,000		72,000		22.0		58.8		5.76		
		97,000	97,300	71,250	72,400	21.5	22.0	56.9	57.7	5.51	5.61	211
		97,000		74,000		22.5		57.3		5.55		

*Q.F. = Tensile Strength x Reduction of Area x 10^{-6} **Average of 10 values
***Defective bar, not included in the average

Twenty-five dry-sand-mold castings were poured from this heat, each weighing approximately 900 lb (Fig. 1). Sulfur additions to produce a range of 0.011 to 0.027 weight per cent were made in the ladle. The sulfur was added in the form of FeS. Aluminum additions ranging from 0 to 2 lb per ton were made in the mold. The aluminum was in the form of turnings for the 1/3 and 2/3 lb/ton additions and 1-in. pieces of pig for the 1 and 2 lb/ton additions.

Following the sulfur additions in the ladle, the metal was stirred with a steel rod to assure thorough mixing and uniformity of composition. The aluminum and sulfur contents of the various castings are summarized in Table 1.

Heat Treatment

All castings were given a heat treatment consisting of a 1050 C (1922 F) austenitize for 10 hours, air cool, temper at 690 C (1274 F) for 20 hours followed by a furnace cool.

Tests

The following tests and analyses were made on sections taken from the same location in each casting (Fig. 1):

1. Room temperature tensile tests
2. Metallographic examination
3. Total oxygen determinations
4. Sulfur and aluminum determinations

Results

The mechanical properties as a function of aluminum and sulfur content are summarized in Tables 2 through 6 and Figs. 2 to 4. The sulfur content of each group of castings was verified by analyzing millings taken from 25 broken tensile bars and averaging the results. The range of sulfur values from each group of castings varied only ± 0.002 per cent from the averages.

Table 4 — Summary of Mechancal Properties for Group III Castings (0.018% Sulfur)

Aluminum Added, Lb/Ton	Residual Aluminum, Spectro	T. S., psi	Average	Y.S., .02% O.S., psi	Average	Elong. % in 2 in.	Average	R. A., %	Average	Q. F.*	Average	BHN**
0	.004	107,500		81,500		20.5		59.4		6.38		
		107,500	107,500	76,250	79,400	21.5	20.8	60.1	59.3	6.46	6.37	241
		107,750		80,500		20.5		58.3		6.28		
1/3	.020	97,500		72,500		20.5		55.5		5.41		
		97,000	96,700	68,750	70,400	18.0	18.8	56.7	55.7	5.50	5.38	207
		95,500		70,000		18.0		54.9		5.24		
2/3	.037	96,500		73,000		23.0		60.7		5.86		
		96,000	96,000	67,500	69,200	22.5	22.5	60.7	59.8	5.83	5.75	207
		95,500		67,000		22.0		58.1		5.55		
1	.053	98,500		75,000		26.0		60.0		5.91		
		98,750	97,900	75,000	74,200	19.0	21.0	50.3	54.0	4.96	5.29	209
		96,500		72,500		18.0		51.7		5.00		
2	.103	97,500		73,750		20.5		56.0		5.46		
		97,000	97,000	73,750	73,800	21.0	21.2	52.2	53.7	5.06	5.21	207
		96,500		73,750		22.0		53.0		5.11		

*Q. F. = Tensile Strength x Reduction of Area x 10^{-6} **Average of 10 values

Table 5 — Summary of Mechanical Properties for Group IV Castings (0.024% Sulfur)

Aluminum Added, Lb/Ton	Residual Aluminum, Spectro	T. S., psi	Average	Y.S., .02% O.S., psi	Average	Elong. % in 2 in.	Average	R. A., %	Average	Q. F.*	Average	BHN**
0***	.005	104,500		80,500		19.0		50.6		5.29		
		105,500	104,800	82,000	80,700	19.5	19.5	53.0	52.5	5.59	5.50	229
		104,500		79,500		20.0		53.8		5.62		
1/3	.021	98,500		71,250		19.0		53.8		5.30		
		98,500	98,300	72,500	72,100	18.5	18.2	45.8	49.9	4.51	4.91	209
		98,000		72,500		17.0		50.1		4.91		
2/3	.038	93,500		71,000		20.0		58.4		5.45		
		94,500	93,800	70,000	69,900	21.0	19.3	55.5	55.3	5.25	5.16	207
		93,500		68,750		17.0		51.2		4.79		
1	.066	97,500		73,750		21.5		58.8		5.73		
		96,500	96,700	70,000	71,300	21.0	21.2	55.2	56.7	5.33	5.48	209
		96,000		70,000		21.0		56.1		5.38		
2	.084	99,500		76,250		19.5		51.9		5.16		
		99,000	99,200	75,000	75,000	19.0	19.3	49.0	50.8	4.85	5.03	217
		99,000		73,750		19.5		51.4		5.08		

*Q. F. = Tensile Strength x Reduction of Area x 10^{-6} **Average of 10 values

***Material contained Type II sulfide inclusions. Reheat treated (additional 8 hr temper at 690 C) material taken from the same casting and adjacent to the original test area was free of any Type II sulfide inclusions and had the following tensile properties:

T. S., psi	Average	Y.S., .02% O.S., psi	Average	Elong. % in 2 in.	Average	R. A., %	Average	Q. F.	Average
101,500		77,000		21		57.5		5.85	
101,000	101,700	75,000	75,800	20	20.8	54.7	57.7	5.53	5.86
102,500		75,500		21.5		60.8		6.25	

The examination of unetched micro samples taken from the broken tensile bars of each casting is summarized in Table 7. Representative photomicrographs of various types of sulfide inclusions described in Table 7 are shown and described in Figs. 5 to 12.

The microstructures of the low-sulfur castings (.011 per cent) as a function of added aluminum are shown in Figs. 13 to 19. These structures and grain sizes were representative of all castings over the range of sulfur contents investigated. All samples were etched with a mixture of ferric chloride and zephiran chloride to bring out the grain boundaries.

The total oxygen and aluminum contents of the castings are summarized in Table 1. Note the high recovery of aluminum. This was not surprising since the additions were made to metal free of slag. There was also some aluminum introduced from the ferroalloy additions.

Discussion

Room Temperature Mechanical Properties

Before aluminum can be considered as a method of deoxidizing turbine grade steels, its effect on mechanical properties must be evaluated. There is a wealth of data on this subject in the literature for low- and medium-carbon cast steel.

TABLE 6 – SUMMARY OF MECHANICAL PROPERTIES FOR GROUP V CASTINGS (0.027% SULFUR)

Aluminum Added, Lb/Ton	Residual Aluminum, Spectro	T. S., psi	Average	Y.S., .02% O.S., psi	Average	Elong. % in 2 in.	Average	R. A., %	Average	Q. F.*	Average	BHN**
0	.006	98,500		75,000		21.0		62.7		6.17		
		98,000	98,300	73,750	74,400	20.0	20.2	62.3	61.9	6.10	6.08	217
		98,500		74,500		19.5		60.6		5.97		
1/3	.025	97,000		71,250		19.5		55.7		5.40		
		97,000	96,800	71,500	71,800	18.0	19.0	51.3	53.2	4.97	5.14	207
		96,500		72,500		19.5		52.5		5.06		
2/3	.038	100,000		76,250		18.0		52.5		5.25		
		100,000	99,700	76,250	76,300	20.0	19.5	51.7	51.2	5.17	5.12	217
		99,000		76,500		20.5		49.9		4.94		
1	.047	100,000		78,000		13.0		29.2***		—		
		99,500	99,000	77,000	76,600	18.5	18.3	49.5	51.2	4.92	5.07	216
		98,500		76,250		18.0		52.9		5.21		
2	.085	96,500		73,750		21.0		56.1		5.42		
		97,500	97,000	75,000	74,200	21.5	20.8	52.4	52.4	5.11	5.10	210
		97,000		73,750		20.0		48.6		4.76		

*Q. F. = Tensile Strength x Reduction of Area x 10^{-6} **Average of 10 values
***Defective bar, not included in the average

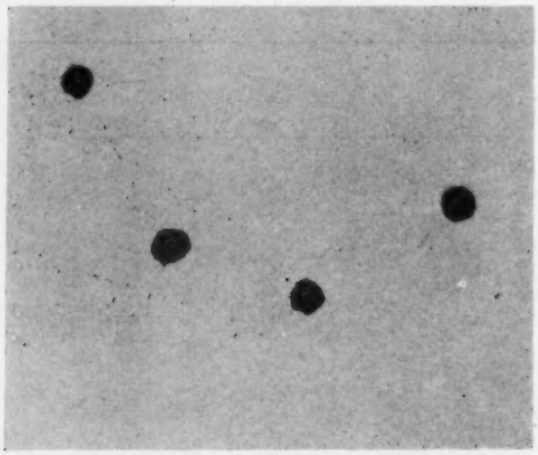


Fig. 5 — Group I, Casting 6, shows a few large Type I sulphide inclusions. Sample contained mostly fine scattered Type I's. Mag. — 500X.

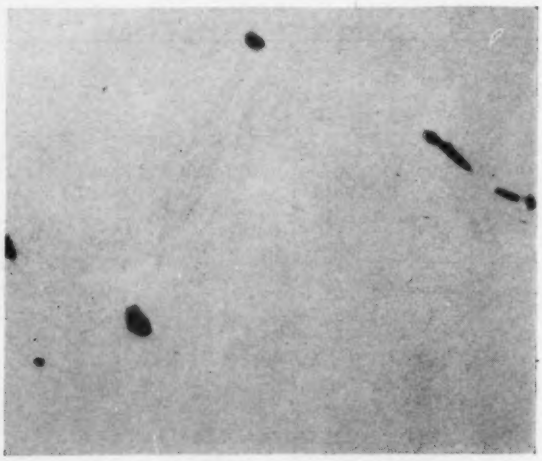


Fig. 6 — Group I, Casting 5, note cigar-shaped sulphides. Mag. — 500X.

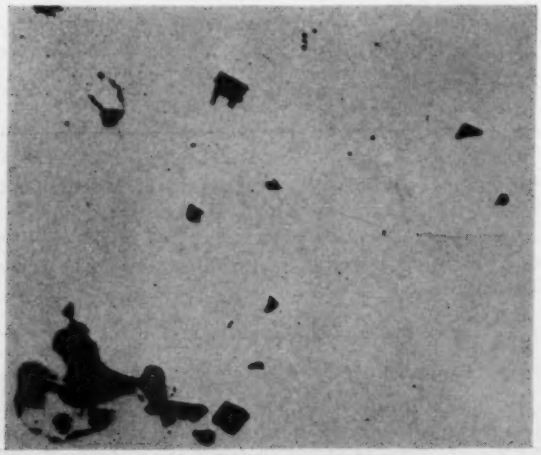


Fig. 7 — Group I, Casting 4, note area of Type III sulphides and microshrink. Sample contained mostly fine scattered Type I's. Mag. — 250X.

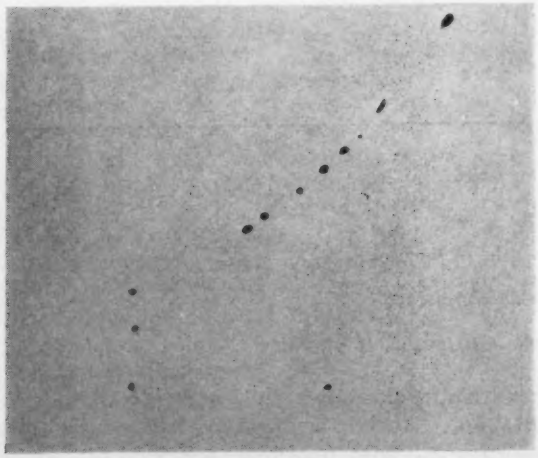


Fig. 8 — Group II, Casting 15, shows some line up of Type I sulphides. Mag. — 500X.

Table 7 — Summary of Sulfide Inclusion Types for Various Sulfur Contents and Aluminum Additions

Aluminum Added, Lb/Ton	Sulfur Content, %	Group and Casting No.	Types of Sulfide Inclusion Found in Microstructure
0	0.011	I, 6	I A
1/3		I, 2	II A
2/3		I, 3	I A
1		I, 4	II A*
2		I, 5	II B
0	0.013	II, 11	I B
1/3		II, 12	I B
2/3		II, 13	II A
1		II, 14	II A
2		II, 15	II B
0	0.018	III, 21	I A
1/3		III, 22	II A
2/3		III, 23	I A
1		III, 24	II A
2		III, 25	II A
0	0.024	IV, 31	II E
1/3		IV, 32	II A
2/3		IV, 33	II A
1		IV, 34	II C
2		IV, 35	II C
0	0.027	V, 41	II C
1/3		V, 42	II C
2/3		V, 43	II D
1		V, 44	II D
2		V, 45	II D

*Some Type III sulfides

I A Fine randomly scattered Type I sulfide inclusions (inclusion size shown in Fig. 8).

II B Fine randomly scattered Type I sulfide inclusions plus a line up (Fig. 8).

II A Fine randomly scattered Type I sulfide inclusions plus a line up and some tendency toward Type II's (Figs. 6 and 8).

II B Large randomly scattered Type I sulfide inclusions plus a line up and some tendency toward Type II's (Figs. 5, 6 and 8).

II C Fine randomly scattered Type I sulfide inclusions plus a line up and a strong tendency toward isolated Type II sulfides (Figs. 8, 11 and 12).

II D Large randomly scattered Type I sulfide inclusions plus a line up and a strong tendency toward isolated Type II sulfides (Figs. 5, 11 and 12).

II E Fine randomly scattered Type I sulfide inclusions and numerous Type II sulfide inclusions (Figs. 9 and 10).

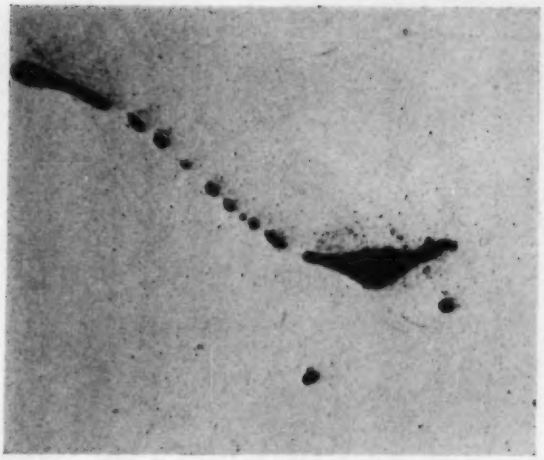


Fig. 9 — Group III, Casting 22, shows a chain Type II sulphide inclusion. Mag. — 500X.

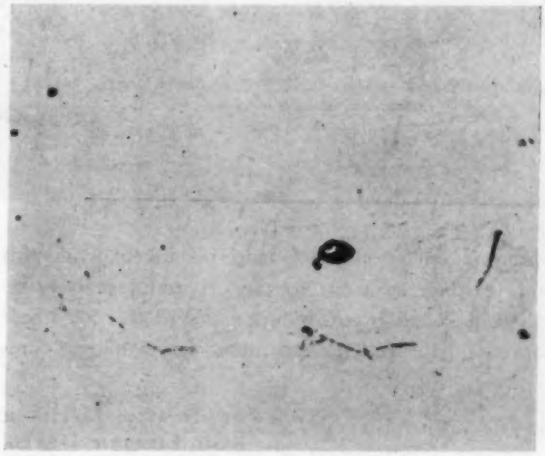


Fig. 10 — Group IV, Casting 31, shows fine, chain Type II sulphide inclusions. Mag. — 100X.

Sims and Lillieqvist,[1] Sims and Dahle[2] and Gagnebin[3] found that aluminum additions had to be carefully controlled to prevent a loss in tensile ductility for plain medium-carbon steels. Sims[4] further reported for a plain medium-carbon steel that additions of aluminum in the range of 1/3 to 1 lb per ton will result in a sharp drop in tensile ductility.

Lillieqvist[5] reported similar results for a basic open hearth plain medium-carbon cast steel. Lillieqvist's data show that the lowest tensile ductilities were obtained with an aluminum addition of 1/2 lb per ton, and high sulfur contents (Fig. 20).

Included in Fig. 20 are the averages from 20 heats of low-sulfur basic electric annealed plain low-carbon steel produced by the Schenectady Steel Foundry (Table 8). Note that the precipitous drop in reduction of area at 1/2 lb Al/ton did not occur on these low sulfur steels. The importance of low-sulfur contents, previously pointed out by Sims and Dahle,[2] Lillieqvist[5] and others, is emphasized by these data, particularly when aluminum deoxidation is practiced.

The effect of aluminum additions and sulfur content on the mechanical properties of cast Cr-Mo-V steel for the same heat treatment is shown in Figs. 2 to 4. The curves shown in Figs. 2 and 3 are the means for all the points plotted where each point is the average of three tensile test values. The points in Fig. 4 are enclosed with an envelope curve. The plots of quality factor (T.S. x R.A.) and reduction of area vs residual aluminum content show a dip in the curves around 1/3 to 2/3 lb per ton of aluminum addition. There was a decrease in both quality factor and reduction of area with increasing aluminum and sulfur contents.

The quality factor and the reduction of area for the 0.024 per cent sulfur casting without aluminum were abnormally low because of a localized segregation of Type II intergranular sulfide inclusions. Ten-

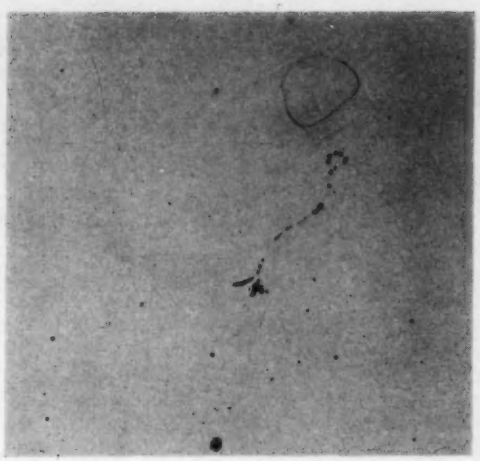


Fig. 11 (Above) — Group IV, Casting 31, shows a fine chain Type II sulphide inclusion. Mag. — 100X.

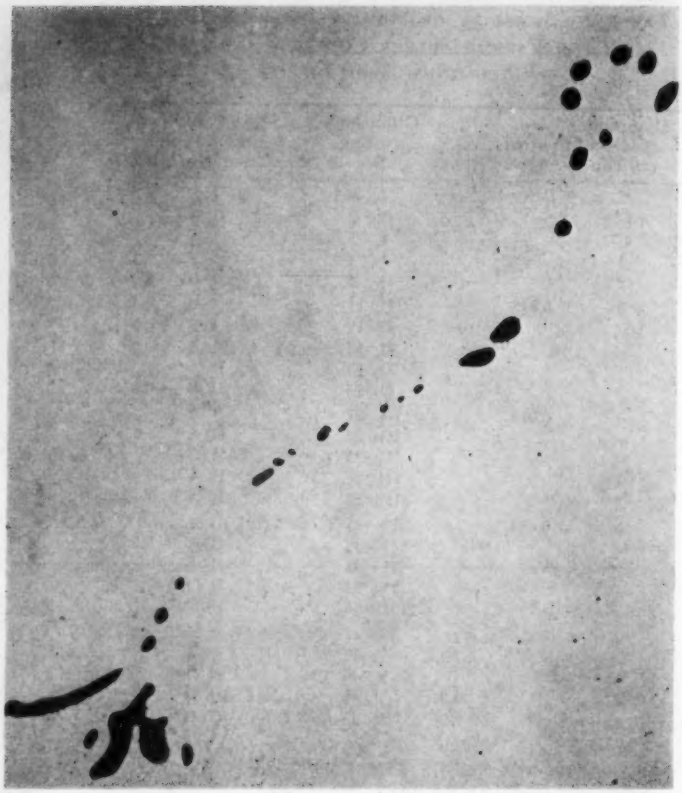


Fig. 12 (Right) — Same as Fig. 8 at higher magnification. Mag. — 500X.

sile tests on retempered material taken from this same casting, adjacent to the original test area, resulted in excellent tensile properties (Table 5). These data are in excellent agreement with that of Sims and Dahle[2] for plain medium-carbon cast steels. The plot of Brinell hardness vs residual aluminum shows a fairly sharp decrease in hardness up to 1 lb per ton of aluminum (.052 per cent) with a leveling off up to 2 lb per ton (Fig. 4).

TABLE 8 — SUMMARY OF MECHANICAL PROPERTIES OF SEVERAL LOW CARBON, BASIC ELECTRIC HEATS, SCHENECTADY STEEL FOUNDRY

Heat	C	Si	Mn	S	P	Cr	Ni	Mo	T.S. psi	Y.S. psi	Elong. % in 2 in.	R.A. %
Aluminum Deoxidized with 1 lb/ton												
2402	.24	.50	.64	.007	—	.12	.15	.12	76,000	42,000	34.0	49.5
2404	.20	.46	.53	.008	—	.09	.11	.16	79,000	39,500	33.0	53.3
2406	.20	.45	.51	.007	—	.10	.08	.17	66,000	36,000	34.0	56.8
2407	.21	.30	.67	.008	—	.09	.10	.12	65,000	35,500	34.0	55.8
2421	.21	.50	.57	.006	—	.09	.10	.17	68,500	36,500	34.5	54.0
2423	.21	.50	.56	.008	—	.10	.16	.14	68,000	37,000	38.5	53.6
2470	.21	.56	.54	.008	—	.05	.15	.04	68,000	37,000	34.0	53.8
2472	.23	.48	.61	.007	—	.10	.13	.14	70,500	40,000	31.0	49.5
2478	.21	.76	.61	.010	—	.08	.13	.10	69,500	39,500	33.0	53.6
2480	.21	.35	.39	.007	—	.24	.70	.17	68,500	36,500	30.5	46.9
			Average	.008							Average	52.7
Aluminum Deoxidized with ½ lb/ton												
1798	.24	.42	.55	.007	—	.16	.19	.26	73,000	44,500	29.0	48.6
1799	.23	.43	.55	.007	—	.10	.18	.11	70,000	38,000	30.5	51.1
1805	.20	.37	.56	.009	—	.05	.08	.11	66,000	36,000	32.5	54.4
1814	.23	.43	.56	.009	—	.13	.10	.11	71,500	41,500	26.0	52.2
1818	.24	.47	.61	.009	—	.17	.15	.22	74,500	39,000	29.5	50.4*
1825	.20	.39	.54	.008	—	.15	.15	.17	66,500	39,500	26.5	52.2
1829	.25	.46	.54	.012	—	.17	.10	.17	73,000	42,500	33.0	50.3*
1830	.21	.45	.53	.008	—	.17	.10	.20	70,500	41,500	32.0	53.8
1839	.21	.43	.57	.011	—	.15	.10	.11	68,000	37,000	35.0	53.8
1847	.23	.45	.61	.010	—	.16	.05	.10	70,500	39,000	31.0	49.8
			Average	.009							Average	51.7

*Fine Type II sulfide inclusions randomly distributed. All other samples showed randomly distributed Type I sulfides.



Fig. 13 — Group I, Casting 6, zero aluminum, upper bainite, ASTM grain size 2-4. Mag. — 100X.

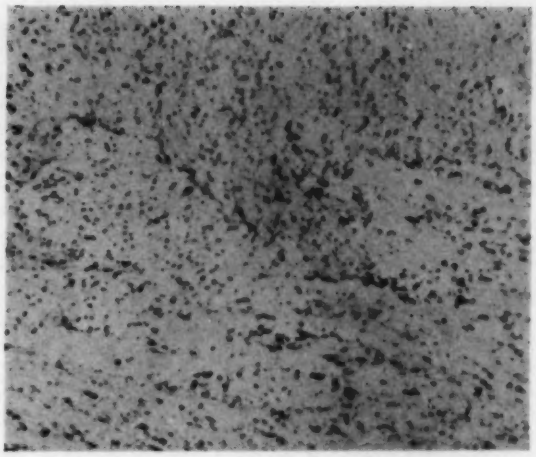


Fig. 14 — Same as Fig. 13 at higher magnification, shows the carbide spacing. Mag. — 1000X.

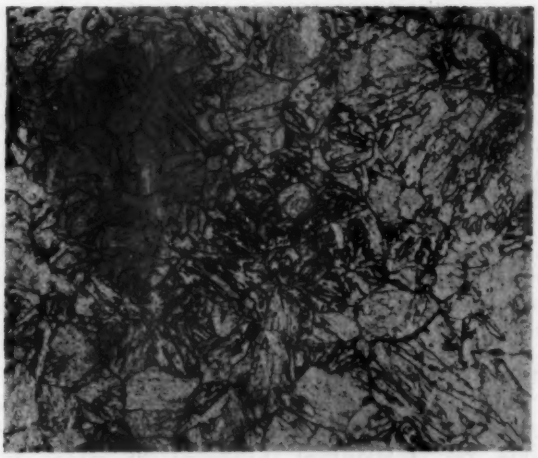


Fig. 15 — Group I, Casting 2, 1/3 lb aluminum per ton, upper bainite, ASTM grain size 2-4. Mag. — 100X.

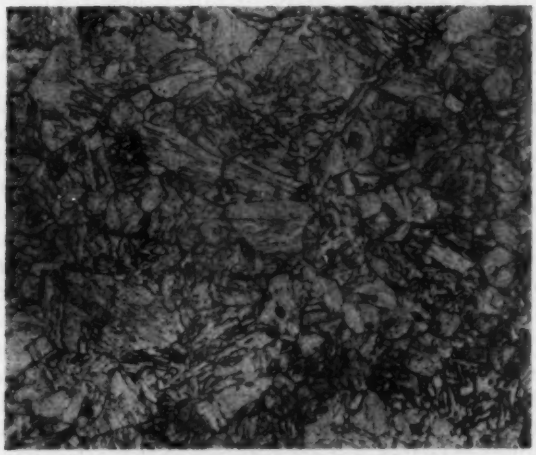


Fig. 16 — Group I, Casting 3, 2/3 lb aluminum per ton, upper bainite, ASTM grain size 3-5. Mag. — 100X.

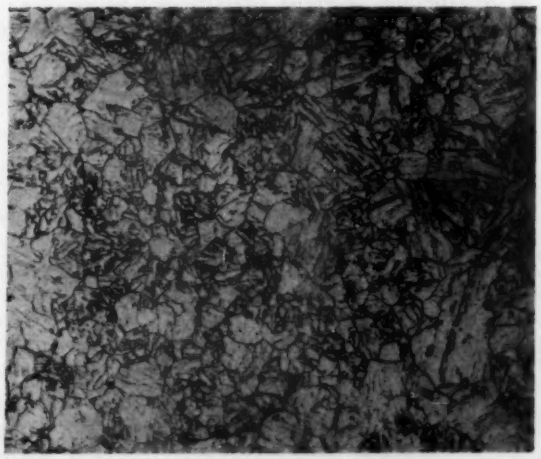


Fig. 17 — Group I, Casting 4, 1 lb aluminum per ton, upper bainite, ASTM grain size 5-6. Mag. — 100X.

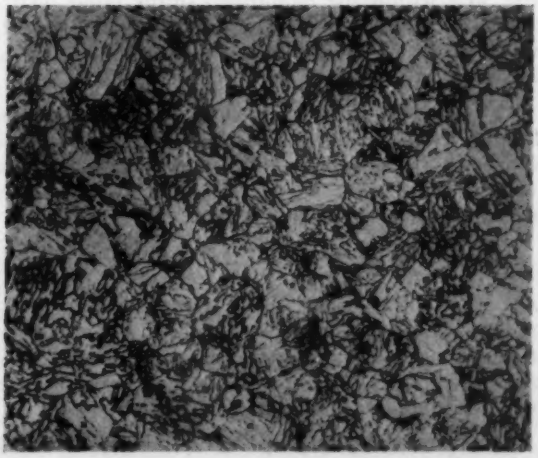


Fig. 18 — Group I, Casting 5, 2 lb aluminum per ton, upper bainite, ASTM grain size 5-6. Mag. — 100X.

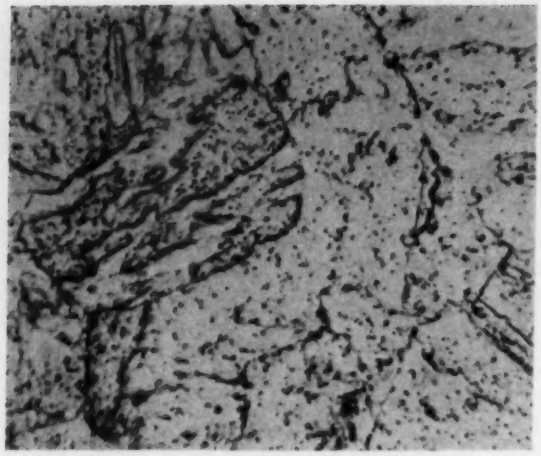


Fig. 19 — Same as Fig. 18 at higher magnification, shows the carbide spacing. Mag. — 1000X.

The causes for changes in mechanical properties with the aluminum additions will be discussed later.

Metallographic Results

Unetched and etched specimens taken from broken tensile bars were examined from each casting. The unetched samples (Table 8) show quite clearly that for the lower sulfur contents (0.011 to 0.018 per cent) the predominant sulfide inclusion type was the rounded globular Type I. In some samples a tendency toward the formation of the Type II sulfide inclusion was noted. For the higher sulfur contents (0.024 - 0.027 per cent) the predominant sulfide inclusion was still the Type I.

It was interesting to note that of all the areas evaluated only one section of one casting contained predominantly Type II sulfide inclusions. This casting (No. 31) did not have any aluminum added, and an adjacent section was entirely free of these inclusions. Additional sulfur determinations were made on material taken from the fracture area of three broken tensile bars from casting Number 31 (Type II area). These results, 0.025, 0.025 and 0.025 per cent sulfur, were well within the established sulfur range (0.024 ± .002) for this casting. The occurrence of Type II sulfide inclusions in some areas of castings will be discussed later.

It was observed that the mean diameter of the Type I sulfide inclusions increased considerably (6 x 10^{-5} to 4 x 10^{-4} in.) as the sulfur range increased from 0.011 - 0.018 per cent to 0.024 - 0.027 per cent for all aluminum contents.

It is interesting to note that the mechanical properties of the cast Cr-Mo-V decrease upon the addition of aluminum (⅓ lb/ton) even in the absence of Type II sulfide inclusions. As there was no change in the grain size or microstructure this suggests the possibility that aluminum per se may have an effect on the mechanical properties.

The etched samples (Figs. 13-19) show quite clearly the effect of the aluminum additions on the ferritic grain size. There was no change in grain size with

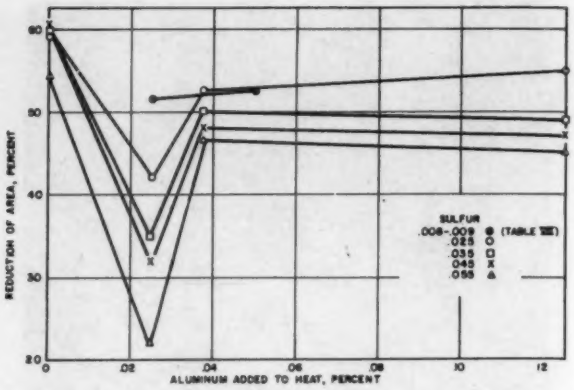


Fig. 20 — Effect of aluminum and sulphur on ductility (After Lillieqvist[5])

the ⅓ pound per ton addition; however, there was a progressive decrease in grain size with the ⅔ and 1 pound per ton additions. The grain size of the material with the 2 pounds per ton addition was the same or slightly coarser than the 1 pound per ton addition.

The work by Baeyertz[7] on low- and medium-carbon cast steel, and by Sims[8] on hot rolled bars of SAE 1020, suggests that under certain conditions sulfur alone is a mild grain refining agent. We were unable to detect any grain refinement in the Cr-Mo-V due to the sulfur per se over the range investigated. In general, the grain refinement observed for cast Cr-Mo-V as a function of aluminum additions and sulfur content was in good agreement with the work of other investigators as summarized by Case and Van-Horn.[9]

Summary of Total Oxygen Content

The total oxygen content of the Cr-Mo-V steel decreased slightly with increasing aluminum additions (Fig. 21). A statistical analysis of the oxygen and aluminum data in Table 1 was made. It was found that the slope of the line (Fig. 21) relating total oxygen to residual aluminum is not different from zero by a statistically significant amount.

It is generally accepted that the solubility of sulfur in molten steel is a function of oxygen content,

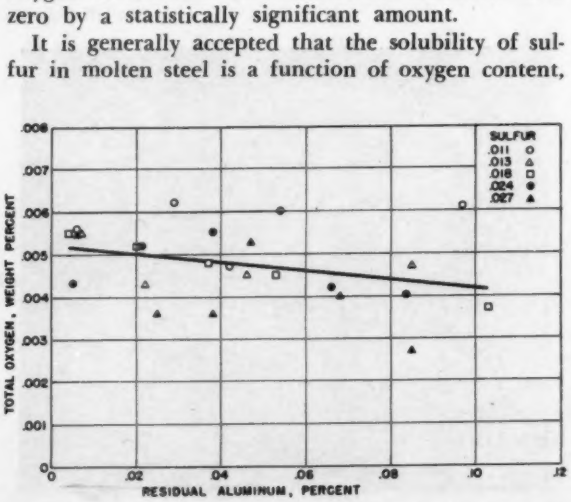


Fig. 21 — Total oxygen contents vs residual aluminum for cast Cr-Mo-V steel.

increasing with decreasing oxygen contents. This has been shown by Hilty and Crafts.[10,11] Since this is true, both the sulfur and oxygen content must be considered when one is discussing sulfide inclusion formation. Unfortunately, sulfide inclusion formation is further complicated by the deoxidation practice used, as shown in Fig. 22 for 29 basic electric production heats of Cr-Mo-V.

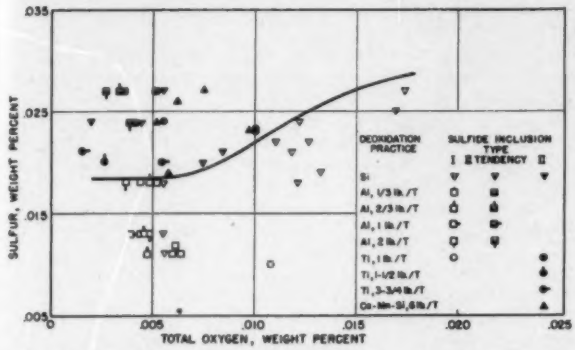


Fig. 22 — Sulphide inclusion type as a function of total oxygen and sulphur content and deoxidation practice for cast Cr-Mo-V steel.

Above the curve, Type II's or a strong tendency to form Type II sulfide inclusions exists whereas below the curve only Type I sulfide inclusions form. The right-hand portion of the curve in Fig. 22 was determined with the following additional data from three silicon deoxidized, acid electric heats of a .20 C, 1 Cr, 1 Mo cast steel containing Type II sulfide inclusions:

Total Oxygen, Weight %	Sulfur, Weight %
0.0212	0.032
0.0125	0.036
0.0132	0.036

These data were not plotted in Fig. 22 since all the other heats investigated contained vanadium. Figure 22 shows quite clearly that sulfide inclusion formation in cast Cr-Mo-V is not only a function of sulfur and oxygen content but the deoxidizing practice as well. This was previously demonstrated for plain medium-carbon cast steel by Gagnebin[3] and others.

Crafts and Hilty[12] have indicated that the melt history as well as deoxidation practice will affect inclusion formation and the resulting mechanical properties. They stress the importance of oxygen content before deoxidation, pointing out the fact that oxygen contents in basic open hearth steel about twice as high as that corresponding to equilibrium with carbon have been observed. This excess oxygen has been called "delta" by other investigators and can be reduced by boiling, the use of a reducing slag in the basic arc furnace or by initial deoxidation with silicomanganese in the basic openhearth.

Crafts and Hilty[12] concluded that initial oxygen and inclusion types are related, in spite of the fact that the final oxygen content after deoxidation appears to be essentially the same regardless of the initial oxygen content. This suggests that means of increasing the oxygen content of the bath prior to the block such as a recarburization practice could promote the formation of a more desirable sulfide inclusion type without any increase in the final oxygen content. Ahles[13] has observed a significant improvement in the quality factor (T.S. x R.A.) of several recarburized heats of 1 Cr, 1 Mo, ¼ V steel.

Since the solubility of sulfur in molten steel decreases with increasing oxygen content one obvious way to minimize Type II sulfide inclusion formation in cast Cr-Mo-V for all types of deoxidation practices is to end up with a total oxygen content of greater than 0.010 per cent (Fig. 22). This would promote the early precipitation of globular sulfide inclusions in lieu of the last to freeze grain boundary type intergranular chain or film type sulfide inclusions. This, however, is not entirely practical since total oxygen contents of 0.010 per cent or greater may be conducive to pinhole and gas formation.

The fact that the aluminum deoxidized cast Cr-Mo-V with low total oxygen and high sulfur contents showed only a strong tendency towards Type II sulfide inclusion formation suggests further that the state of the oxygen rather than its total content is the more important parameter. Gagnebin,[3] Sims and Lillieqvist,[1] Christopher[14] and others discuss the FeO rather than the total oxygen content of the steel in reference to its effect on sulfide inclusion formation.

The FeO hypothesis may be valid for plain carbon cast steels; however, there are enough anomalies in data from various sources regarding the effect of oxygen and sulfur content and deoxidation practice on sulfide inclusion formation to suggest that some other oxygen distribution, for a given sulfur level, should be explored.

An examination of free energy of formation data at steel making temperatures obtained by Chipman[15] and some calculations using these data by Malin and Zeno[16] suggest that for Cr-Mo-V steels the following oxides may form in the following order:

		Free Energy of Formation Δ F at 1600 C (2912 F) in Cal/Mol
1.	Al_2O_3	−139,620
2.	V_2O_3	− 66,650
3.	SiO_2	− 37,330
4.	Cr_2O_3	− 25,510
5.	CO	− 22,590
6.	MnO	− 8,500
7.	FeO	− 5,500
8.	MoO_2	+ 12,000

For alloy steels it is apparent that a further evaluation of oxygen distribution in the melt is necessary to promote the understanding of sulfide inclusion formation as a function of sulfur and oxygen content and deoxidation practice. Along this line Ishizuka[17] has reported that aluminum additions usually lead to an increase in the inclusion content of basic electric steel owing to the formation of oxides on deoxidation.

Type II sulfide inclusions and thickly populated colonies of Type I sulfide inclusions are often found in the last to freeze areas of castings; e.g. under ris-

ers. This is mostly attributable to sulfur segregation; however, these sulfide inclusions have been found in low sulfur (0.010 - 0.020 per cent) areas suggesting further that the mechanism of inclusion formation in alloy steels requires additional study.

The data presented in this study and those of the other investigators show quite clearly that low sulfur contents are extremely desirable irrespective of oxygen content or deoxidation practice; Type II sulfide inclusions cannot form if the basic constituent, (sulfur), is eliminated.

Conclusions

Several castings of cast Cr-Mo-V with sulfur contents of 0.011 to 0.027 per cent and aluminum additions of 0 to 2 lb/ton were examined. The base composition and heat treatment were the same for all castings. The mechanical properties, inclusion types and distribution and microstructure of these castings were determined. The following conclusions are made:

1. For sulfur contents ranging from 0.011 to 0.018 per cent, the addition of 0 to 2 lb of aluminum per ton caused little loss in tensile ductility and quality factor (T.S. x R.A.). The predominant inclusion was the Type I sulfide.

2. For sulfur contents of 0.024 - 0.027 per cent, the addition of 0 to 2 lb of aluminum per ton produced an appreciable loss in tensile ductility and quality factor and a strong tendency towards the formation of Type II sulfide inclusions.

3. For these lower sulfur castings (0.011 to 0.018 per cent), the addition of aluminum (1/3 lb/ton) caused a decrease in the mechanical properties with no change in the grain size or microstructure. This suggests an effect of aluminum per se on the mechanical properties of alloy steels.

4. Type II sulfide inclusion formation in Cr-Mo-V can be minimized for all oxygen levels and many deoxidation practices provided the sulfur contents are below 0.018 per cent.

5. The role of oxygen in sulfide inclusion formation in alloy steels for a given sulfur level and deoxidation practice has not been clearly defined to date.

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GAS POROSITY IN OXYGEN-FREE COPPER CASTINGS

By

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ABSTRACT

Gas porosity in copper castings poured from oxygen-free melts was studied. The porosity studied was that due to gaseous compound formation. An investigation of the oxygen sources after pouring indicated that atmospheric oxygen, water in the molds, and water vapor provided such sources in large enough quantity to cause porosity while monatomic oxygen released from silicate surfaces (silica and bentonite) did not. Carbon dissolved in the copper melt was shown to yield increased porosity when an oxygen source was present after pouring. The addition of two alloying elements was investigated (tin and nickel).

Gas porosity in castings of copper and copper-based alloys may result from the evolution of gas during solidification by either of two mechanisms. These are:

1. A gas dissolved in the melt may precipitate during solidification due to its decrease in solubility.
2. A gaseous compound may be formed in the melt by the reaction of dissolved oxygen with dissolved hydrogen, sulfur or carbon to form H_2O, SO_2, CO or CO_2.

Porosity of the first type has been demonstrated for the case of hydrogen by several investigators.[1-5,16,17] Porosity of the second type due to the reaction between hydrogen and oxygen;[3,15] and sulfur and oxygen[13,15] are also well known. There is, however, little information regarding the role of carbon in reactions of this type. The solubility of carbon in copper has been established[7] and though small (0.003 per cent at 2190 F), is of sufficient magnitude to produce appreciable porosity if conditions are suitable for its conversion to the insoluble gaseous oxides during solidification.

This investigation has dealt with the gaseous compound type of porosity in copper castings poured from oxygen-free melts. Pouring atmospheres and mold materials were investigated as sources of oxygen after pouring. The results of some subsequent work on the effect of dissolved carbon on casting porosity has been included.

Experimental Procedure

Melting and pouring were carried out in a two-compartment furnace (Fig. 1) in which the melting and pouring atmospheres could be varied independently. Each chamber was about 12 in. x 12 in. x 15 in. high and was equipped with a window. The upper (melting) chamber contained the crucible assembly

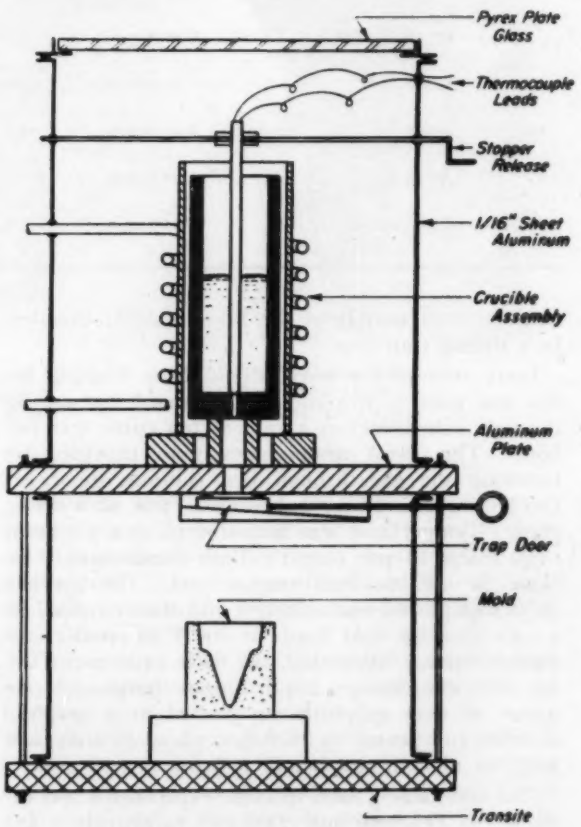


Fig. 1 — Melting and casting apparatus.

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TABLE 1

Casting	Mold	Melting Atmosphere	Pouring Atmosphere	Casting Porosity (% voids)	Radiograph
1	Alumina	Dry Argon	Dry Argon	0.0	Sound
2	Silica	Dry Argon	Dry Argon	0.3	Sound
3	Alumina	Wet Argon	Dry Argon	0.2	Sound
4	Silica	Wet Argon	Dry Argon	1.0	Sound
5	Silica plus 4% Bentonite	Dry Argon	Dry Argon	0.0	Sound
6	Silica plus 4% Bentonite	Wet Argon	Dry Argon	0.0	Sound
7	Silica Flour	Wet Argon	Dry Argon	0.1	Sound
8	Green Silica	Dry Argon	Dry Argon	22	Small and large holes in casting and riser.
9	Silica and 11% Cupric Oxide	Dry Argon	Dry Argon	25.5	Dendrites outlined in riser and casting, small holes in riser and casting, large holes in riser, elongated holes in casting.
10	Silica and 11% Manganese Dioxide	Dry Argon	Dry Argon	23.6	Dendrites outlined in riser and casting, small holes in riser, elongated holes in casting.
11	Silica and 11% Manganese Dioxide	Wet Argon	Dry Argon	35.8	Dendrites outlined in riser and casting, small holes in riser, elongated holes in casting.
12	Silica and 4% Bentone 34	Dry Argon	Dry Argon	0.0	Sound
13	Silica and 4% Bentone 34	Wet Argon	Dry Argon	1.0	Elongated holes in riser and casting.
14	Green Silica and 4% Bentone 34	Wet Argon	Dry Argon	2.6	Elongated hole in casting connecting to shrink hole.
15	Silica	Open Atmosphere	Open Atmosphere	8	Dendrites outlined in casting and riser, small holes in riser.
16	Silica	Wet Argon	Wet Argon	0.1	Sound casting, riser has few small holes near shrink hole.
17	Alumina	Wet Argon	Dry Argon	6.2	Dendrites outlined in casting and riser, small holes in riser.
18	Protected Silica	Wet Argon	Wet Argon	2.2	Elongated holes in casting and riser.

and was separated from the lower (mold) chamber by a sliding trap door.

Inert atmospheres were provided by flushing for 2½ hrs with lamp-grade argon (dried by passing through anhydrone) at a rate of ten cubic feet per hour. The "wet" atmospheres were provided by bubbling the gas through two columns of water. Oxygen used in these experiments was of welding grade. Temperature was measured using a platinum – platinum 13 per cent rhodium thermocouple located in the alundum stopper rod. The cathode copper employed was scrubbed and then oxidized in a core oven for four hours at 800 F to remove oxidizable surface impurities. In each experiment unless otherwise noted, a 2¼-lb copper charge with one ounce of flake graphite was melted in a graphite crucible and heated to 2370 F in about 20 min. and held for ten minutes before pouring.

The test casting used in these experiments was designed by O'Keeffe and Taylor[14] to provide a test piece free from shrinkage porosity. The castings were sandblasted, washed in dilute nitric acid and in water, and sectioned as in Fig. 2. Densities were determined by weighing the indicated section to 0.001 gm in air and in distilled water. Relative void volumes were calculated on the basis of maximum density of 8.92 grams per cubic centimeter. Sections for macro-examination were milled to uniform thickness, polished to 00 grit emery, radiographed, and etched in 1:1 nitric acid.

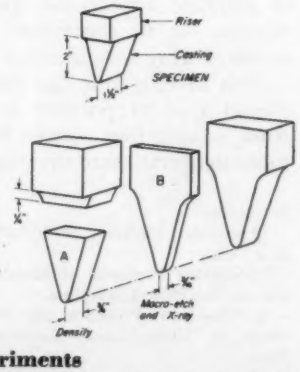
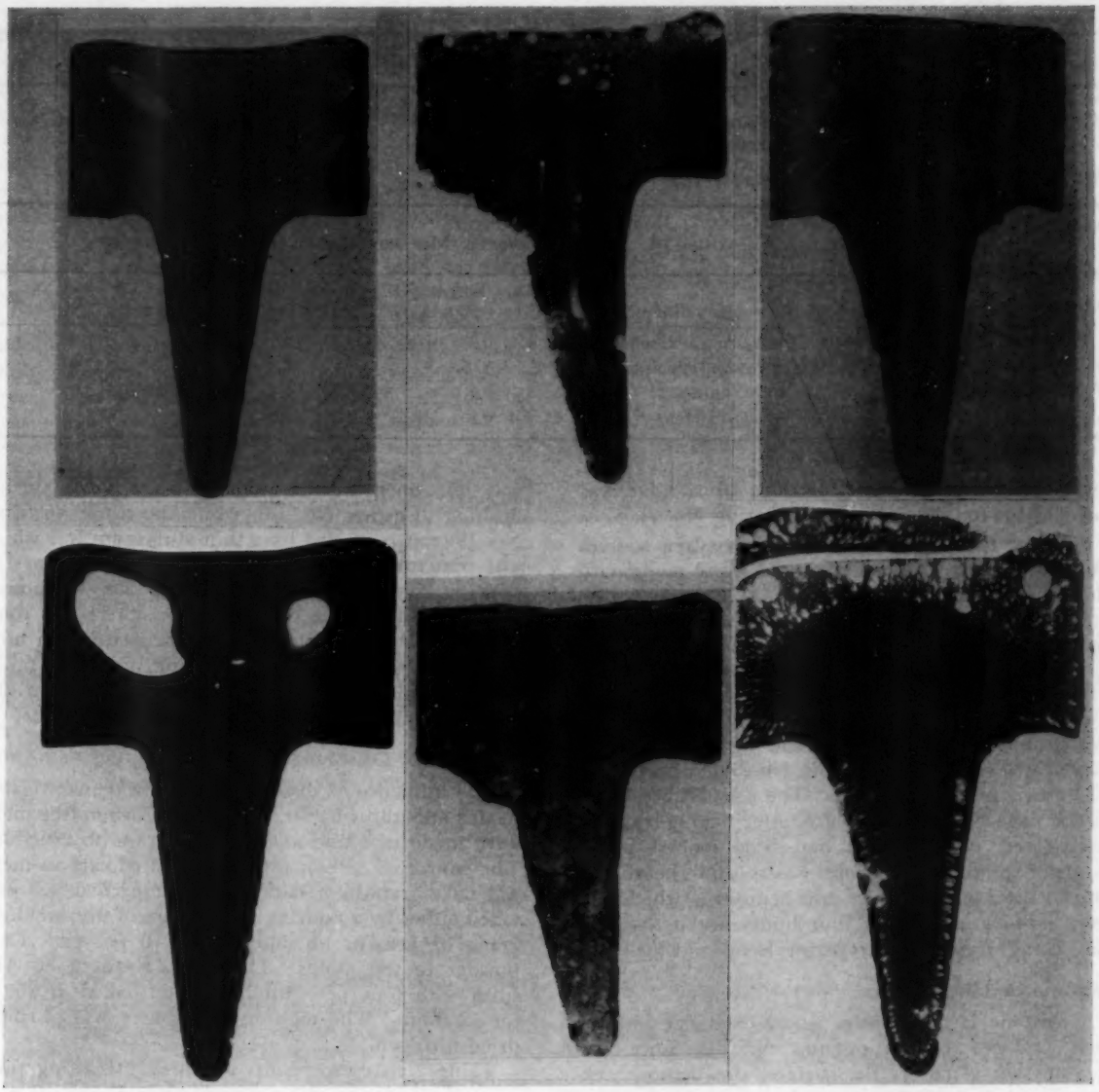


Fig. 2 — Sketch shows sectioning of specimens.

Experiments

The experiments reported first are those dealing with possible sources of oxygen in mold materials and



Mold: Baked Silica
Top: Wet Argon
Bottom: Dry Argon

Fig. 3

Mold: Silica and MnO_2
Top: Wet Argon
Bottom: Dry Argon
Voids: 35.8%

Fig. 4

Mold: Green Silica
Top: Dry Argon
Bottom: Dry Argon

Fig. 5

in the pouring atmosphere. The second set of experiments are the work of Dr. D. Schutt* and deal with the effect of dissolved carbon on casting porosity when poured under conditions permitting access to oxygen.

Silica as an Oxygen Source

Hauser[10,12] has shown that the heating of silicate surfaces and the fracture of silica grains is accompanied by the release of monatomic oxygen in sizable proportions. He also indicates that this phenomenon is absent in the case of alumina surfaces.

Molds of pure silica were prepared by ramming a mixture of washed AFS 80 silica sand with four per cent silica flour and four per cent water. The molds were then baked at 300 F for four hours to remove water. Similar molds were prepared of pure alumina as standards. To insure detection of oxygen in the mold, hydrogen was introduced into the melt by using water in the melting atmosphere. ($H_2O + C = CO + H_2$.)

The castings made in the alumina molds were sound showing that the level of hydrogen was not high enough to cause porosity of type 1 by exceeding the solubility of hydrogen in solid copper (Table 1, Casting No. 1 and 3). The castings made in silica molds were also sound (Casting No. 2 and 4). A radiograph and a macroscopic section are shown in Fig. 3. This showed that the oxygen from the surface of silica grains (of this size) was not sufficient to cause porosity. To investigate the effect of size of

*This work was done at M.I.T. on the apparatus discussed in this paper.

TABLE 2 — POROSITY (AS % VOIDS) OF PURE COPPER MELTED UNDER PURIFIED NITROGEN

Melting Conditions	Pouring Conditions	Pouring Temperature 1920 F	Pouring Temperature 2370 F	
			a	b
No Graphite	Oxygen Pouring atmosphere	0.8	3.0	3.1
Graphite Addition	Oxygen Pouring atmosphere	4.7	7.8	10.8
No Graphite	CuO in Mold	1.5	3.4	4.4
Graphite Addition	CuO in Mold	4.2	9.5	10.0

TABLE 3 — POROSITY (AS % VOIDS) OF ALLOYED COPPER MELTED UNDER PURIFIED NITROGEN

Melting Conditions	Pouring Conditions	Cu - 10% Sn Pouring Temp. 1920 F	Cu - 10% Sn Pouring Temp. 2370 F	Cu - 10% Ni Pouring Temp. 2370 F
No Graphite	Oxygen Pouring atmosphere	0.2	2.5	1.5
Graphite Addition	Oxygen Pouring atmosphere	3.6	5.3	12.9
No Graphite	CuO in Mold	0.2	1.6	3.0
Graphite Addition	CuO in Mold	0.9	4.1	8.5

silica grain, a baked mold of pure silica flour was tested. This too proved sound (Casting No. 7).

To show that molds with positive oxygen sources produce porosity, baked silica molds were made with additions of either 11 per cent CuO (Casting No. 9) or 11 per cent MnO_2 (Casting No. 10 and 11). These molds yielded castings of very high porosity (Fig. 4).

Water as an Oxygen Source

Copper melted and poured under dry argon into a green (unbaked) mold of silica and five per cent water showed gross porosity (Casting No. 8). A section of this casting is shown in Fig. 5. The retention of water on the surface of bentonite clay after baking was also investigated. Copper that had been melted under both wet and dry atmospheres was cast into bentonite-bonded silica molds (4 per cent bentonite) which had been baked at 300 F for four hours. In both cases, the castings were sound (Casting No. 5 and 6).

Bentone Bonding Agents

A bentone is a bentonite whose exchange ion has been replaced by an organic radical. Since this organic radical covers the surface, the bentones exhibit no evolution of monatomic oxygen.[11] Benzene was used as a plasticizer since bentones are organophilic. Two molds made of silica, four per cent bentone and four per cent benzene were baked at 300 F for four hours. The one cast with copper that had been melted under dry argon (Casting No. 12) showed no porosity. The one cast from copper melted under wet argon (Casting No. 13) showed some porosity of the type caused by hydrogen alone (large, thin bubbles). An unbaked bentone-bonded mold yielded a high degree of porosity (Casting No. 14).

Casting Atmosphere as an Oxygen Source

Two sources of oxygen in the casting atmosphere were considered, atmospheric oxygen and water vapor. Castings made in inert molds through pure oxygen (Casting No. 17) and through air (Casting No. 15) both showed a high degree of porosity and some copper oxide content.

Two additional castings were melted and poured under wet argon into baked silica molds. One mold was left open to the pouring atmosphere during flushing (Casting No. 16), while the other (Casting No. 18) was protected by a thin aluminum foil which was punctured by the metal stream as soon as the pouring commenced. While casting No. 16 showed little porosity in the casting section, there was some evidence of gas porosity in the riser section. In both cases these were not perfectly sound, but were much more sound than those exposed to oxygen in the pouring atmosphere.

Influence of Dissolved Carbon

The influence of dissolved carbon was investigated in this apparatus by Dr. D. Schutt. Carbon-free melts were made in alundum-lined clay-graphite crucibles. The source of carbon for the saturated carbon melts was flake graphite. Detection of the carbon was provided either by a pouring atmosphere of dry, welding-grade oxygen, or an addition of 10 per cent CuO powder to the molds. The molds were of 80 AFS silica with four per cent bentonite, baked at 300 F for 12 hours The melting atmosphere was purified, dried nitrogen.

Table 2 lists the results obtained in casting pure copper from two temperatures (2370 F and 1920 F) using carbon-free and carbon-saturated melts.

Table 3 shows similar results for alloys of copper with tin (10 per cent Sn) and with nickel (10 per cent Ni).

Discussion

There is no doubt that monatomic oxygen is released when the silicate surfaces are heated or fractured. However, these experiments demonstrated that this source was not effective in introducing enough oxygen to produce type 2 (compound gas) porosity in pure copper castings. The surfaces of large silica grains, silica flour and bentonite were considered. Casting No. 8 (Fig. 5) clearly showed that water in the mold could produce gross porosity. Soundness of the castings in baked bentonite-bonded molds showed not only that the bentonite did not retain enough water on its surface to be harmful, but also that the evolution of monatomic oxygen from the surface of bentonite was not powerful enough to cause porosity. The bentones do not act as oxygen sources but may

act as hydrogen sources.

Tests on pouring atmospheres showed that both atmospheric oxygen and water vapor could act as oxygen sources, but that oxygen itself was a far more powerful source.

The experiments on dissolved carbon showed that presence of oxygen in either the mold or the pouring atmosphere resulted in porosity in the casting. This porosity was greater by a factor of two or three in cases where the melts had previously been saturated with carbon. This increased porosity was due to a reaction between carbon dissolved in the melt and oxygen picked up during pouring or filling of the mold. The porosity of castings poured from graphite-free melts was probably the result of the reaction between small quantities of hydrogen dissolved in the electrolytic copper melting stock and oxygen sources provided. The reduction in both types of porosity with decrease in pouring temperature was expected as a result of decreased carbon content of the melt and decreased oxygen pickup in the mold at the lower pouring temperature.

The experiments on copper alloys indicated that an addition of ten per cent tin to copper melts resulted in a decreased porosity in both carbon-free and carbon-saturated castings. Dr. Schutt attributed this to a mild deoxidizing effect of the tin which resulted in a decrease in oxygen available in the melt for gas forming reactions. The influence of pouring temperature on porosity can be explained in the same way as for pure copper. No marked effect on porosity was observed to result from the addition of nickel to copper melts. This was surprising in view of the increase in carbon solubility resulting from such additions.[8]

Conclusions

With respect to compound gas formation in copper castings poured from oxygen-free melts:

1. Water in the mold produces porosity.
2. Bentonite, after baking at 300 F for four hours, does not retain sufficient water to produce porosity.
3. The evolution of monatomic oxygen from silica or silicate surfaces is not strong enough to produce porosity.
4. Oxygen in the pouring atmosphere is easily absorbed by the copper poured through and solidifying in it.
5. Water in the pouring atmosphere is a weak oxygen source.
6. Bentone in a mold acts as a source of hydrogen.

With respect to the experiments on the influence of dissolved carbon:

7. Carbon-saturated melts yield castings with greater porosity when an oxygen source is present in the pouring conditions. This indicates that carbon can act like hydrogen or sulfur in compound gas formation.
8. Lower pouring temperatures decrease the above effect due to decreased carbon solubility.
9. Additions of tin decrease porosity of this type presumably due to the deoxidizing effect of tin.
10. A ten per cent addition of nickel does not markedly change the results obtained using pure copper.

Acknowledgment

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FLUXING AND DEOXIDATION TREATMENTS FOR COPPER

By

M. G. Neu* and J. E. Gotheridge**

ABSTRACT

Development of a practical test for the visual estimation of the gas content of high-purity copper melts is described, and this is used in a series of experiments to determine the efficiency of the principal types of fluxing media normally employed commercially, i.e. reducing, neutral and oxidizing.

Effectiveness of three types of deoxidizing agents, viz. phosphor copper, calcium boride and lithium is compared. Results of radiographic, microscopic and chemical analyses are reported. The findings are applied to a duplex deoxidation treatment of high-conductivity copper using phosphor copper to remove the major part of the oxygen present and finally deoxidizing with calcium boride or lithium.

The deoxidizing efficiency of the materials has been determined and the results of the investigation are discussed in the light of present knowledge.

Introduction

The most important application for pure copper either in the wrought or in the cast form is under conditions where a high electrical or thermal conductivity is essential. Although the mechanical properties of pure copper are inferior to all copper-base alloys it possesses with the exception of silver, the greatest electrical conductivity of any material known and this is the property mainly required of high-purity copper castings.

The founding of pure copper involves an extremely specialized technique which even at the present time has only been mastered by relatively few nonferrous founders. The main problem is centered around producing a sound casting having the necessary high electrical (or thermal) properties. Figure 1 shows the effect of impurities on the electrical conductivity[1] of copper. It may be seen that even small amounts of recognized additions – for example phosphorus – drastically reduce the conductivity.

The question of unsoundness may be subdivided into (a) that due to inefficient feeding, i.e., pure shrinkage porosity and (b) that caused by the evolution of gases during the solidification of the casting. The former is a problem encountered in the founding of all types of alloys to a greater or lesser extent and its rectification is purely dependent on adequate feeding of the casting. It is the latter problem – that is the evolution of gases during solidification – which is more difficult to avoid in high conductivity copper and some alloy castings.

Probably the first notable researches of gas unsoundness in pure copper were those carried out by Allen[2] in 1930 and by Allen and Hewitt[3] in 1933 and these are still recognized as standard works on this subject. The major cause of gas unsoundness was found to be due to a combination of two gases, hydrogen and oxygen.

The need for deoxidation had, however, been recognized previously and phosphorus, calcium,[4] boron carbide,[5] beryllium,[6] and lithium,[5,6] have been mentioned in the literature as suitable deoxidants. Since that time a large number of other materials have been added to this list and at present the choice of deoxidants is extremely wide. The first essential of a deoxidant must of course be to produce a sound, oxide-free casting without adversely affecting the conductivity. This implies that the deoxidation products must separate easily from the metal bath,

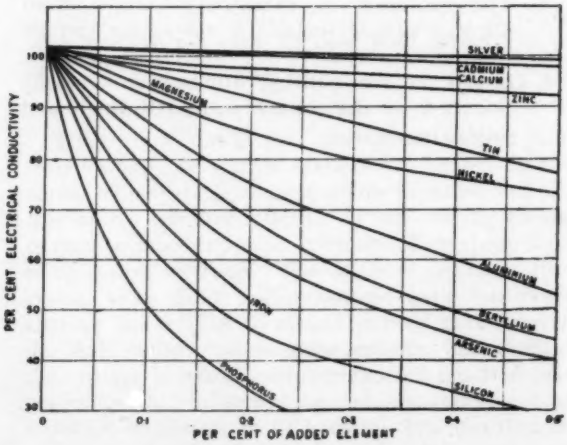


Fig. 1 — Influence of added elements on electrical conductivity of copper.

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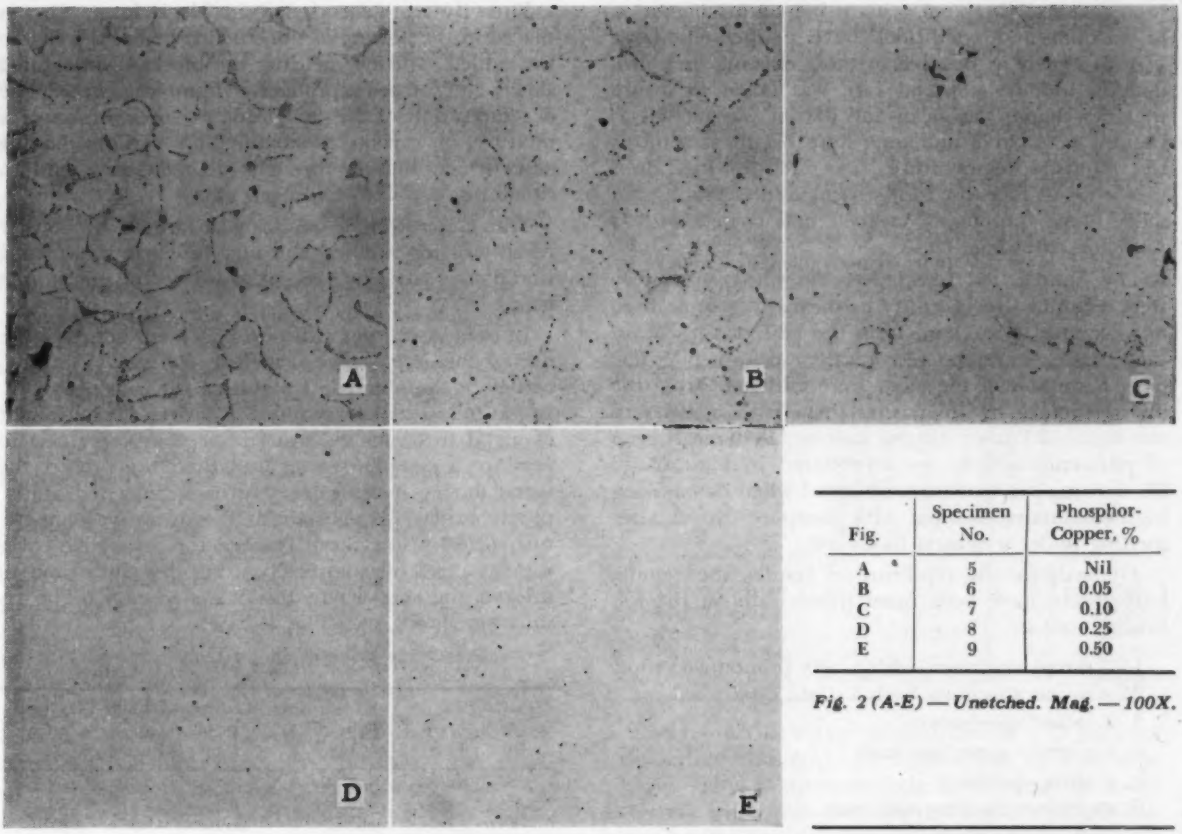


Fig.	Specimen No.	Phosphor-Copper, %
A	5	Nil
B	6	0.05
C	7	0.10
D	8	0.25
E	9	0.50

Fig. 2 (A-E) — Unetched. Mag. — 100X.

and ideally any residual deoxidant remaining should not affect the conductivity.

The subject of melting and fluxing has not received such marked attention. It has generally been accepted that pure copper should be melted in a reducing atmosphere. Charcoal is normally used as a surface cover, in order to keep the oxygen content of the melt to a minimum value, although at least one authority[7] advises that slightly oxidizing conditions should be employed. This method limits the hydrogen content of the melt which is then deoxidized immediately prior to casting.

The object of the work which is described in this paper has been to assess practically, the value of the various types of flux materials which may be used to provide a protective cover during melting, namely those exerting a reducing, a neutral or an oxidizing influence. At the same time the efficiency of the deoxidants which are currently being used in the production of high-conductivity copper castings, such as phosphor copper, calcium boride and lithium has been studied.

Experimental Work

The standard methods for assessing the gas content of copper melts such as microscopic examination, density determinations or x-ray techniques, are too time consuming to permit their use in the production foundry for checking melt quality before casting. A simple test similar to that already used in copper refineries was required so that the quantity of deoxidant to produce a fully deoxidized melt could be determined and leave the minimum amount as a residual impurity in the metal. This is in some ways analogous to the chill wedge test used in cast iron foundries.

In order to obtain some idea of the degree of oxidation of a melt of high conductivity copper, a number of simple shaped specimens, such as cylinders, cones, rectangular blocks have been cast in both sand and chill molds. From observation of the tops of these specimens during solidification, it was found that a cylindrical rod 1-in. diameter, 4-in. long, cast in a green-sand mold gave the clearest indication of the gas content of the melt. A bar solidifying with a pronounced pipe is indicative of a gas-free melt, while a gassy melt is characterized by a bar solidifying with a bulging concave head. This simple test has been used over a lengthy period with considerable success in establishing the amount of deoxidant required and has been used throughout the present work.

Electrolytic copper (99.98 per cent purity) was used throughout the series, new metal being used for each test. Melting was carried out in clay graphite crucibles in a gas-fired injector furnace. Each melt consisted of 50 lb of copper. The procedure adopted throughout was to superheat the melt to 2280 F, remove the flux cover, pour test samples and add deoxidant as required. It was found that with this procedure the temperature fell by 120 F during the time required to complete one series of tests.

It will be noted that no special precautions have

been taken to carry out tests in inert atmospheres or in vacuum and conditions have deliberately been kept as closely as possible to those existing in a production foundry. Special care was taken to insure that the charge was clean and dry; it was melted as rapidly as possible and tests were begun as soon as the requisite temperature (i.e. 2280 F) had been reached. All tools used for stirring and skimming the metal were thoroughly cleaned and coated with a refractory dressing.

The simple test specimens described previously were examined for porosity by means of gamma rays, and sections taken 1 in. from the base of the bar in each case, were microscopically examined. The oxygen content of the melts were estimated from the microstructure by comparing them with a series of standards of known oxygen content. A typical series of photomicrographs are reproduced in Figs. 2A to 2E showing the structures obtained when deoxidizing high conductivity copper with phosphor copper after melting under a neutral flux cover.

To facilitate the reporting of results, the gamma radiographs have been rated numerically in the following way:

1. a sound specimen with a very pronounced pipe.
2. a sound specimen with a slight pipe.
3. a sound specimen.
4. a slightly gassy specimen.
5. a gassy specimen.
6. an extremely gassy specimen.

Figure 3 shows the radiograph of the series of specimens from which the numerical ratings were adopted.

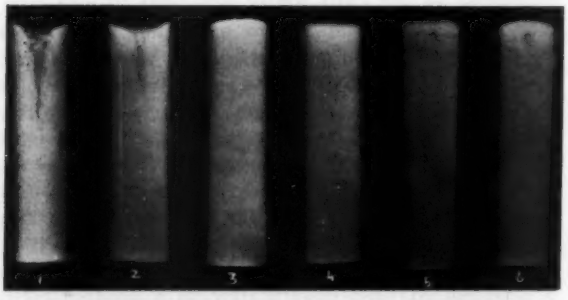


Fig. 3 — Radiograph of series of specimens from which numerical ratings were adopted.

The practical tests may be conveniently subdivided into two parts, namely:

1. The effect of fluxing procedures.
2. The efficiency of various deoxidants.

Methods of Fluxing

Proprietary fluxes were used throughout the investigation having the following characteristics:

Flux A is based primarily on charcoal with fluxing agents which insure a continuous cover throughout the melting operation. The flux which is in powder form, is entirely reducing in its action. It therefore eliminates oxidation *during melting* and due to the continuous cover minimizes the solution of hydrogen.

Flux B is a neutral flux in block form. It was placed in the bottom of the crucible before the charge was added. During melting, the blocks disintegrated slowly and formed a protective liquid slag cover. The decomposition of the flux leads to an evolution of a mixture of carbon monoxide and carbon dioxide gases which bubble through the melt and tend to sweep out dissolved hydrogen gas.

Flux C is a strongly oxidizing surface flux in powder form. During melting the flux fused and formed a liquid slag, covering the entire surface of the molten metal.

In each case, 1 per cent of the flux was added either during the early stages of melting for fluxes A and C or before charging the crucible in the case of flux B. It was found that flux, added as 1 per cent by weight of metal to be melted was sufficient in every case to produce a complete cover and therefore protect the metal during melting from furnace gases and atmospheric oxidation. As soon as the operative temperature (2280 F) had been reached the protective cover was skimmed off completely using dry silica sand to thicken the slag where this was necessary. The results are shown in Table 1.

TABLE 1

Specimen No.	Flux	Type of Flux	Type of Slag	Gamma Graph Rating	Oxygen Content, %
1	None	—	—	6	0.060
2	A	reducing	dry	1	Nil
3	B	neutral	liquid	6	0.040
4	C	oxidizing	liquid	4	0.100

Although the results will be discussed in detail at a later stage it is here of interest to note a number of points. Except for flux C where the oxygen content is sufficiently high to exclude hydrogen from the melt almost completely, the gammagraph ratings and the heads of the test bars are indicative of the oxygen content of the melt. The reducing flux has prevented any oxidation and the resulting test specimen is completely sound. This indicates that solution of hydrogen has been kept to within the limits of solid solubility. However, during pouring of more complex castings, the molten metal is likely to become oxidized due to contact with atmospheric oxygen and it is for this reason that some slight deoxidation is still required even when melting is carried out under a reducing atmosphere.

Throughout melting, when using the block type of flux B, a steady flow of bubbles was observed to break through the liquid cover until some five minutes before pouring, when the block apparently had completely decomposed. The gammagraph shows that even this scavenging action was not enough to limit the hydrogen content of the melt sufficiently to suppress gas evolution in the presence of cuprous oxide. It will be shown that complete deoxidation is necessary before obtaining a sound test specimen.

The shape of the head of the test bar is misleading when highly oxidizing conditions exist, as was the case with flux C.

If these conditions are employed for melting pure

copper it is advisable to fracture the test piece. A completely deoxidized melt is indicated by a pronounced sink in the test bar head and a tough specimen showing metallic salmon pink fracture. If a pipe is observed in the specimen but on fracturing it is found to be brittle and the fracture is dark red in color, the indications are that the melt contains large quantities of oxide and drastic deoxidation is necessary.

Efficiency of Commonly Used Deoxidants

In order to test the efficiency of three of the more commonly used deoxidants the melting procedure was standardized using 1 per cent of flux B in every case. The materials to be investigated were as follows:

1. **Phosphorus.** This was added in the form of a 15 per cent phosphorus copper alloy, graded to allow the maximum rate of solution of the compound and to reduce to a minimum any loss of phosphorus due to fines floating on the surface of the melt. The graded alloy was contained in tubes of pure copper strip and these were plunged to the bottom of the crucible by means of a forked, refractory coated, steel rod. The phosphate slag formed, floated to the surface and was skimmed off before test specimens were taken.

2. **Calcium boride.** This material is essentially CaB_6 but as used in industry it may contain small amounts of calcium oxide. For this reason the compound was carefully dried, weighed and sealed in air tight, pure copper tubes before use. The tubes were plunged into the molten metal in the same manner as described for phosphor copper.

3. **Lithium.** This is an extremely reactive element and is rapidly attacked by the atmosphere. For this reason it is normally stored under oil. For use in these tests it was cleaned of oil, weighed and transferred to sealed tubes as speedily as possible.

The results of the test are summarized in Table 2.

Table 2 — Phosphorus Introduced as 15 Per Cent Phosphor Copper
1% flux B used as cover

Specimen No.	% Phosphor Copper Added (Total)	% Phosphorus Added (Total)	% Residual Oxygen	Gammagraph Rating	% Residual Phosphorus
5	Nil	Nil	0.040	6	Nil
6	0.05	0.008	0.010	5	0.003
7	0.10	0.016	0.010	4	0.007
8	0.25	0.038	Nil	1	0.022
9	0.50	0.075	Nil	1	0.051

The test bar heads are shown in Fig. 4A. Figure 4B shows the radiographs of these specimens. These are typical of the heads and radiographs obtained throughout the present series. As may be seen, the introduction of phosphorus rapidly reduces the oxygen content, a feature which is clearly indicated by the progressive sinks in the test specimens.

An addition of only 0.25 per cent phosphor copper leaves a residual phosphorus content of 0.022 per cent in the melt. This is sufficient to reduce electrical conductivity of the metal to 75 to 80 per cent. When, however, only 0.1 per cent phosphor copper was added, the oxygen content of the metal (0.010 per cent) was still sufficiently high to promote gas porosity. As has been known for some time, oxygen and phosphorus can co-exist in pure copper melts. As may be seen from these results, it is not until a residual phosphorus content of approximately 0.02 per cent has been reached that the oxygen is completely eliminated. It is for this reason that phosphor copper is seldom used as the sole deoxidizing agent for high-conductivity copper. See Table 3.

Table 3 — Calcium Boride Additions
1% flux B used as cover

Specimen No.	% Calcium boride added (Total)	% Calcium added (Total)	% Boron added (Total)	% Residual Oxygen	Gammagraph	% Residual Calcium	% Residual Boron
10	Nil	Nil	Nil	0.050	5	Nil	Nil
11	0.019	0.007	0.012	0.040	4	Nil	0.0005
12	0.038	0.014	0.024	0.040	4	Nil	0.0030
13	0.085	0.032	0.053	0.020	3	Nil	0.0048
14	0.235	0.089	0.146	Trace	2	Nil	0.0063
15	0.478	0.182	0.296	Nil	1	Nil	0.0120

The results show that an addition of 0.49 per cent calcium boride is required to deoxidize the melt completely. This is approximately twice as much as was necessary when phosphor copper was added but calcium boride has the important advantage that the residual deoxidant remaining in the melt is very low and this will have a negligible effect on electrical properties. The deoxidation products were found to separate quickly and easily from the metal bath and the resulting structure was free from non-metallic inclusions.

Calcium was not detected analytically in any of the samples treated with calcium boride and it may therefore be assumed that the element has been removed completely as a deoxidation product. This

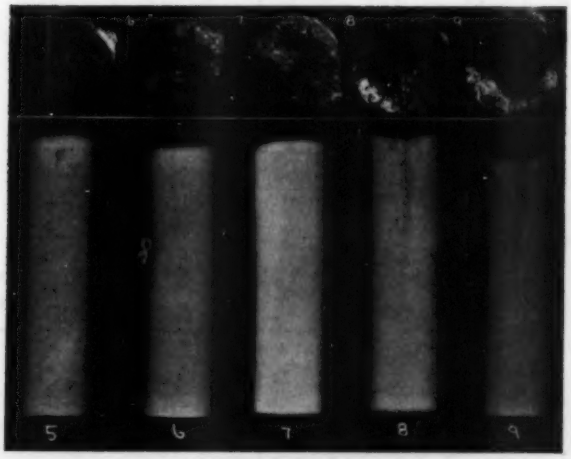


Fig. 4A (Top) — Test bar heads and Fig. 4B (Below) — Radiographs of high conductivity copper deoxidized with increasing amounts of 15% phosphor copper.

result contradicts that of previous workers using elemental calcium, where a tendency for dross inclusions to form at the casting surface has been reported. See Table 4.

TABLE 4 – LITHIUM ADDITIONS
1% flux B used as cover

Specimen No.	% Lithium added (Total)	% Oxygen	Gammagraph Rating	% Residual Lithium
16	Nil	0.040	5	Nil
17	0.005	0.010	4	Nil
18	0.010	Trace	2	0.002
19	0.025	Nil	1	0.005
20	0.050	Nil	1	0.015
21	0.100	Nil	2	0.065
22	0.250	Nil	5	0.150
23	0.500	Nil	6	0.312

The data show that the amount of lithium required to produce a sound specimen is only 0.025 per cent. This is approximately one-tenth of the quantity of phosphor copper and one-twentieth of the quantity of calcium boride necessary to achieve the same result. The residual lithium content of 0.005 per cent is not sufficient to affect the electrical conductivity. There can be no doubt that a controlled addition of lithium is a very effective deoxidant.

When the lithium addition was increased beyond this optimum value, very gassy specimens resulted although no cuprous oxide could be detected in the microstructure. This apparent contradiction is discussed in the next section.

Probably the greatest disadvantage of calcium boride and lithium – particularly of the latter – is the high cost of raw materials. Phosphor copper, on the other hand, is relatively cheap but is not really suitable for the complete deoxidation of high-purity copper due to its adverse effects on electrical conductivity.

A duplex deoxidizing treatment using phosphor copper and calcium boride or lithium is a possible economy and in the following series therefore, phosphor copper was added in quantity sufficient to remove the major part of the oxygen present. The small residual oxygen content insured a negligible residual phosphorus content and therefore minimum reduction of electrical conductivity. The small amount of cuprous oxide remaining was finally reduced with calcium boride or lithium. As with the other tests, the deoxidation products were skimmed off the metal surface before pouring test specimens. See Table 5.

The results show that a considerable saving in the quantity of calcium boride necessary for complete deoxidation can be effected. In this case under the conditions of melting and fluxing employed an addition of 0.15 per cent phosphor copper served to reduce the oxygen content to a trace and yet leave only 0.008 per cent phosphorus in the metal bath. A further 0.10 per cent of calcium boride i.e. only 20 per cent of the amount previously required was sufficient to reduce the remaining oxide and remove all traces of gas porosity. The amounts of impurities remaining have little effect on the electrical properties. See Table 6.

TABLE 6 – PHOSPHORUS AND LITHIUM ADDITIONS
1% flux B used as cover

Specimen No.	Type of addition	Total % added	% Phosphorus added	% Residual Oxygen	Gammagraph rating	% Phosphorus Residual	% Lithium Residual
27	Nil	Nil	Nil	0.040	5	Nil	Nil
28	Phosphor copper	0.15	0.023	0.010	3	0.007	Nil
29	Lithium	0.005	Nil	Nil	1	0.007	Nil
30	Lithium	0.015	Nil	Nil	3	0.006	0.008

The duplex treatment with 0.15 per cent phosphor copper and 0.005 per cent lithium has produced a gas-free melt having an extremely low impurity content. This is a saving of 80 per cent of the lithium required to achieve the same result as a single addition. When the lithium addition was increased to 0.015 per cent however, the gas content of the test specimen was correspondingly raised although no oxide could be detected in the microstructure.

Discussion

Unsoundness, with the exception of that caused by shrinkage, in high conductivity copper may be due to one or both of the following reasons. Porosity caused by dissolved hydrogen alone or that due to a combination of hydrogen and oxygen. The latter gives rise to the phenomenon known as "steam reaction". While it is not correct to say that the former type of unsoundness is never met under normal production conductions it is certain that the majority of instances of gas unsoundness are due to the formation of water vapor in the metal.

TABLE 5 – PHOSPHORUS AND CALCIUM BORIDE ADDITIONS
1% flux B used as cover

Specimen No.	Type of addition	Total % added	% P added	% Ca added	% B added	% Residual Oxygen	Gamma-graph rating	% P	% Ca Residual	% B
24	Nil	Nil	Nil	Nil	Nil	0.04	5	Nil	Nil	Nil
25	Phosphor copper	0.15	0.023	Nil	Nil	Trace	2	0.008	Nil	Nil
26	Calcium boride	0.10	Nil	0.038	0.062	Nil	1	0.008	Nil	0.0050

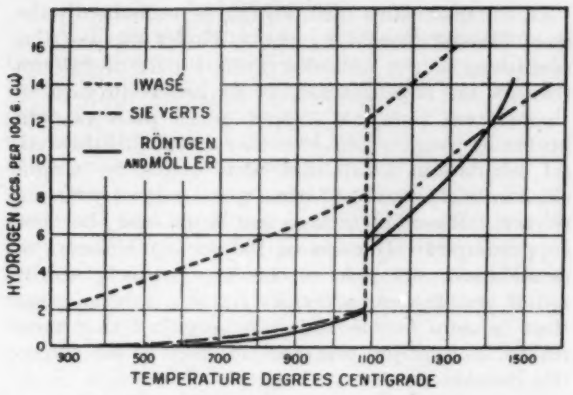


Fig. 5 — Solubility of hydrogen in copper[1].

Unsoundness Due to Hydrogen Alone

Pure oxygen-free copper when molten is capable of dissolving appreciable quantities of hydrogen. The precise amount which may be held in solution is dependent entirely on the temperature, providing of course that the metal is free from oxygen. Figure 5 shows the solubility of hydrogen in copper with respect to temperature.[1] The determination of solubility by Röntgen and Möller[8] are the most recent and are usually regarded as being most accurate. It may be seen that there is a large fall in solubility as the metal solidifies. This is in order of 3.27 ccs (measured at N.T.P.) per 100 grams of copper. It is this difference in solubility of hydrogen between liquid and solid copper which is the cause of unsoundness due to hydrogen alone.

It has been shown that[2] if oxygen-free copper containing hydrogen is allowed to solidify slowly, a sound specimen will result. If however the same metal is chill-cast a specimen having a high porosity is produced due to the entrapment of hydrogen as it is expelled from solution. The voids produced by hydrogen alone are normally large trumpet-shaped cavities easily visible to the naked eye on polished surfaces and very different to the type of porosity normally met in industrial practice.

It is important to differentiate between atomic and molecular hydrogen in copper when considering it as the principal source of gas. It is only the former which is soluble in copper. This is produced by the decomposition of organic substances present in the metal bath or surface oxidation by steam contained in the furnace atmospheres. Molecular hydrogen is said to be completely insoluble. According to Frear[9] under certain circumstances it may be bubbled through the melt as a scavenging gas – although this treatment is not to be recommended generally.

Unsoundness Due to Steam Reaction

Oxygen exists in copper in the form of cuprous oxide. In the liquid state this is completely miscible with copper. On solidification the oxide forms an eutectic with the pure metal containing 0.39 per cent oxygen. This large amount of oxygen is not normally encountered. It is more usual to see the copper – (copper oxide eutectic) – only in the grain boundaries of the structure as already shown in Fig. 2. No oxygen is liberated on solidification and the presence of cuprous oxide alone does not lead to unsoundness of the structure. It does however depress the electrical conductivity of the metal and reduce its ductility.

The quantities of oxygen and hydrogen which may be contained in a given mass of copper are interdependent at any given temperature. Figure 6 shows this equilibrium relationship at 2100 F.

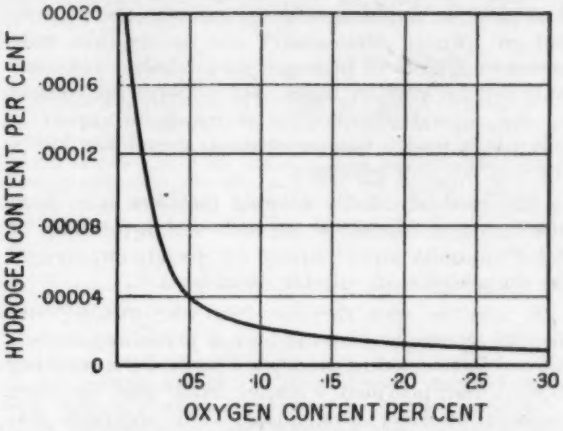


Fig. 6 — Variation of hydrogen content of molten copper at 2100 F with oxygen content[2].

On cooling, these equilibrium conditions are disturbed and lead to the reversible reaction indicated by the equation

$$H_2 + Cu_2O \rightleftarrows 2Cu + H_2O$$

The steam which is formed is considered to be completely insoluble in solid copper. It may be assumed that, at the time of casting, the contents of hydrogen and oxygen are in equilibrium. During freezing the first crystallites of copper to appear will be free from cuprous oxide which therefore enriches the remaining liquid. At the same time hydrogen is expelled from solution due to its decreasing solubility on a falling temperature gradient and the reaction will therefore

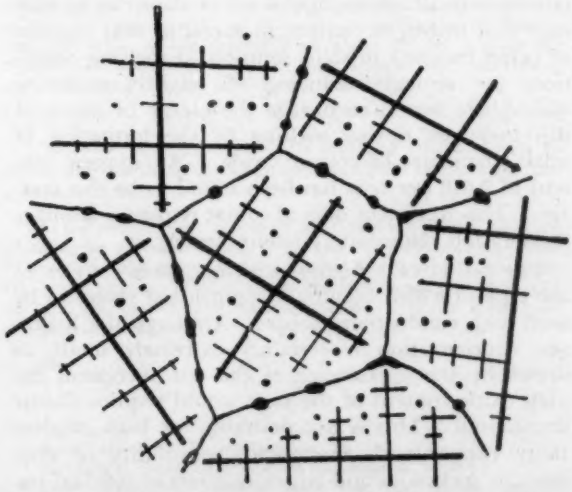


Fig. 7 — Arrangements of blow holes in cast copper.

proceed in the direction of the upper arrow. Steam will thus be liberated in the interstices of the primary dendrites and in those places where the metal solidifies last.[2] This is shown diagrammatically in Fig. 7 and is the fine type of gas unsoundness principally met in industrial practice.

From these observations it becomes apparent that the production of sound copper castings depends on the complete elimination of either hydrogen or oxygen. The former course is usually not possible as hydrogen, in the form of water vapor, is always present in furnace atmospheres and it requires only minute amounts of hydrogen to produce a considerable volume of water vapor. For example the volume of the normal solidification shrinkage of copper is offset by a similar volume of steam containing 1.08 x 10^{-6} wt per cent hydrogen.

The method usually adopted therefore is to keep the hydrogen content of the melt within the limit of solid solubility and eliminate completely any oxygen by the addition of suitable deoxidants.

It may be seen therefore that the melting and fluxing techniques employed are of prime importance. Of the fluxes used in the work described the reducing flux A has provided a sound oxide free specimen without need for deoxidation. This indicates that the hydrogen content has been kept to within the limits of solid solubility but it is almost certain that the metal has a definite hydrogen content below this upper limit.

There is, however, a need for a small amount of residual deoxidant in practice; due to the turbulence of the metal during casting, large surfaces of the metal unavoidably come in contact with atmospheric oxygen. The resulting oxidation may lead during solidification to the steam reaction described. It is for this reason that some form of mild deoxidation treatment is normally employed.

Use of a neutral or slightly oxidizing cover such as provided for example by flux B has much to recommend it particularly when the material making up the metal charge is not clean virgin metal. Use of foundry returns with adhering sand, borings containing a small percentage of oil or damp scrap may lead to a hydrogen content in excess of that capable of being retained in solid solution, if melting conditions are strongly reducing. A slightly oxidizing atmosphere would encourage the escape of much of this hydrogen during melting by the formation of small quantities of copper oxide. (An oxygen content of 0.040 per cent has been found to be the average.) This treatment does of course require a stronger deoxidation immediately before casting.

The action of a highly oxidizing surface cover of the type provided by flux C is much too severe to be used with conductivity copper. Although the hydrogen concentration is certainly extremely small, as shown by the appearance of the test specimen, the high oxide content of the melt would require drastic deoxidation. This is not desirable for high conductivity copper with the greater possibility of non-metallic inclusions and larger amounts of residual impurities, remaining in the melt.

Of the deoxidants used, lithium is undoubtedly the most efficient deoxidizing agent. Under the practical conditions of the tests described, 1 part of lithium removes the same quantity of oxygen as 10 parts of 15 per cent phosphor copper or 20 parts calcium boride by weight. This does not agree with theoretical calculations which show that weight for weight calcium boride has the highest potential for removing oxygen, followed by lithium and 15 per cent phosphor copper respectively assuming 100 per cent efficiency of deoxidation. In practice, however, it is generally found that lithium additions are of a smaller order than calcium boride and it is suggested that these results indicate the practical efficiency of the respective deoxidants.

The use of phosphor copper as the sole deoxidant is not to be recommended due to the deleterious effect of residual phosphorus on electrical conductivity. Neither calcium boride nor lithium have this disadvantage to such a marked extent. The duplex treatment, however, using phosphor copper as an initial deoxidant, followed by either of the other materials to complete the deoxidation treatment, has been shown to be an attractive alternative. This is capable of leaving a negligible amount of phosphorus in the melt and yet permit a considerable reduction in the amount of more expensive alternative deoxidants required.

Deoxidizing Action of Lithium

The results obtained with lithium are of special interest. It is known that apart from the high affinity this element has for oxygen it also exerts a strong degassing influence. This is not due to any scavenging action but to a strong tendency for lithium to form lithium hydride. For example 20 grams of lithium (i.e. 0.01 per cent of a 200-kilogram charge) are capable of removing 200 liters of hydrogen at a temperature in the region of 2200 F.[7]

Lithium hydride which may be formed initially is unstable at these temperatures and with oxygen in the form of copper oxide forms lithium hydroxide according to the following equation:

$$LiH + Cu_2O \rightarrow 2Cu + LiOH$$

Both lithium oxide and lithium hydroxide separate cleanly from the metal bath. If, however, a quantity of lithium is added in excess of that required to reduce the copper oxide present, the remaining elemental lithium will react with water vapor in the atmosphere during pouring and the moisture contained in the mold, to produce lithium oxide and hydrogen. The latter will be dissolved by the still molten copper and expelled from solution during solidification.

This explains the gross gas unsoundness obtained in specimens no. 22 and 23 and the slight porosity shown in specimen no. 30. The addition of lithium therefore must be carefully adjusted to amounts depending on the oxygen content of the melt. With this end in view the simple test described at the beginning of the paper has been found of great value in estimating accurately the precise amount of deoxidant required.

Conclusions

1. With normal fluxing and deoxidation techniques the degree of oxidation of a melt of pure copper may be rapidly assessed by casting a simple test bar in green sand and observing the contour of the head of the bar. By means of this test the necessary quantity of deoxidant can be controlled.

2. Due to the importance of small quantities of dissolved gases in molten copper, rigorous control of melting conditions must be observed. For the production of high conductivity copper castings from virgin metal, a reducing flux has been shown to be the most effective in limiting hydrogen solution and keeping the quantity of deoxidant required to a minimum. If foundry returns, damp scrap, oily borings or similar materials are used as part of the charge, then a neutral or slightly oxidizing protective cover should be employed.

3. Lithium has been shown to be the most efficient deoxidant, followed by 15 per cent phosphor copper and calcium boride under the conditions in which they were used. This result agrees in principle with industrial practice.

The duplex deoxidation treatment has been used successfully and its application can result in a considerable saving in the use of expensive deoxidants.

Acknowledgment

The work described in this paper was carried out in the Research and Development Laboratories of Foundry Services Ltd., Birmingham, England. The authors wish to thank the directors of Foundry Services Ltd., England and Foundry Services, Inc., U.S.A. for their encouragement and permission to publish this paper. The authors' thanks are also due to their many colleagues for their invaluable assistance in compiling the data, particularly Mr. E. J. Jago for the preparation of radiographic and microscopic specimens.

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DISCUSSION

Chairman: R. A. COLTON, Federated Metals Div., American Smelting & Refining Co., Newark, N. J.

Co-Chairman: H. C. AHL, National Bearing Div., American Brake Shoe Co., Meadville, Pa.

Secretary: R. F. SCHMIDT, Ajax Metal Div., H. Kramer & Co., Philadelphia, Pa.

J. J. MARKOWICH:[1] Regarding lithium as a deoxidizer, why do you have gas when too much has been added? Is there any practical means of judging the proper lithium addition?

MR. GOTHERIDGE: Lithium is not only an efficient deoxidizer but will also exert a degassing influence which is not due to a scavenging action but to the tendency for lithium to produce lithium hydride. At normal melting temperatures lithium hydride which may be formed initially, is unstable and with oxygen in the form of copper oxide produces lithium hydroxide. If the quantity of lithium added is in excess of that required to reduce the copper oxide present, the remaining lithium reacts with water vapor in the atmosphere and water in the mold to produce lithium oxide and hydrogen. The hydrogen is dissolved by the molten copper and expelled from solution during solidification.

The use of a simple test piece such as described in the paper will serve as a practical means for controlling the amount of lithium or any deoxidant to be used. If lithium is added to a melt in small quantities, a test piece poured after each addition and the solidification characteristics noted, only sufficient deoxidant to take care of the oxygen present need be added. After a period of operation in this manner in the foundry, a practice may be adopted whereby larger initial additions of deoxidant can be made but this can only be established with experience.

F. J. DANIELS:[2] Is the safest procedure double deoxidation of phosphorus and then lithium?

MR. GOTHERIDGE: It is perhaps incorrect to say that the double deoxidation technique for high-conductivity copper is the safest but it is certainly cheaper than using lithium or calcium boride alone. With a practical control such as described, phosphor-copper in conjunction with either calcium boride or lithium may be used to efficiently deoxidize high-conductivity copper melts without impairing electrical conductivity properties.

J. J. CONNELLY:[3] By theory calcium boride is the best deoxidizer, yet the author's experiments indicate it to be the poorest. Either one or the other must be wrong. What can be done? Using lithium and the gas problem, is it a time relationship?

MR. GOTHERIDGE: Theoretically, weight for weight, calcium boride has the highest potential for removing oxygen. However, in the method of usage described, calcium boride is the least efficient. Theory assumes 100 per cent efficiency of deoxidation and in actual practice, this efficiency is never attained and it is only necessary to refer to everyday foundry practice, where smaller amounts of lithium than calcium boride for instance are used, for verification of this point. Therefore, it is suggested that the results indicate the practical efficiency of the respective deoxidants.

The gas problem with lithium is dependent on the amount of elemental lithium remaining in the melt, and is not directly associated with a time relationship. However, the question arises as to what are the build up effects on remelting. To the best of our knowledge at the present time, about 70 per cent of the residual lithium is lost on the first remelt, with decreasing amounts being lost on subsequent remeltings.

J. E. FRIES, JR.:[4] When you were plunging calcium boride in tubes, did you have a violent action which would make the operation difficult?

MR. GOTHRIDGE: No trouble at all was encountered with the plunging of calcium boride tubes and no noticeable bubbling or reaction took place. If the calcium boride which may contain small amounts of calcium oxide is not thoroughly clean and dry and the copper tube efficiently sealed, a rather violent reaction may occur.

1. Foundry Supt., W. W. Alloys, Inc., Div. of Fansteel Metal Corp., Detroit, Mich.

2. George Sall Metals Co., Philadelphia, Pa.

3. Metallurgist, American Brake Shoe Co., Mahwah, N. J.

4. Chief Metallurgist, American Brake Shoe Co., National Bearing Division, Meadville, Pa.

J. D. ALLEN, JR.:[5] There are two kinds of calcium boride. One is effective and the other is not. Which did the author use, black or white? Also what is your method of analyzing for lithium?

MR. GOTHERIDGE: The calcium boride used in the experiments was CaB_6 which is black, and at this time, I have no knowledge of a white calcium boride. The most suitable method we have found for determining lithium in small amounts without recourse to laborous time-consuming methods, is to utilize a flame photometer.

CHAIRMAN COLTON: What is the effect of residual elements on electrical conductivity? Would the duplex treatment give a minimum electrical conductivity of 85 per cent?

MR. GOTHERIDGE: Figure 1 in the paper shows the effect of various elements on the electrical conductivity of pure copper. In our experience, using the duplex treatment of deoxidation, minimum electrical conductivity of 85 per cent can easily be obtained.

HARRY STERN:[6] If vacuum degassing of copper were carried out nothing would be added to reduce the electrical conductivity. What information do you have on this point?

MR. GOTHERIDGE: By vacuum degassing copper, only hydrogen would be removed from the metal which in itself has not been found to be a major problem in the casting of high conductivity copper. Porosity problems are more commonly found to be due to a combination of oxygen and hydrogen and eliminating the hydrogen would most definitely overcome this problem. However, no matter how closely controlled the melting technique employed, a small amount of oxidation will occur during pouring, and it is because of this that a small amount of deoxidant must be added. Copper oxide present in the melt will reduce the electrical conductivity properties.

CHAIRMAN COLTON: In melting back scrap with new copper, what are the melting conditions and phosphorus buildup regarding lower electrical conductivity?

MR. GOTHERIDGE: In melting back scrap with new copper, a reducing flux will give good results providing this scrap is not excessively dirty or oily. If very dirty or oily scrap is utilized, an oxidizing flux will give best results since this will limit the hydrogen content to within limits of solid solubility.

Using a practical test piece as described, excessive build-up of phosphorus can be avoided. When the metal is molten, a test piece can be poured and the solidification characteristics noted. If the test piece exhibits a level or a slightly sinking surface, only lithium or calcium boride should be used. If the test piece exhibits severe swelling, it is an indication of excessive oxidation, and phosphorus can be utilized for the initial deoxidation purposes. As pointed out, if oxidizing fluxes are used, the test specimen must be fractured before going ahead with deoxidation treatments. In our experience, it is possible to maintain with the duplex deoxidation method, minimum electrical conductivity properties of 85 per cent.

5. Federated Metals Corp., Newark, N. J.
6. U.S. Naval Gun Factory, Washington, D.C.

INORGANIC BINDERS SOLVE SHELL MOLDING PROBLEMS

By

Jose Navarro* and H. F. Taylor**

ABSTRACT

Problems arising from casting low-carbon steel in shell molds were studied from the standpoint of lowering mold cost, improving gating and risering technique, and producing a more acceptable cast surface.

Different types of gates and risers were studied, as was rate of filling the mold cavity. Best results were obtained when mold cavity was filled slowly.

Carbonate additions to sand-resin mixtures improved surface finish of plate castings up to ½ in. thick. Heavier sections required, additionally, that forsterite or zircon be substituted for the silica in the mixture.

Sodium silicate was found to be an inexpensive and acceptable inorganic binder. It also solved the problem of producing an acceptable surface finish. Shells were made by blowing the mixture on a hot pattern and rate of drying was investigated for different pattern temperatures. Other methods for producing shells were suggested.

Use of the shell molding process for production of low-carbon steel castings has been held back because of surface imperfections on the finished product. These surface defects seem to be caused by gases evolved from the organic binder while the surface of the freezing metal was at a critical state of solidification. A previous investigation[1] in which MnO_2 was used as an additive to the sand-resin mixture provided some specific knowledge of the problem and a limited solution. This pointed the way to better solutions and further promising research.

A second deterrent to wider commercialization of the process has been the high cost of the phenol-formaldehyde resin. The basic cost of raw material and production costs are already relatively low in the plastics industry, so it is improbable the base cost of the phenolic resin will be reduced in the near future. Therefore, a solution to the unfavorable economics would be development of a cheaper binder (preferably inorganic) which would act similarly to phenol resins although the mechanism of bonding might be completely different. Inorganic binders are inherently cheaper than organic types and have the added great advantage that when in contact with molten metal there is practically no evolution of gas. Little work has been published on low "setting" temperature inorganic binders and their development offers new promise for the shell molding process.

Data on the effects of gating and risering on castings made in shell molds is contained in a very few papers,[2,3] these are generally descriptive and non-specific, and for low-carbon steel castings the only published work available is by Powell.[1]

Accordingly, a research program was planned along the following lines:

a. A study of gating and risering shell molds to determine the effect of these variables on the surface condition of steel castings.

b. An improvement of the surface finish of plain and low alloy steel castings by altering mold composition and other casting variables.

c. Development of a method to obtain a reduction in resin consumption or of an acceptable substitute for resin.

Experimental Procedure

In all cases shell molds were made by the following procedure: The aluminum pattern was preheated in a core oven for about 1 hour at 600 F. After the first shell was made the temperature of the oven was increased to 700 F to shorten the production cycle.

A hand-operated dump box mounted on a frame and free to pivot about a horizontal axis was used to invest the pattern. The pattern was simply clamped upside down over the box for investing. Stripping the shell from the pattern after curing was accomplished by pushing a plate against ejector pins mounted on the base plate.

The shell molding mixtures were made in different mullers according to requirements. When quantities of about 150 pounds were needed, the batches of sand and resin were mulled for 4 minutes in a barrel-type mixer. When smaller amounts were required and chemical additives used, a heavy-wheel laboratory muller was used and the procedure was as follows: the sand and dust suppressant (0.1 per cent of kerosene) were mulled for 10 minutes to achieve a good distribution of the kerosene on the surface of the

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NOTE: Patents have been applied for use of MnO_2 and forsterite by Steel Founders' Society of America, Terminal Tower Bldg., Cleveland, Ohio under whose sponsorship the initial work on these materials was carried out; patents have also been applied for use of $CaCO_3$ by the Research Corp., 405 Lexington Ave., New York 17, N. Y.

grains of sand and to avoid formation of lumps of sand. Resin was then added and mulled with the sand for 4 minutes. Finally the chemical (when used) was added and the mixture mulled for another 4 minutes. Belts and clamps were used to hold the shell molds together, the molds being placed in flasks and backed up with gravel as support for pouring.

All heats were made from Armco iron with additions of carbon, silicon, and manganese to achieve the desired composition. Chemical analyses were made as required to keep compositions within acceptable limits. The nominal analysis was 0.18 to 0.22 per cent C, 0.3 to 0.5 per cent Si and 0.6 to 0.7 per cent Mn.

A platinum-platinum 13 per cent rhodium industrial thermocouple was used for measuring metal temperatures. Throughout the experiments the initial pouring temperature was controlled at about 2950 F except when pouring temperature was studied. A drop of 25 F was detected for each casting when successive slab castings were poured from the same heat.

Effect of Pouring Temperature, Gating and Risering

Among the variables included in this study were: (a) pouring temperature, (b) step or finger gating, (c) top gating, (d) bottom gating, (e) top risering, (f) side risering and (g) rate of filling the mold.

The patterns used for the test casting were two slabs 7 in. x 5 in. x 1 in. and 7 in. x 5 in. x ½ in. No taper was provided on the slabs, in spite of the fact that this could lead to fine center-line shrinkage in the casting, since it was decided to eliminate any influence due to the variation of section. The sprue system (runners and gates) selected was based on the work of Johnson, Bishop and Pellini[4] and a sketch of the system is shown in Fig. 1a. A runner was chosen to provide a reserve flow so the metal would be directed upward in a vertical runner (a). Pressurization of the spue is thus permissible and a choke (b) was introduced between the sprue and the runner to regulate the rate of filling the mold. By changing the choke and removing certain ingates, all variables except side risering could be studied using the same pattern.

For accuracy all experiments described in the following paragraphs were performed at least twice.

Effects of Pouring Temperature

Effect of pouring temperature was investigated first to determine if it would influence results obtained from other variables of the test program. These experiments were carried out under standard conditions using both 1-in. and ½-in. slab castings, step-gated with a 1-in. choke in New Jersey silica sand of 60 AFS grain fineness bonded with 4 per cent phenolic resin. Results indicated that pouring temperature has slight influence on surface conditions of both 1-in. and ½-in. slabs.

Experiments on Gating

Gating of 1-in. Slabs. The conditions of experiments performed on 1-in. slabs were the same as above except that the gating was varied and temperature held constant. The results are shown in Fig. 2. Each group of three castings were poured from the same heat.

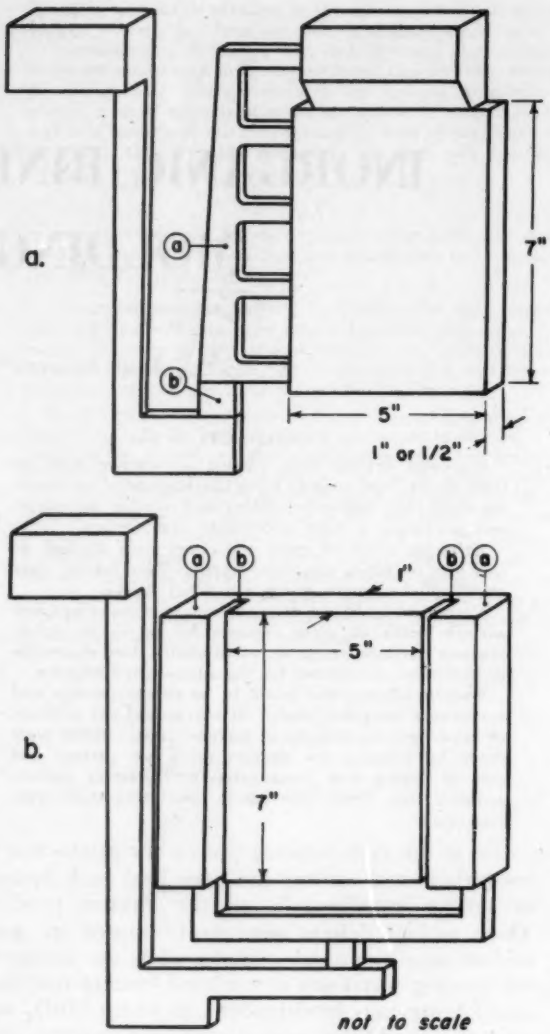


Fig. 1 — Sketches of test castings; (a) step gated, (b) side risered.

It will be observed that the rate of filling the mold has some significance on the gating of the castings. When a 1-in. choke is used the metal fills the mold more rapidly and the gating has almost no influence on the surface finish of the casting. On the other hand, when a ½-in. choke is used a sharp difference on the surface of the casting was detected for different gating systems; the best results were achieved when top gating was used.

Gating of ½-in. Slabs. The same experiments described above were performed on ½-in. slabs. From the results obtained it seemed that neither the rate of filling the mold nor the gating system had much influence on the surface finish of the ½-in. slabs. This was probably because of the relatively small volume of the casting, and the fact that smallest gate chosen was adequate to fill the mold relatively rapidly.

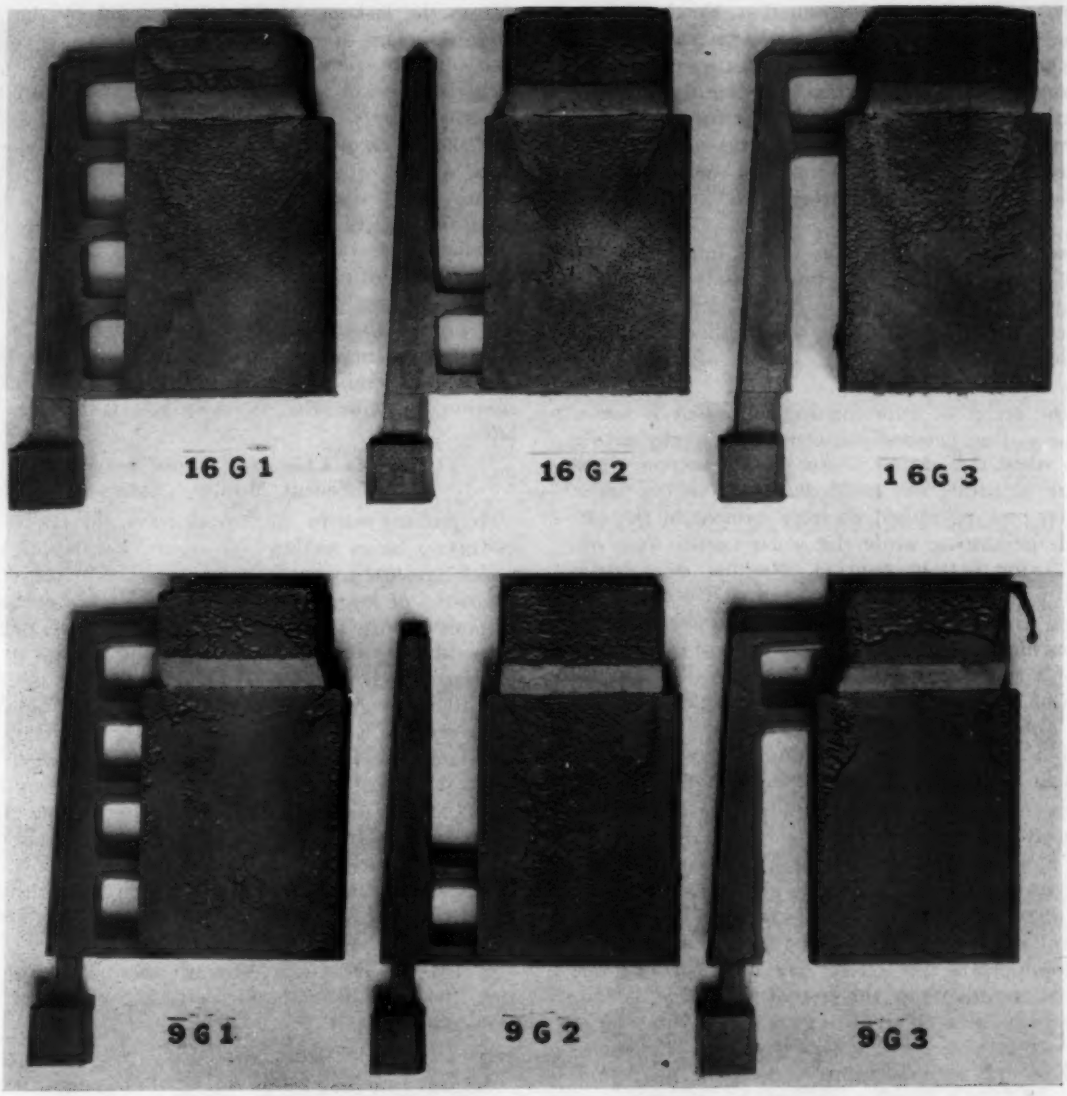


Fig. 2 — Effect of different gating systems on 1-in. slabs using: upper row 1-in. choke, lower row ½-in. choke.

Experiments on Risering

Side risering was studied next, together with the rate of filling the mold. The object was to determine if gating through the side risers would be more efficient in eliminating surface defects, and to determine how important ferrostatic pressure might be. A sketch of the system used is shown in Fig. 1b. The two side risers (a) are connected to the slab through web (b). Although this arrangement is not very satisfactory from the point of view of feed metal requirements, it served to prevent any possible influence of the risers on the rate of skin formation of the slabs in their proximity. The slab was filled through the risers. An interchangeable choke (c) was provided to regulate the rate of filling the mold.

Castings resulting from this experiment were comparable to those made with top risering and indicated that side risering did not improve casting surface.

Discussion of the Results on Gating and Risering

The above study was not concerned with gating and risering from the standpoint of turbulence during filling the mold or of internal soundness of the castings. Since occurrence of surface defects is a major obstacle to the production of steel castings in shell molds, efforts were concentrated upon determining to what extent these surface blows could be influenced by drastic changes in gating and risering.

A difficulty in this study was the lack of reproducibility in shape and location of the blows, especially in 1-in. slabs. Sometimes they appeared as round cavities trapped almost completely in the surface of the casting, while at other times they appeared as rough surface depressions. On ½-in. slabs the defects were always of the second type. A further difference was that on 1-in. slabs the defects tended to be, in most cases, on the upper half of the casting, while on ½-in. slabs defects were spread over entire casting.

It was suggested in the introduction that the defects seemed to be caused by the back pressure of gas on the solidifying metal. Therefore, it was intended, by changing the gating system, to alter the thermal pattern and to vary the rate of skin formation. No appreciable change was detected on ½-in. slabs, whereas on 1-in. slabs improvement of the surface was achieved by using top pouring with slow rates of filling the mold.

It seems, therefore, that as far as improving the surface quality of the casting under study, gating has a limited influence. This influence depends upon two factors: (1) rate of filling the mold, and (2) section thickness of the casting, and so is sensitive to casting size and shape.

In the series of experiments performed it seems that the casting becomes more sensitive to the gating system when the section of the casting increases and the rate of filling the mold decreases. In the latter case, top pouring caused an improvement of the surface of the casting while the worst results were obtained using bottom pouring. Regular step gating showed a surface condition between that obtained by top and bottom gating. However, it is the authors' opinion that the improving effect observed in casting 9G3 of Fig. 2 was due to the flushing effect of the stream of the metal rather than to the gating system itself. Thus in casting 9G2 (Fig. 2), the metal filled the mold cavity smoothly and the bubbles of the gas remained trapped by the solidifying metal.

These findings seem to contradict the experiences of Powell and others, in which a very rapid filling of the mold through the top was a necessary requisite to the production of reasonably good castings. However, these two improving effects may respond to different causes. As pointed out before, when using slow top pouring the effect may be a flushing one while in the other case the improvement may be due to the full effect of ferrostatic pressure. This latter effect will be discussed later. An important point to be taken into consideration is that top pouring is nearly always associated with turbulence which may detract somewhat from the usefulness of this method.

The appearance of defects on the upper part of 1-in. slabs suggested two possibilities: (1) the retarding effect on the rate of skin formation by the proximity of the riser, and (2) the smaller ferrostatic pressure in the upper part of the casting.

To check these possibilities the side risering system described above was tried. In spite of the absence of a top riser the defects were concentrated on the upper half of the slabs, indicating that the proximity of the riser, when the top riser is used, does not account for the location of the defects in that area.

On the other hand, the decrease of ferrostatic pressure did not seem to impair, to a great extent, the surface of the castings. An explanation of this may lie in the fact that when step gating was used, the skin of the metal solidified gradually as the metal filled the mold, the ferrostatic pressure failing to exert its full effect. However, on this basis it is difficult to explain why bottom poured castings, which must have been subjected to a considerable ferrostatic pressure due to the lack of solidification caused by the continuous stream of metal passing through the bottom part, did not yield best results. On ½-in. slabs the increased rate of solidification possibly overshadows all other effects.

It should be noted that the appearance of the defects is not caused by lack of ferrostatic pressure alone; other effects such as the rate of evolution of gases and rate of skin formation of the castings must be taken into consideration as well.

The point of major importance shown in this work is that varying gates and risers does not provide a full solution to satisfactory shell molding of plain carbon steel castings. There is evidence, however, these variables do minimize the problems when used in conjunction with other improving variables discussed later.

Use of Chemical Additives and Different Molding Sands

As pointed out in the introduction, the occurrence of surface blows on low-carbon steel castings in shell molds appears to be caused by the solidifying characteristics of low-carbon steel and the large evolution of gases coming from the decomposition of the phenol-formaldehyde resin. Accordingly, those parts of the castings which, because of their design or position, solidify while gas is being evolved will show blows on their surface, whereas the parts which solidify before any large quantity of the resin is decomposed will show no defects. The above consideration indicates the desirability of delineating and altering the effects of skin formation and of gas evolution separately. This can be achieved as follows:

1. Use of chemical additives which will influence the decomposition of the resin, and act as accelerators or inhibitors to the evolution of gases.

2. Use of sands with different thermal conductivities that will increase or decrease the rate of skin formation.

Use of Chemical Additives

Information on the products of decomposition of resins is limited. According to Bolton, Cullingworth, Ghob and Cobb,[5] the decomposition of phenol-formaldehyde resin under vacuum follows three steps:

1. Up to 900 F — Breakdown of the phenol-formaldehyde resin leaving solid and liquid residues. Hexagonal nuclei of carbon start to form.
2. 900 to 1000 F — Reaction between the residues:
   $R_1H + R_2CH_3 \ldots R_1R_2 + CH_4$ (hydrocarbons)
   $R_1OH + R_2H \ldots R_1R_2 + H_2O$ (maximum amount of water evolved)
3. From 1100 F up — Formation of water gas: $H_2O + C$ (hexagonal nucleus) $\ldots CO + H_2$ The amount of $CO + H_2$ tends to be constant.

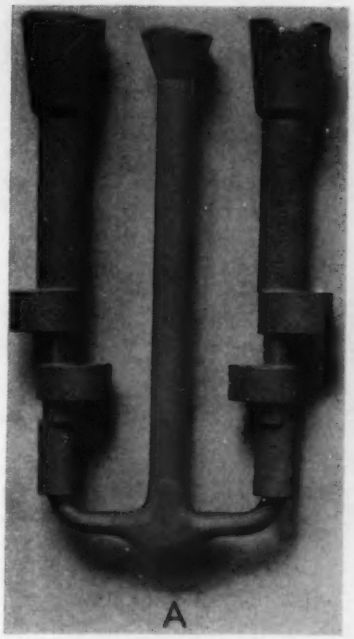


Fig. 3 — Camshaft casting made in shell mold without chemical additives.

Conditions prevailing in the mold during the pouring of metal are very similar to those referred to above.

The manner by which the whole process may be influenced by the chemical additives is not well understood. Previous work by Powell[1] has shown that some oxidizing substances, especially MnO_2, were able to improve the surface finish of some castings, which added support to the belief that the whole process could be modified by oxidizing effects.

The first series of experiments were directed to check previous results with different grades of MnO_2. A camshaft pattern was used, and the procedure for making shell molds was described before (Fig. 3, 4). It was surprising to find that the cheapest brand of MnO_2 is not effective in preventing the surface defects. This suggests that MnO_2 may have a catalytic effect, besides an oxidizing effect, the former being possibly suppressed by the presence of impurities.

TABLE 1 — EFFECT OF DIFFERENT GRADES OF MnO_2 ON SURFACE QUALITY OF CASTINGS

80 AFS Silica Sand, 4% Resin, Pouring Temp. 2950 F

Chemicals Used	Comments
2% MnO_2 (c.p. granular grain)	Remarkable improvement; almost no defects.
2% MnO_2 (Company A tech. grade)	Remarkable improvement; almost no defects.
2% MnO_2 (Company B ind. grade)	No improvement detected.
2% MnO_2 (Company C ind. grade)	Very rough surface finish.

The following series of experiments were made to find new chemicals which might perform more favorably in efficiency and price than MnO_2. The results are shown in Table 2.

From this series of experiments it may be seen that some very strong oxidizing agents did not improve the surface finish of the casting while other inert oxides such as Al_2O_3 improved the surface markedly. PbO_2 improved the casting whether used alone or together in small amounts with MnO_2 (ind.).

It is interesting to note that both activated alumina and PbO_2 are used as catalysts in chemical reactions. This, together with the fact that many strong oxidizing agents did not improve the surface of the cast-

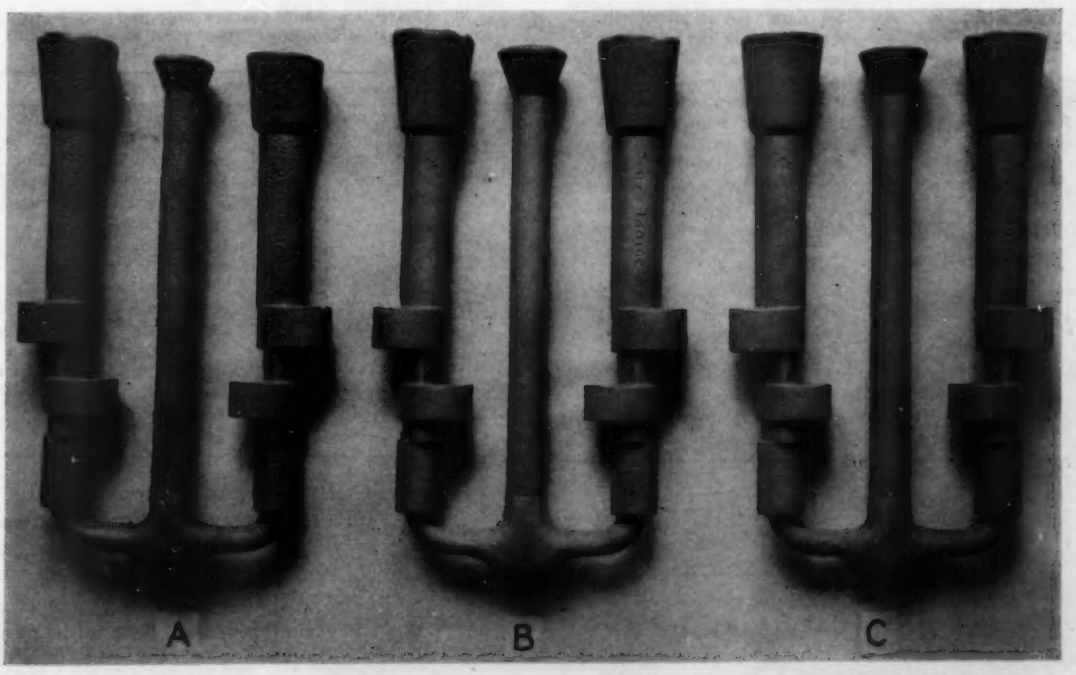


Fig. 4 — Effect of MnO_2 on castings: (A) 2% MnO_2, Co. C ind.; (B) 2% MnO_2, Co. A tech; (C) 2% MnO_2 c.p.

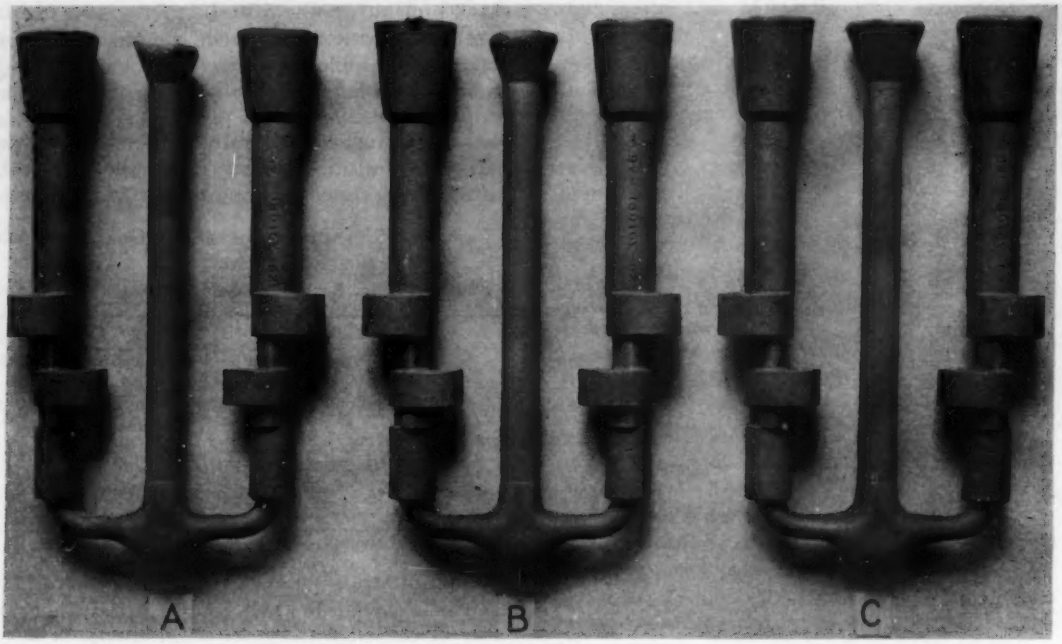


Fig. 5 — Effect of carbonates on castings: (A) 2% Na_2CO_3 c.p.; (B) 2% $CaCO_3$ c.p.; (C) 2% $MgCO_3$ c.p.

TABLE 2 — EFFECT OF DIFFERENT CHEMICALS ON SURFACE QUALITY OF STEEL CASTINGS

70 AFS Silica Sand (except as noted), 4% Resin, Pouring Temp. 2950 F

Chemical Used	Comments
2% KNO_3 (c.p.)	Casting worse than the reference made with plain resin.
2% $K_2Cr_2O_7$ (c.p.)	Casting worse than reference.
2% $PbCrO_4$ (tech.)	No significant improvement.
2% Pb_3O_4 (tech.)	Improvement over the reference casting.
½% PbO_2 (tech.)	No significant improvement.
2% PbO_2 (tech.)	Remarkable improvement, smooth surface; almost no defects.
4% PbO_2 (c.p.)	Almost a perfect casting.
2% MnO_2 (ind. Company C)+ 1% PbO_2 (tech.)	Remarkable improvement over the casting made only with MnO_2 (Conneaut) Fig. 12, 3A.
2% $CaSO_3$ (tech.)	Good improvement over the reference.
5% MgO (c.p.) (over 400 mesh)	Improvement over the original.
None	Rough surface.
None	100 AFS Zircon. Slightly defective casting.
2% $CaSO_3$	100 AFS Zircon. Very good casting.
½% PbO_2 (tech.)	100 AFS Zircon. Excellent casting.
2% Activated Alumina (c.p.)	Very good casting.
2% Ferric Oxide (c.p.)	Very poor result.

ings, suggests that the improvement obtained with several chemicals is due mainly to their catalytic effect rather than to a chemical effect. This theory could explain why some grades of MnO_2 are not effective due, possibly, to the presence of catalytic poisoners.

Some information was obtained about the products of carbonization of the phenol-formaldehyde type resins and the effect of certain carbonates on the rate of combustion.[5] According to these findings it seems that some metallic oxides added to the resin as carbonates greatly increase the rate of combustion. Therefore, the experiments listed in Table 3 were performed. The results achieved are important because the carbonates are much cheaper than MnO_2 of technical grade. See Fig. 5 for effects of carbonates.

TABLE 3 — EFFECT OF VARIOUS CARBONATES ON SURFACE QUALITY OF CASTINGS

80 AFS Silica Sand (except as noted), 5% Resin, Pouring Temp. 2950 F

Chemical Used	Comments
2% Na_2CO_3 (c.p.)	Very good improvement.
2% Na_2CO_3 (ind.)	Although the casting was not good because the grading of the Na_2CO_3 was too coarse, all signs of surface blows practically disappeared.
2% Na_2CO_3	140 AFS Silica Sand. Very good casting.
2% $NaHCO_3$ (c.p.)	Very good casting.
1% K_2CO_3 (c.p.)	No sign of surface blows. Defects due to the coarse size of the K_2CO_3 grains.
2% $CaCO_3$ (c.p.)	Very good casting. Repeated 8 times with the same result.
2% $MgCO_3$ (c.p.)	Very good casting.
2% $BaCO_3$ (c.p.)	Very poor result.

Some experiments were carried out with the hope of throwing some light on the mechanism by which the carbonates gave good results. From the use of a carbonate the following effects may result.

1. Decomposition of the carbonate at the temperatures involved, giving place to a metallic oxide to catalyze the decomposition of the resin under the actual conditions.

2. The CO_2 coming from the breakdown of the carbonates, at the temperatures involved may act as a strong oxidizing agent according to the reaction

$$CO_2 + C \ldots 2\ CO$$

The first series of experiments were oriented to determine the effects of basic oxides by themselves. The oxides were added as peroxides or hydroxides for convenience.

TABLE 4 — EFFECT OF METALLIC OXIDES
80 AFS Silica Sand, 5% Resin, Pouring Temp. 2950 F

Chemical Used	Comments
2% $Ca(OH)_2$	Slight improvement.
2% BaO_2	Very poor result.
1% Na_2O_2	Marked improvement.

The results are shown in Table 4 and show that the oxides alone are not responsible for the effect of the carbonates. It is interesting that $BaCO_3$ in the previous table gave a very poor result which suggests that CO_2 by itself is not always effective.

The following experiments were designed to give both effects, i.e. the effect of CO_2 and the metallic oxide, working together. NaOH was used as the basic oxide. It was dissolved in alcohol and added to the sand, and the mixture mulled until dry. MnO_2, of industrial grade, was added as an oxidizing agent. MnO_2 would have at least the same oxidizing power as CO_2 but would give half the amount of gas. By using this mixture it was expected to obtain great improvement. However, the resulting castings showed very poor surface quality. Similar experiments using ½ per cent $Ca(OH)_2$ plus 2 per cent MnO_2 also produced highly defective castings.

This seems to indicate that the complete mechanism by which some carbonates work is of more complex nature than was previously thought. Possibly processes like catalysis and nucleation of the gas phase that would slow down the evolution of gases in the form of a chain reaction may be involved. Therefore, more elaborate methods of study should be used if a better understanding of the process is to be obtained.

The next step in the study of chemical additives was to try, on larger castings, the chemicals which proved to work successfully on smaller castings. The big slab pattern used for gating experiments was selected because of its size and section. This pattern permits easy control of gating and of rate of filling the mold. It was thought that if castings of 1-in. sections could be produced satisfactorily, the problem of shell molding steel castings would be practically solved.

The following chemicals Na_2CO_3 (c.p.), $CaCO_3$ (c.p.), MnO_2 (c.p.) and MnO_2 (tech. Company A) were tried together with different gating systems. Step, top, and bottom gating were each tried with 1-in. and ½-in. chokes. The sand was 80 AFS silica bonded with 5 per cent resin.

All these castings showed a bad surface condition suggesting that no matter what the gating and rate of filling the mold, castings of 1-in. section are not likely to have their surfaces dependably improved by the action of chemical additives.

A further experiment was performed using ½-in. thick slabs. The back of the casting, Fig. 6, was poured against a mixture of sand and resin alone while the front, Fig. 7, shows the improving effect of the $CaCO_3$ added. Therefore, it seems that chemical additives will improve the surface of castings up to certain thicknesses, as in the case of $CaCO_3$ up to ½-in. However, the effect on the surface will also depend on the shape of the casting and, more generally speaking, on the rate of skin formation. Therefore, the determination of the maximum section on which a chemical will work is difficult to establish for all conditions likely to be met in practice.

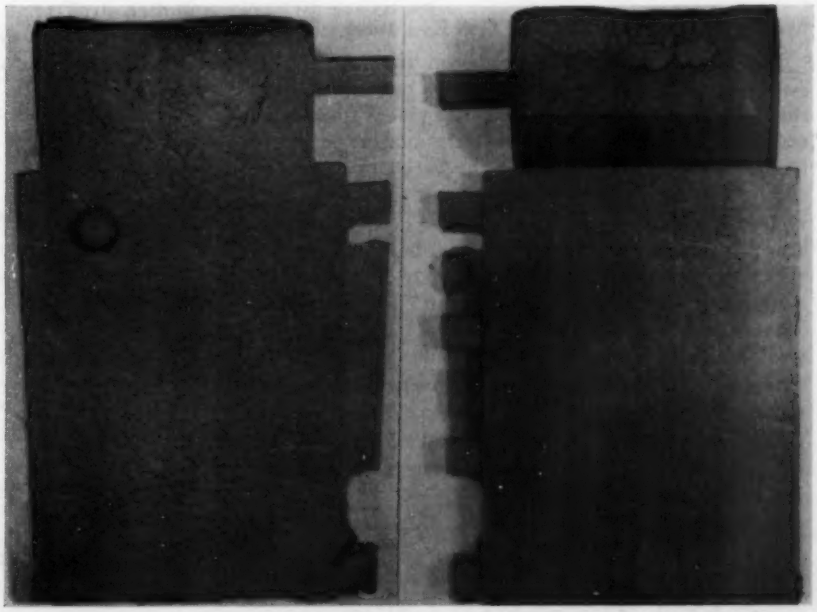


Fig. 6 (Left) — Back of ½-in. slab poured against a mixture of sand and resin.

Fig. 7 (Right) — Front of same slab but poured against a mixture of sand, resin, and 2 per cent $CaCO_3$, c.p.

Use of Different Kinds of Sands

Use of different kinds of sands without chemical additives was studied. Sands ranging from low to very high thermal conductivities were tried. These included kanamite, which consists of unicellular spherulized clay particles and produced a bad casting; forsterite (Mg_2SiO_4), an excellent casting; zircon ($ZrSiO_4$), a very good casting; and silicon carbide (SiC), a very bad casting.

Thus it may be seen that sands with low or very high thermal conductivities gave bad results while sands that, according to Locke, Briggs and Ashbrook,[6] have intermediate thermal conductivities produced good castings, especially forsterite which gave a casting with a surface completely free of defects. This makes the understanding of the problem even more difficult.

The next step was to determine whether chemical additives to zircon and forsterite would further improve the surface of the castings. MnO_2 and $CaCO_3$ were tried on zircon sand and $CaCO_3$ on forsterite. It was found that additions of MnO_2 impaired the surface quality of the castings, whereas, $CaCO_3$, improved it. $CaCO_3$ did not show any further improving effect on forsterite.

Zircon or forsterite sands are higher in cost than silica sand so it was decided to build up a composite shell using a facing layer of the effective material and backing it up with regular silica sand. This procedure was also suggested by Powell[1] although not tested. Experiments on composite shells with facing layers of $\frac{1}{32}$ in. and $\frac{1}{16}$ in. of forsterite and zircon sand were performed. A very good casting was obtained with $\frac{1}{16}$-in. layer of forsterite, whereas a marked improvement was observed when similar experiments were performed with zircon sand.

From the above experiments it follows that more radical methods than using chemical additives are required if castings with sections thicker than $\frac{1}{2}$ in. are to be produced successfully.

Measurement of the Rate of Gas Evolution

Some experiments were made on the rate of gas evolution and are quoted here only for information since the results obtained would not give a reasonable statistical analysis. They were performed to get some basic information concerning the mechanism of the reaction at the mold-metal interface.

A weighed sample of 500 mg of the desired mixture was put in a bottom sealed vycor tube, heated to its thermosetting point, and connected to a manometer containing a solution of potassium permanganate. The tube was then plunged into a bath of molten steel at 2950 F and a movie camera recorded the reading of the manometer every sixteenth of a second for a period of 10 seconds. The experiment was repeated with different mixtures and the results are plotted in Fig. 8 showing the change of height of liquid column with time.

Thus it is shown that the decomposition of the resin occurs at a very high speed; most of the gas being produced during the first two or three seconds. It is interesting to note that when MnO_2 was added a more uniform decomposition resulted.

Discussion of Results on Use of Chemical Additives and Sands

As pointed out in the introduction, the appearance of surface defects is probably due to the simultaneous effects of gas evolution from the organic binder and the solidifying characteristics of low carbon steel. However, it is surprising that even with shell molds of very high permeability at room temperature, the defect still persists. A possible explanation to this effect may be the following: the gaseous products of decomposition of the first layers of sand and resin in contact with the molten steel might condense in the back layers, decreasing, very much, the permeability of the shell and preventing any further gaseous products escaping from the shell.

Under these conditions any further evolution of gas will be reverted to the surface of the casting. This possibility has been somewhat confirmed during the experiments carried out to determine the rate of gas evolution of different mixtures of sand and resin described above. Liquids of high viscosity were found in the upper part of the vycor tube which contained the mixtures and which was plunged in molten steel. However, there is need for further experimentation in order to prove this theory.

The reactions of the chemicals are not well understood although several effects may be pointed out as follows:

1. Oxidizing effect. The additive may supply oxygen which increases the rate of combustion of the resin (MnO_2 and CO_2 at high temperatures) or it may break down the liquid residues condensed in the back layer of the shell thereby increasing the permeability.

2. Nucleation of gas phase. It is possible that when molten steel comes in contact with the resin an evolution of gas very similar to an explosion might occur.

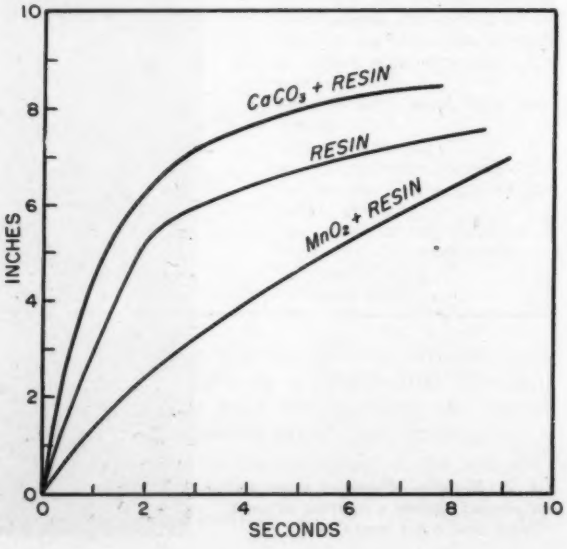


Fig. 8 (Right) — Rate of evolution of gases for different mixtures of sand, resin, and chemical additives.

Nucleation of the gas phase by the products of decomposition of MnO_2 of some carbonates may slow down the reaction allowing the gas to form more gradually.

3. Catalytic effect. The effect of activated alumina can only be explained as a catalytic effect because it is very unlikely that at the high temperature involved the alumina has any effect on adsorption of the gases produced.

The effect of sands is also confusing since, as shown above, the thermal conductivities of the sands do not account for the changes on the surface quality of the castings.

Production of Shells With Inorganic Binder

Development of a shell mold which would not evolve gases under casting conditions would be highly desirable but the selection of an inorganic binder which would set in a period of 2 to 4 minutes and give appreciable strength to the mold is not easy. Since inorganic chemical reactions occur in ionic state, the use of water solutions and, therefore, of wet sands is necessary. A blowing machine would have to be used instead of a dumping box because of the poor flowability of wet sands.

There are several mechanisms by which a molding material and inorganic binder could be made to set:

1. Gas through a mixture of wet sand with the binder, for instance, CO_2 through a mixture of sand and water glass.
2. Gas through a mixture of dry sand and binder, for example, water vapor through a mixture of sand and sodium silicate powder.
3. Reaction between two chemicals giving a cement type bond. Main trouble would be the mix may set in the hopper and other equipment.
4. Exothermic reaction of two compounds giving up heat that can be used to dry an inorganic binder. This method suffers from the same trouble as No. 3.
5. Elimination of water from the inorganic binder by blowing the mixture on a hot pattern.

This last method was the simplest and was selected for investigation. Sodium silicate was used as the bonding agent. The mixture of sodium silicate, water and sand was blown on a hot pattern and the loss of water caused the bonding of the grains of sand by solid sodium silicate.

Use of sodium silicate has some disadvantages such as lack of collapsibility, and lowering of the sintering point due to the presence of basic oxide Na_2O. Nevertheless these difficulties could be overcome. The effects of contraction, expansion and collapsibility could be minimized by using carbonaceous substances or different inorganic materials like perlite, etc. or appropriate mixtures of sand. Regarding the refractoriness of the sand, there is always the possibility of using other types of refractories such as magnesite or, perhaps, alumina.

Preliminary experiments with a small sand blowing machine gave very promising results. Before commencing the manufacture of actual shells, it was decided to carry out some experiments on the bonding properties of sodium silicate. The following factors were studied.

Effect of Temperature

It was not expected that temperature would have an effect on the mechanical strength of the sand bonded with sodium silicate provided the temperature was high enough to eliminate all the water from the sand. This was borne out in experiments carried out with 3 per cent $NaSiO_3$, 4 per cent water and 80 AFS silica sand. The temperature of the plate on which the sand was blown was measured with an iron-constantan thermocouple; temperatures ranging from 400 F to 800 F were studied. In all cases the ultimate tensile strength was about 150 psi. The thickness of the shell was a function of the temperature for a given length of time.

Effect of Amount of Sodium Silicate, Water and Size of the Sand

These experiments were to determine the optimum conditions for blowing a mixture on a hot pattern, that is, the minimum amount of water and sodium silicate which will give a reasonable strength. The procedure was as follows: the solution of water and sodium silicate was added slowly to the sand and mulled. After mulling, specimens were made in the shape of a "dog bone" which had a minimum cross-sectional area of 1 sq in. The specimens were then baked in a core oven for 20 minutes at a temperature of 700 F. After baking, the specimens were allowed to cool to room temperature and tested in a universal testing machine. The results as plotted in Figs. 9 and 10 are the average of ten measurements.

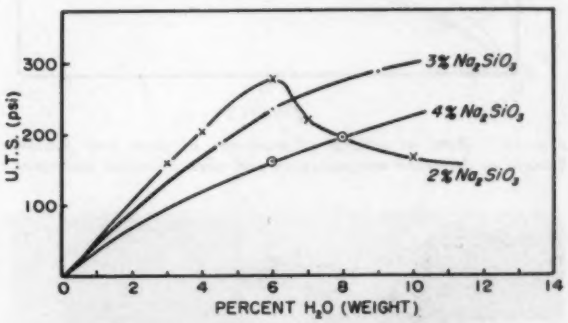


Fig. 9 — Effect of amount of water and sodium silicate on tensile strength for 60 AFS silica sand mixtures.

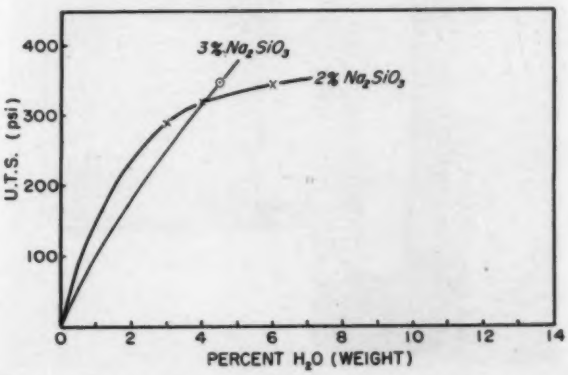


Fig. 10 — Effect of amount of water and sodium silicate on tensile strength for 140 AFS silica sand mixtures.

From the results obtained it can be seen that for a certain content of water higher values of tensile strength are obtained with the lower contents of sodium silicate. This might be due to the fact that for a certain content of water, the higher the concentration of sodium silicate the more viscous becomes the solution and, therefore, more difficult to spread over the entire surface of the grains, thus reducing the total area of contact. The results with 140 AFS silica sand showed ultimate tensile strength values much higher than those with 60 AFS sand, probably due to an increase of the total surface of contact of the grains.

Rate of Setting

A mixture of sand and water glass was blown against a hot plate, and the thickness of the layer versus time was determined. The results shown in Fig. 11 are the average of two measurements indicating that a variation in moisture from 4.5 to 6 per cent does not affect the rate of setting. The values obtained are not absolute but relative values because the rate of setting is dependent upon the heat capacity of the plate or pattern.

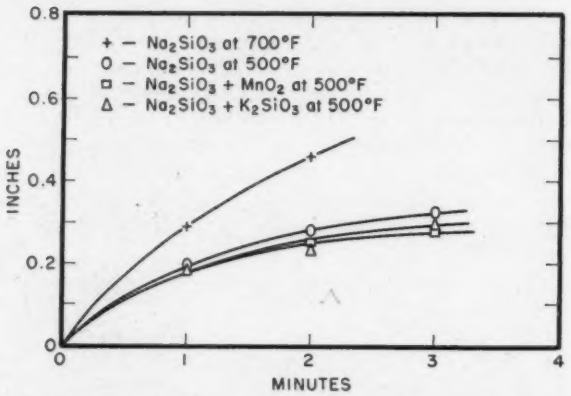


Fig. 11 — Rate of setting of mixtures of sand and sodium silicate at different temperatures and with different additives.

According to Vail[7] several substances increase the speed of setting of the solution of sodium silicate in water, of which, sodium fluosilicate, manganese dioxide and potassium silicate were tried. These were mixed with sand and water glass and the mixture blown on a plate held at 500 F. These results also are plotted in Fig. 11.

Because there was not an appreciable change in the rate of setting, it was decided to use sodium silicate alone without any additive. As pattern for making an actual shell the usual 1-in. slab pattern was selected. A contour plate was made and vents located conveniently.

A mixture consisting of 60 AFS sand, 3 per cent silicate and 6 per cent water was blown on the pattern heated at about 500 F using a sand blowing machine. After the shell set for 3 minutes it was removed from the pattern. The shell was strong, although it showed some defective areas, probably due to improper location of the vents.

Figures 12 and 13 show the front and back of a casting poured in a shell mold, of which the front shell was made of sand plus sodium silicate and the back shell made of sand plus resin. The part of the casting which solidified in contact with the shell bonded with sodium silicate showed no surface blows while the back of the casting showed the usual surface defect. By this method it seems possible to make shell molds which would not evolve gas when molten metal was poured into them and thereby produce castings free from surface blows.

Summary of Results

1. Changing the rigging, gating, and risering did, in some cases, eliminate surface defects. On 1-in. thick slabs some improvement was achieved by top

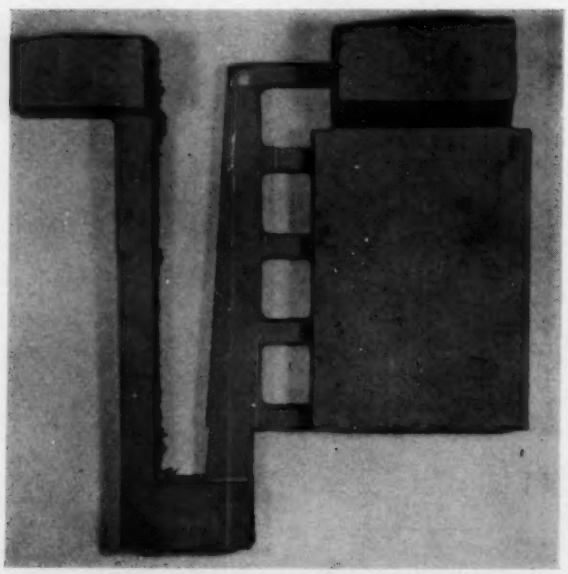


Fig. 12 — Casting poured in a shell mold made with sodium silicate.

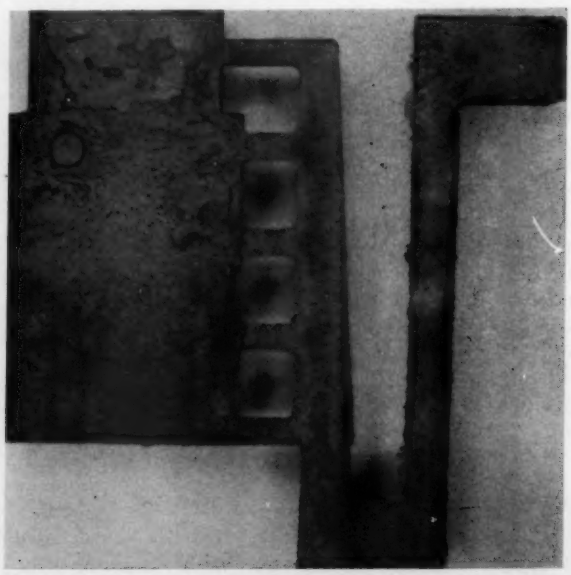


Fig. 13 — Back of same casting but poured against a mixture of sand and resin.

pouring at a slow rate.

2. Changing the ferrostatic pressure did not modify the surface quality of the castings studied.

3. The chemical additives tried did not improve the surface of castings with plate-type sections thicker than ½ in., but were beneficial for lighter sections.

4. Some industrial and technical grades of MnO_2 do not produce any improvements on the surface, even in small castings, while the addition of PbO_2 to these inactive grades of MnO_2 rendered them active and able to improve castings.

5. The addition of carbonates, especially sodium, magnesium or calcium carbonate, to the mixture of sand and resin improved the surface finish markedly. Chemically precipitated calcium carbonate (industrial grade) proved successful on plate-type castings less than ½ in. thick.

6. For plate-type sections over ½ in. thick the only successful method of producing good castings is to use zircon sand or forsterite, preferably the latter.

7. Calcium carbonate together with zircon sand further improved the surface of the casting while addition of manganese dioxide impaired it.

8. Castings were produced successfully with a composite shell having a 1/16 in. facing of forsterite and the remainder of the shell made of regular silica sand.

9. Shell molds were produced using sodium silicate as an inorganic binder by blowing the mixture of sand and sodium silicate on a hot pattern. Steel castings made in this type shell mold were free of surface defects.

Acknowledgments

The authors wish to express their thanks to Watertown Arsenal and the Ordnance Department of the United States Army for their sponsorship of this research program, to the members of the Foundry Section at the Massachusetts Institute of Technology for their continuing assistance, and to George E. Schmidt, Jr. for his aid in preparing the illustrations.

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METAL SOLIDIFICATION IN A FLOWING STREAM

By

M. C. Flemings, C. M. Adams, E. E. Hucke, and H. F. Taylor*

ABSTRACT

The grain structure of a casting is commonly described in terms of alloy composition, cooling rate, and nucleation. It is shown here that a fourth variable must sometimes be considered; the flow of liquid metal across the solidifying interface. Several unusual cast structures caused by the flow of liquid metal are presented and explained in terms of a "boundary layer theory". Methods are presented for the elimination of the effects described.

In most castings, pouring time is relatively short in comparison with freezing time, and pouring is complete before solidification begins. In these castings solidification generally progresses into a quiet pool of liquid metal. However, in some cases such as large chilled non-ferrous castings, solidification may begin before pouring is complete, and here the solidifying front is progressing into a moving stream of metal. Solidification may also progress into a flowing stream in centrifugal castings and in certain ingot practices, or in large sand castings due to convection. The effect of this liquid metal movement on the solidification structure is therefore of practical importance.

Solidification in a Quiet Liquid

Consider first a metal freezing into a quiet pool of liquid; a normal sand or chilled casting. This may be described in a manner similar to that of Winegard and Chalmers[1] in terms of a "boundary layer".

Absolutely pure metals must freeze at one temperature and with a plane front liquid-solid interface. The addition of even small amounts of impurities or alloying elements, however, complicates the solidification pattern, and the interface must be described in terms of the appropriate phase diagram. In a freezing alloy such as that of Fig. 1, the first metal to form is of lower solute concentration than the bulk liquid. The solute atoms must therefore pile up and diffuse through a gradient away from the freezing interface into the bulk liquid. This local high concentration of solute atoms is referred to as a "boundary layer" and affects the solidification structure in a manner which may be seen by considering the effect in greater detail.

Assume the alloy C_o of Fig. 1 to be initially freezing with a plane front interface as sketched in Fig. 2. At the start of freezing, the liquid is of composition C_o and the first solid is S_s as predicted by the phase diagram. These initial concentrations are drawn in Fig. 3. Rejection of solute atoms by the freezing solid, however, rapidly builds up the interface composition in the liquid to some level C_i (Fig. 4) governed by the rate of rejection of the solute from the solid and the rate of diffusion of the solute into the liquid. This buildup represents the boundary layer. Thus, if

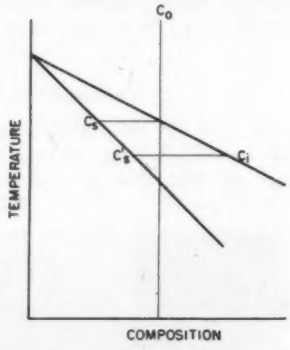


Fig. 1 — Phase diagram.

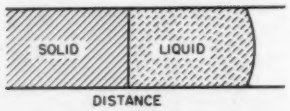


Fig. 2 — Plane front solidification.

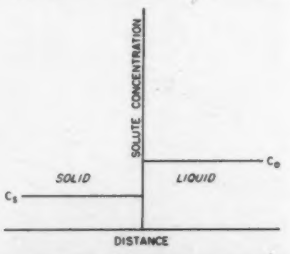


Fig. 3 — Concentration gradients at start of freezing.

*Flemings, Adams, and Taylor are on the faculty at Massachusetts Institute of Technology, Cambridge, Mass., while Hucke is at University of Michigan, Ann Arbor.

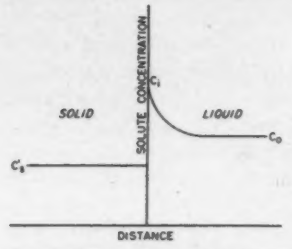


Fig. 4 — Concentration gradients as freezing progresses.

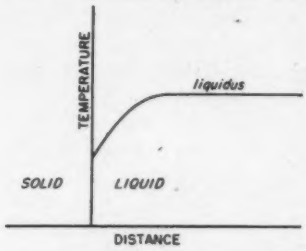


Fig. 5 — Freezing temperature gradient in boundary layer.

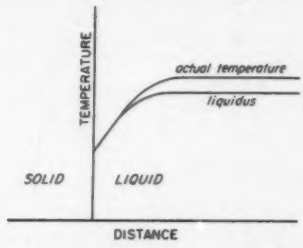


Fig. 6 — Temperature gradients required to maintain plane front freezing.

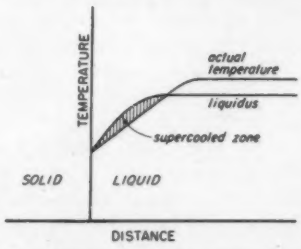


Fig. 7 — Temperature gradients to maintain columnar freezing.

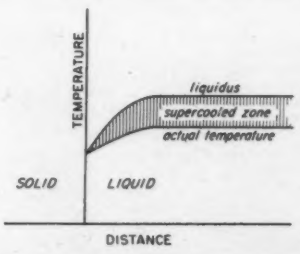


Fig. 8 — Temperature gradients promoting equiaxed grain formation.

plane front freezing is maintained, large grains and violent macro-segregation must ensue. This is not the usual case in an alloy casting.

The reason lies in a breakdown of the interface and heterogeneous nucleation. Because the boundary layer has a higher alloy concentration than the bulk liquid it has a lower freezing point than the bulk liquid, and acts as a barrier to further solidification. Dendrite arms then tend to reach into the somewhat super-cooled liquid, or heterogeneous nucleation takes place there.

This is seen with reference to Fig. 5-8. For each composition of the boundary layer shown in Fig. 4, the phase diagram indicates a discreet freezing point (liquidus) as drawn in Fig. 5. To maintain this plane front interface now requires very steep temperature gradients in the liquid as shown in Fig. 6. A temperature gradient in the liquid which is less than that of Fig. 6 will result in a zone away from the interface being slightly super-cooled. This is shown in the shaded area of Fig. 7, and results in a casting with columnar dendritic structure.[1,2] That is, the barrier is unstable and dendrite arms reach out into the supercooled liquid. Finally, if the casting temperature gradients become even less with respect to the liquidus gradients, a greater zone of super-cooled metal may occur, complete breakdown of the growth barrier results, and nucleation may take place away from the interface. Thermal gradients are shown in Fig. 8. These latter two thermal conditions are the cases commonly met in commercial castings. A plane front interface is seldom obtained except in zone melting highly purified materials.

Solidification in a Flowing Stream

The action of a flowing stream on a growing solid may now be considered. The moving liquid sweeps away the impurities from the solidifying interface, and results in a reduction of "boundary layer" thickness,[3] or reduction in the "solidification barrier". Since the "barrier of solidification" is reduced there is less tendency for its breakdown, and less tendency for nucleation to occur away from the interface.

Consider a casting which, without liquid movement, freezes under thermal conditions such as those sketched in Fig. 8. The resulting grain structure is equiaxed. If, however, the liquid metal is flowing past the interface, the concentration gradient and maximum concentration C_i (Fig. 4) will be reduced. The liquidus temperature gradient of Fig. 5 will be correspondingly reduced and the freezing temperature slightly raised. The net effect of these changes is a reduction in the super-cooled zone sketched in Fig. 8. If the super-cooled zone is sufficiently reduced, the thermal conditions of Fig. 7 are obtained, and additional reduction results in the thermal conditions of Fig. 6. Thus, if a thermal gradient exists in a casting, sufficient stirring may alter a fine grain casting to a columnar casting and additional stirring may permit the metal to freeze in a "plane front" without dendrites.

Figure 9 shows this alteration. Both castings were rigged and risered from the right, and bottom chilled. Both were poured of commercially pure aluminum (99.85 per cent) at the freeze point (1215 F). Casting 9B was of conventional rigging with no metal flow-off. It presents a typical equiaxed fine grain structure, the impurities being sufficient to make the casting freeze as an alloy with thermal characteristics of Fig. 8. The left end of the bar of Fig. 9A was cut open and the metal allowed to flow out of the mold until choked off by freezing. Thus the solidifying front progressed into a rapidly moving stream of metal. Thirty pounds of metal flowed out before freeze-off

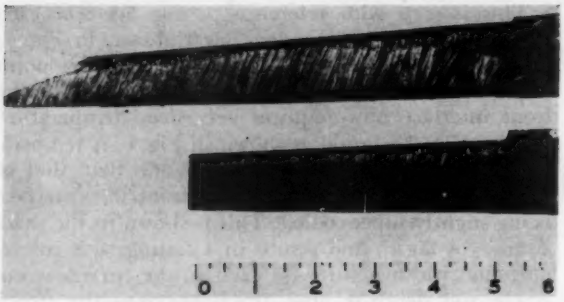


Fig. 9 — Bottom chilled commercial purity (99.85%) aluminum bars. Macro-etch of vertical section; (A) Metal spilled out left end during solidification, (B) Left end plugged; static cast.

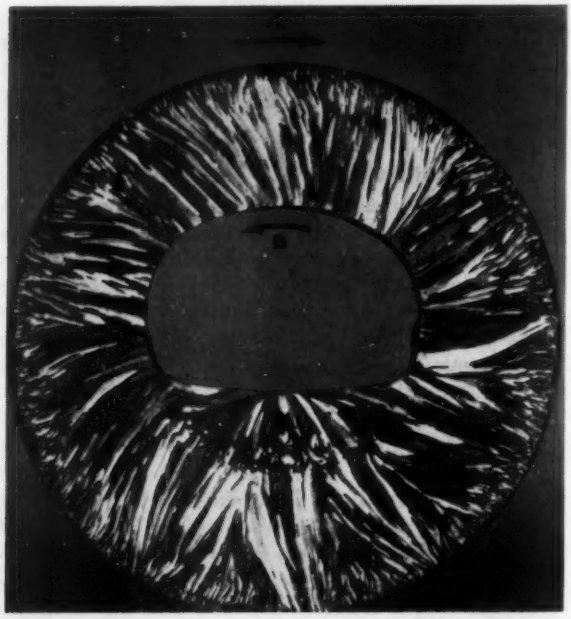


Fig. 10 — Commercial purity copper centrifugal casting. "A" is direction of mold rotation, "B" is direction of slip velocity of liquid metal with respect to mold.

occurred, after 22 seconds. This stream was sufficient, as shown, to alter the fine grain structure of Fig. 9B to the columnar grains of Fig. 9A. Metallographic examination confirmed the grains to be columnar dendritic and not plane front non-dendritic.

It is of interest to note that the columnar grains point "upstream". If the growing dendrites are pictured as protruding slightly into the boundary layer, it is apparent the upstream side of the grain will be washed by the flowing metal to a greater extent than the downstream side. Thus the "growth barrier" is more efficiently removed on the upstream side, and most rapid growth occurs in this direction.

A similar effect is seen in the centrifugal casting of Fig. 10, a commercially pure (99.99 per cent) copper casting. Mold spin direction was A, relative liquid movement with respect to the solid was B. The grains point upstream, and point more sharply upstream in the early period solidification when the relative slip velocity B is greatest.

It is of further interest to examine the case of a casting solidifying into a flowing stream when the "growth barrier" is not maintained constant. A variation in the growth barrier may be obtained in several ways. The most common of these are variations in solidification rate or liquid flow rate.

A sudden growth barrier variation was induced in the casting of Fig. 11 by the latter means. Rigging and chilling of this casting was identical to that of the preceding bars; the metal was allowed to flow out the left end. An aluminum alloy containing 6.5 per cent tin was poured at 1225 F. Freeze-off time was 43 seconds and 40 pounds of metal spilled over in this time. After approximately 20 seconds, the liquid flow was interrupted by plugging the end for one second, and then allowed to continue spilling.

The casting shows a tendency for columnar grains, pointing in the upstream direction. These are not as pronounced as in the preceding bar due to the higher alloy content. A band is apparent in the etched structure approximately three-quarters of the total distance from the chill. This band contains a high percentage of low-melting constituents, or solute atoms, and its presence is explained by the more complete breakdown of the boundary layer when flowing was interrupted. The "band" consists of the suddenly entrapped solute atoms concentration at the interface.

A more pronounced band is seen in the centrifugal casting of Fig. 12. This casting was poured of the aluminum-tin alloy. Note the columnar grains, followed by a band, and subsequently by fine grains. In this case the boundary layer breakdown may have



Fig. 12 — Aluminum-tin alloy chilled centrifugal casting. Macro-etch of radial section.



Fig. 11 — Bottom chilled Al-Sn alloy bar. Macro-etch of vertical section. Metal spilled out left end during solidification.

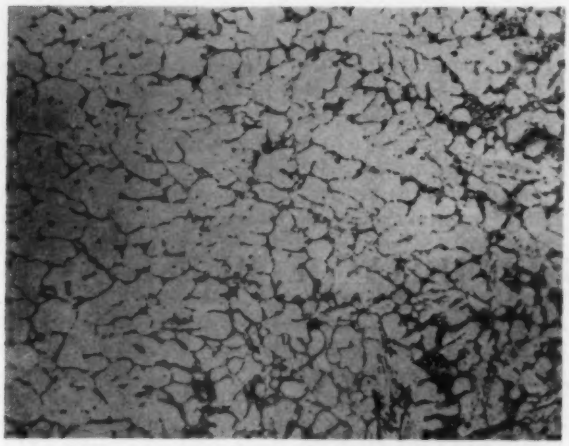


Fig. 13 — Photomicrograph of "band" of aluminum-tin centrifugal casting. Mold circumference is to the right. Mag.—100X.

been initiated by mold vibrations, liquid flow variations, or simple accumulation of constituents at the interface. The tendency for band formation repeats itself as seen in Fig. 12. A photomicrograph of the outer band is shown in Fig. 13.

Discussion

The effect of liquid metal movement on casting structure, when present at all, may not be harmful. The bands in the aluminum-tin alloy centrifugal casting shown, however, present planes of weakness and can not be tolerated. Bands in castings of this type were eliminated both by reducing chill rate and by lowering alloy content, as would be expected from the boundary layer theory.

The grain coarsening effect of liquid flow shown in Fig. 9 may also be undesirable. The variables which should reduce the tendency for this coarsening (formation of columnar grains) have been described in general terms here, and more completely elsewhere.[4] They are:

1. Reduction of liquid movement
2. Increase of alloy content
3. Reduction of thermal gradients
4. Use of alloying elements which possess
   a. Wide liquid-solid range
   b. Liquidus of steep slope
   c. Low diffusion rate in the liquid metal.
5. Addition of nucleating agents.

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SOME FACTORS AFFECTING FLUIDITY OF METALS

By

D. V. Ragone,* C. M. Adams,** and H. F. Taylor***

ABSTRACT

The fluidity of metals in the foundry sense (the distance that liquid metal can flow in a channel before being stopped by solidification) was studied both analytically and experimentally. Metal properties such as heat of fusion, heat capacity, viscosity, melting temperature and density were considered. The influence of test variables such as applied pressure head, channel diameter, and mold material properties was considered.

Analytically, equations were derived relating fluidity of pure metals to metal properties and test variable. Fluidity was found to vary directly with channel diameter, volumetric heat of fusion, applied pressure head, and superheat. Fluidity varied inversely with friction factor, temperature difference between metal and mold, and heat-absorbing ability of the mold. Change in viscosity was found to have but a small effect.

Some of these relationships were checked experimentally using a new fluidity test designed to conform with the conditions of the derivations.

In order to discuss the subject of fluidity, there must first be a clarification of the ambiguous terms used in the field. The term "fluidity" has been used as a fundamental physical constant of materials (reciprocal of the viscosity). Foundrymen, however, have adopted the term and have changed the meaning somewhat. In the foundry sense, fluidity is an empirical measure of the distance a liquid metal will flow before it solidifies. The values of fluidity thus obtained are not fundamental physical constants of the metal, but are rather a function of many physical constants and of the variables of the test used. Fluidity values are generally given as distance flowed through a channel of particular cross section before solidification.

*Submitted to the Department of Metallurgy, Massachusetts Institute of Technology, in partial fulfillment of the requirements for the degree of Doctor of Science; now, Assistant Professor of Metallurgical Engineering, University of Michigan, Ann Arbor, Mich.

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History of Fluidity Testing

T. D. West [37] in 1898 was the earliest investigator to report on the properties of molten metals as cast into sand molds. He poured metal into a horizontal wedge and considered the distance flowed as a measure of fluidity. Lebedure in 1904 [38], Sexton and Primrose [38] in 1911 and Moldenke in 1917 [39] modified the wedge test somewhat.

Ruff [26] ran metal in a long cylindrical channel and used the length flowed in the channel as his value of fluidity. This test was particularly sensitive to errors in leveling. Evans [6] has tried an inverted "U" type of test in which he uses several vertical sections of various cross-sectional areas fed from a common channel. The heights to which the metal rises in the various sections is a gage of fluidity.

The familiar fluidity spiral was first tried by Saito and Hayaschi [40] in 1919. Their channel was wound into a spiral thereby simplifying handling and leveling problems. Many investigators have used and improved this spiral type of test, and it is now accepted in America as a standard for fluidity of iron and steel. [2, 6, 9, 11, 20, 27, 28, 32, 34, 41] Eastwood and Kempf [5] have used a spiral casting of flat cross section for studying aluminum and its alloys. Because the length of such a spiral is not well defined, they use the volume of metal in the channel as a measure of fluidity.

The Variables

The variables that affect the fluidity data may be divided into three broad classifications; the inherent properties of the metal, the extrinsic properties of the metal, and the variables of the test equipment.

Properties of the Liquid

The intrinsic properties of the liquid which have significant effects on its fluidity are heat content (enthalpy), composition, surface tension, and viscosity. The literature is consistent with respect to the effect of heat content [2, 10, 18, 19, 20-25, 32]. Fluidity varies directly with heat content measured in terms of superheat (temperature increment above the liquidus). The effect of composition on fluidity has been investigated by pouring metal at a constant temper-

ature interval above the liquidus across binary and ternary systems [20-25, 10]. The results of these tests indicate that the fluidity is inversely proportional to the solidification interval (liquidus temperature—solidus temperature). The explanation advanced for this behavior hinges on the manner of solidification of the alloy. In the case of a pure metal or eutectic, the metal solidifies with smooth regular interfaces inward from the mold wall. Alloys with a solidification range solidify by forming dendrites which impair flow considerably.

It seems logical that variations in surface energy should affect the fluidity of a metal. In pouring the test, a new surface is being created which requires a source of energy. This source must be the moving stream of metal. This energy sink, however, proves to be a small one.

The question of whether or not the fluidity tests are viscosity-sensitive has been the subject of much discussion. Desch [4] and Briggs [34] hold that viscosity has no effect, while some other authors hold the opposite view [9, 35, 36]. The opinions of the above authors are based on experimental data. This investigation treats the subject from an analytical point of view.

Extrinsic Properties of the Liquid

Dissolved gas, surface oxide films, and insoluble impurities can affect fluidity. Dissolved gases can either change physical properties of the metal such as viscosity, or provide shear mechanical interference as the gas comes out of solution upon solidification. The properties of oxide films that must be considered are the strength and specific gravity of the film and whether the film is solid or liquid at the casting temperature. A solid oxide film (such as that on aluminum) might form a continuous envelope around the stream and restrict its flow, or the film might be torn apart and mixed with the moving stream which would increase the effective viscosity. It has been found that solid films increase, and liquid films decrease the fluidity [9].

Variables of the Test Equipment

The casting system for a fluidity spiral consists of (Fig. 1) (1) a pouring basin and a downgate, (2) a "cushion" to dissipate kinetic energy of the falling stream, and (3) a spiral channel. Among the important test variables are the channel shape, design of the cushion, molding materials, and pouring rate. Greene [8] has shown that fluidity varies inversely as the perimeter to cross-sectional area ratio of the channel. Andrew et al [1], investigating the effect of cushion design, have demonstrated that under similar casting conditions different cushion designs gave fluidity values that varied by as much as a factor or two. The thermal diffusivity of the mold material, mold temperature, and pouring speed all influence fluidity [3, 9, 11, 20, 28, 32, 41]. Spiral length varies directly with pressure head applied [9], hence it is not unreasonable to assume that the length would vary with the rate of application of pressure head. The difficulties involved in controlling the above-mentioned variables make it almost impossible to compare the date of the various authors.

Quantitative Contributions

Ruff [26] evaluated friction factors in steel flowing in a straight, cylindrical channel and found them to correspond closely to those predicted by the Reynolds number correlation. Portevin and Bastein [20-25] derived an equation which fits their data involving empirical constants which are a function of the equipment used. Taylor and Rightmire [33] have derived the relationship between fluidity and temperature of the metal and some equations for evaluating friction forces. These show good correlation. However, their data are taken from many investigators and are difficult to evaluate.

The purpose of the present investigation is to relate quantitatively fluidity to fundamental properties of metals, molding materials and test equipment design. Specifically, the plan is to derive analytically equations relating the fluidity of pure metals to the above mentioned properties and to conduct controlled experiments to check these relationships.

Analysis Of Solidification In A Cylindrical Mold

This section deals with heat transfer from a cylindrical mold, namely, the fluidity spiral channel. The analysis deals with a pure metal at its melting point, with the result that any heat withdrawn from the mold results in the solidification of some metal. A relationship is derived for the radius of the liquid stream at a point in the channel as a function of time of contact between the mold and the metal. Two different modes of heat transfer are considered.

In both of the following cases, the following assumptions are made:

1. The liquid metal is at its melting point.
2. There is no supercooling.
3. Metal solidifies on the inside of tube with smooth interface.
4. Metal that has frozen out provides no barrier to heat transfer.

Case I, (h-type)

In this case the main resistance to heat flow is at the inner surface of the tube, that is, at the mold-metal interface. This analysis applies to situations where the channel diameter is small (ratio of surface to volume is high) and when the mold material may be considered a chill. q, the amount of heat transferred through this surface per unit time is:

$$q = hA(T_s - T_r) \qquad \text{(A-1)}$$

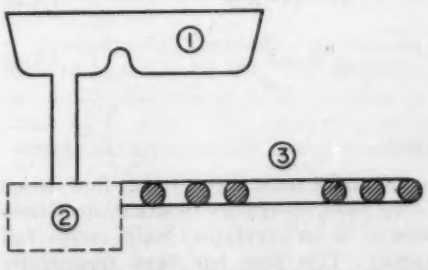


Figure 1—Casting system for fluidity spiral

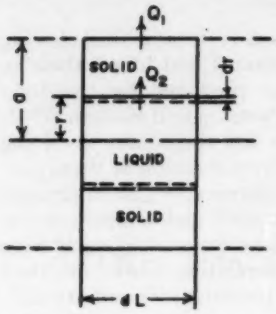


Fig. 2—Differential element of tube.

Consider the element dL in Fig. 2. In time $d\theta$, the element dr freezes out.

$$dQ_1 = 2\pi r \rho H_f \, dr \, dL \qquad (A\text{-}2)$$

and

$$dQ_2 = -h(T_s - T_r)\, 2\pi a \, d\theta \, dL \qquad (A\text{-}3)$$

where:

dQ_1, dQ_2 = Differential quantities of heat as indicated on Fig. 2
r,a,L = Dimensions as indicated on Fig. 2
h = Heat transfer coefficient as defined by Eq. (A-1)
ρ = Density of metal
H_f = Heat of fusion of metal (cal/gram)
T_s = Equilibrium solidification temperature of metal
T_r = Room temperature, that is, the initial temperature of the mold or tube.

since

$$dQ_1 = dQ_2,$$

$$r\,dr = -ah\left(\frac{T_s - T_r}{\rho H_f}\right) d\theta \qquad (A\text{-}4)$$

Integrating

$$r^2 = a^2 - 2ah\left(\frac{T_s - T_r}{\rho H_f}\right)\theta \qquad (A\text{-}5)$$

The time for final solidification θ_f is (r = O):

$$\theta_f = \frac{a}{2h}\left(\frac{\rho H_f}{T_s - T_r}\right) \qquad (A\text{-}6)$$

For purposes of notation let

$$\beta = \frac{2h\,(T_s - T_r)}{\rho H_f} \qquad (A\text{-}7)$$

Then

$$r^2 = a^2 - a\beta\theta \qquad (A\text{-}8)$$

$$\theta_f = \frac{a}{\beta} \qquad (A\text{-}9)$$

Case II (θ-type)

In this case the main resistance to heat flow arises in the mold. This analysis applies to situations where the mold material is an insulator. Sand molds fall into this category. This case has been treated by Adams [43] who gives as a very good approximation of the heat transferred into the mold:

$$q = A(T_s - T_r)\left(\frac{K'\,\rho'\,C_p'}{\pi}\right)^{1/2} \theta^{-1/2} \qquad (A\text{-}10)$$

where:

K' = Thermal conductivity of the mold
ρ' = Density of the mold material
C_p' = Heat capacity of the mold material
θ = Time after contact of the metal and mold.

For purposes of notation let

$$\gamma = 4\left(\frac{K'\,\rho'\,C_p'}{\pi}\right)^{1/2} \frac{(T_s - T_r)}{\rho H_f} \qquad (A\text{-}11)$$

Again referring to Fig. 2

$$dQ_1 = 2\pi r\, \rho H_f \, dr \, dL \qquad (A\text{-}2)$$

$$dQ^2 = -2\pi a\,(T_s - T_r)\left(\frac{K'\,\rho'\,C_p'}{\pi}\right)^{1/2} \theta^{-1/2}\, d\theta \, dL \qquad (A\text{-}12)$$

Following the same pattern as in the previous case:

$$r^2 = a^2 - a\gamma\theta^{1/2} \qquad (A\text{-}13)$$

$$\theta_f = \frac{a^2}{\gamma^2} \qquad (A\text{-}14)$$

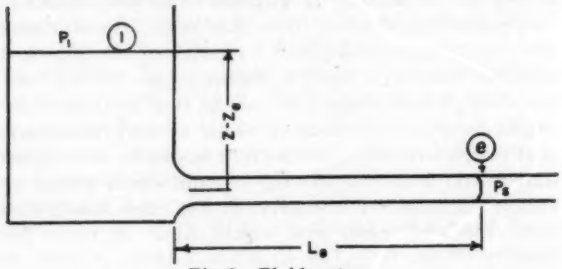


Fig. 3—Fluid system.

Analysis Of Flow In A Tubular Mold With Solidification

The system being considered in this section is illustrated in Fig. 3. The assumptions made are:

1. The fluid is continuous.
2. The pressure P_e is constant.
3. The effects of acceleration may be neglected.

Taylor and Rightmire [33] have performed an analysis of this last assumption and have shown it to be quite justifiable (accurate to within 2 per cent).

With the acceleration terms neglected, the equations of motion reduce to a total energy balance for the steady state (B-1)

$$Z_1 + \frac{V_1}{2g} + \frac{P_1}{\rho} = Z_2 + \frac{V_e^2}{2g} + \frac{P_e}{\rho} + \int_0^{L_e} \frac{fV^2}{2rg}\, dL \qquad (B\text{-}1)$$

where:

g = Acceleration of gravity
$Z_e - Z_1$ = Difference in height between E and 1
$P_e - P_1$ = Difference in pressure between E and 1
ρ = Density of metal
V_e = Velocity at the end of the stream
V = Local average velocity across any cross section of the tube
r = Local radius of tube
f = Friction factor.

For purposes of notation:

$$2g\left[(Z_1-Z_e)+\frac{P_1-P_e}{\rho}\right]=H \qquad \text{(B-2)}$$

Since the area of the reservoir is large compared to the cross section of the tube, $V_1=0$. Eq. (B-1) reduces to:

$$V_e^2+\int_o^{L_e}\frac{fV^2}{2r}\,dL=H \qquad \text{(B-3)}$$

With continuity, $AV=A_eV_e$

$$\pi r^2V=\pi a^2V_e;\quad V=V_e\frac{a^2}{r^2} \qquad \text{(B-4)}$$

where a = radius of container. At the end of the liquid column (point e) the diameter of the liquid stream is the diameter of the container. Substituting (B-4) into (B-3):

$$V_e^2\left(1+\int_o^{L_e}\frac{fa^4}{2r^5}\,dL\right)=H \qquad \text{(B-5)}$$

With the metal at its melting point flowing into a tube, after a time θ, the situation illustrated in Fig. 4 exists. As time θ increases, L_e increases, r_o decreases until finally at $\theta=\theta_f$ (Eq. A-6, A-14), $r_o=0$ and flow ceases. Thus the mechanism by which the moving stream is stopped is a "choking off" phenomenon occurring near the beginning of the stream.

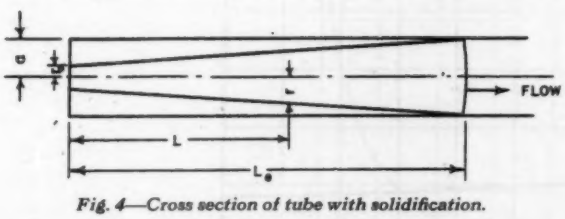


Fig. 4—Cross section of tube with solidification.

In order to solve Eq. (B-5); assume:

$$r=r_o+\frac{a-r_o}{L_e}L \qquad \text{(B-6)}$$

Assume too that the friction factor, f, is constant. Both these assumptions are justified in Appendix C.

The friction term of (B-5) can now be simply evaluated.

$$\int_o^{L_e}\frac{fa^4}{2r^5}\,dL=\frac{f}{8r_o^4}\left(\frac{a^4-r_o^4}{a-r_o}\right) \qquad \text{(B-7)}$$

Eq. (B-5) then becomes

$$V_e^2\left(1+\frac{f}{8r_o^4}\left[\frac{a^4-r_o^4}{a-r_o}\right]L_e\right)=H \qquad \text{(B-8)}$$

or

$$\left(\frac{dL_e}{d\theta}\right)^2\left(1+\frac{f}{8r_o^4}\left[\frac{a^4-r_o^4}{a-r_o}\right]L_e\right)=H \qquad \text{(B-9)}$$

since $H=V_o^2$;

$$\left(\frac{dL_e}{d\theta}\right)^2\left(1+\frac{f}{8r_o^4}\left[\frac{a^4-r_o^4}{a-r_o}\right]L_e\right)=V_o^2 \qquad \text{(B-10)}$$

Eqs. (A-5) and (A-13) give $r_o=f(\theta)$. Eq. (B-10) becomes an equation in L_e and θ which can be solved in various fashions.

Solution for L_e As A Function Of θ

Case I (Friction factor $f=0$)

The integration of Eq. (B-10) yields

$$L_e=V_o\theta$$

and

$$L_f=V_o\theta_f$$

where L_f is the final length. For h-type heat transfer

$$L_f=V_o\frac{a}{\beta}=V_o\frac{a}{2h}\frac{\rho H_f}{T_s-T_r} \qquad \text{(C-2)}$$

For θ-type

$$L_f=V_o\frac{a}{\gamma^2}=V_o\frac{a^2}{4}\frac{\pi}{K'\,p'\,C_p'}\left[\frac{\rho H_f}{T_s-T_r}\right]^2 \qquad \text{(C-3)}$$

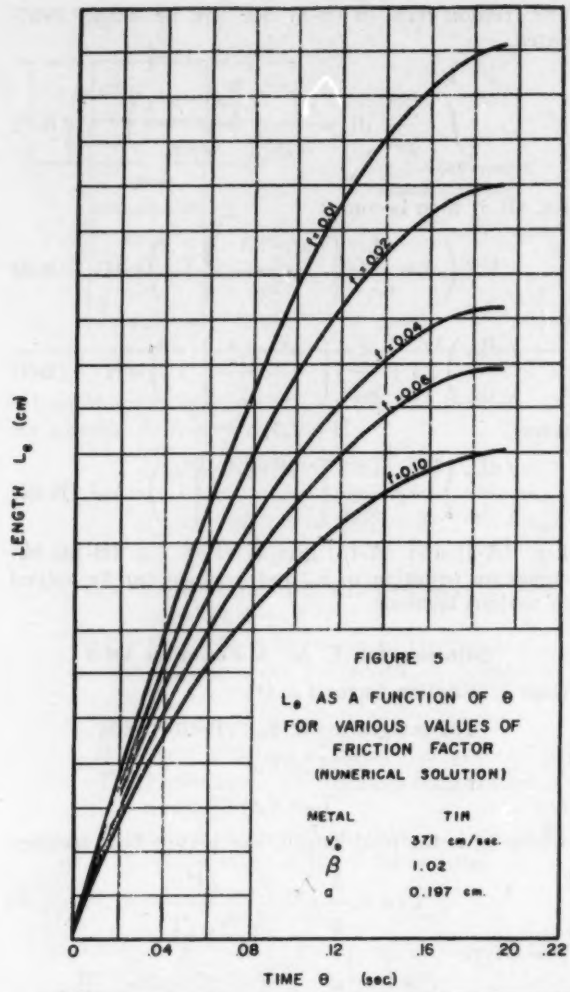
Case II (Constant friction factor)

The equation (B-10) can be solved by either a numerical method (modified Euler technique) ([45]) or by application of Simpson's rule ([46]). To use the modified Euler technique, increments were taken in the independent variable, θ. Through the use of equation (A-5), the factor $f(a^4-r_o^4)/8r_o^4(a-r_o)$ was evaluated as a function of θ. Successive approximations to L_e as a function of θ were made as outlined in Appendix A. The solution converges very rapidly and is quite stable after the fifth approximation. Fig. 5 lists the solutions to the equation for several values of friction factor.

Application of Simpson's rule provides another solution to equation (B-10). (See Appendix B). In order to apply this rule, it is first necessary to assume a form of $L_e=f(\theta)$. The numerical integrations have shown that a reasonable approximation is:

$$V_e=\left(1+\frac{\theta}{\theta_f}\right)V_o \qquad \text{(C-10)}$$

since



$$L_e = \int_0^{\theta} V_e d\theta, \qquad \text{(C-11)}$$

$$L_e = \frac{1}{2}\frac{V_o}{\theta_f}\left[\theta_f^2 - (\theta_f - \theta)^2\right] \qquad \text{(C-12)}$$

Eq. (B-10) becomes

$$\left(\frac{dL_e}{d\theta}\right)^2\left[1 + \frac{f}{8r_o}\left(\frac{a^4 - r_o^4}{a - r_o}\right)\left(\frac{1}{2}\frac{V_o}{\theta_f}\right)\left(\theta_f^2 - (\theta_f - \theta)^2\right)\right] = V_o^2 \qquad \text{(C-13)}$$

The solution to (C-13) is simplified if the independent variable is changed from θ to r_o. Considering the h-type of heat transfer, this substitution yields:

$$\frac{dL_e}{dr_o} = -\frac{2\,V_o\,r_o}{a\beta}\left[1 + \frac{f}{16}\left(\frac{a^4 - r_o^4}{a - r_o}\right)\left(\frac{V_o}{a^2\beta r_o^4}\right)\right]^{-1/2} \qquad \text{(C-14)}$$

$$L_f = -\frac{2V_o}{a\beta}\int_a^{\circ}\left[1 + \frac{f}{16}\left(\frac{a^4 - r_o^4}{a - r_o}\right)\left(\frac{V_o}{a^2\beta r_o^4}\right)\right] r_o dr \qquad \text{(C-15)}$$

Applying Simpson's Rule (4 parts)

$$L_f = \frac{V_o a}{\beta} F\left(\frac{V_o f}{\beta}\right) \qquad \text{(C-16)}$$

where

$$F\left(\frac{V_o f}{\beta}\right) = \frac{1}{6}\left[1 + \left(1 + 21.3\,\frac{V_o f}{\beta}\right)^{-1/2} + \left(1 + 1.76\,\frac{V_o f}{\beta}\right)^{-1/2} + \left(1 + .370\,\frac{V_o f}{\beta}\right)^{-1/2}\right]$$

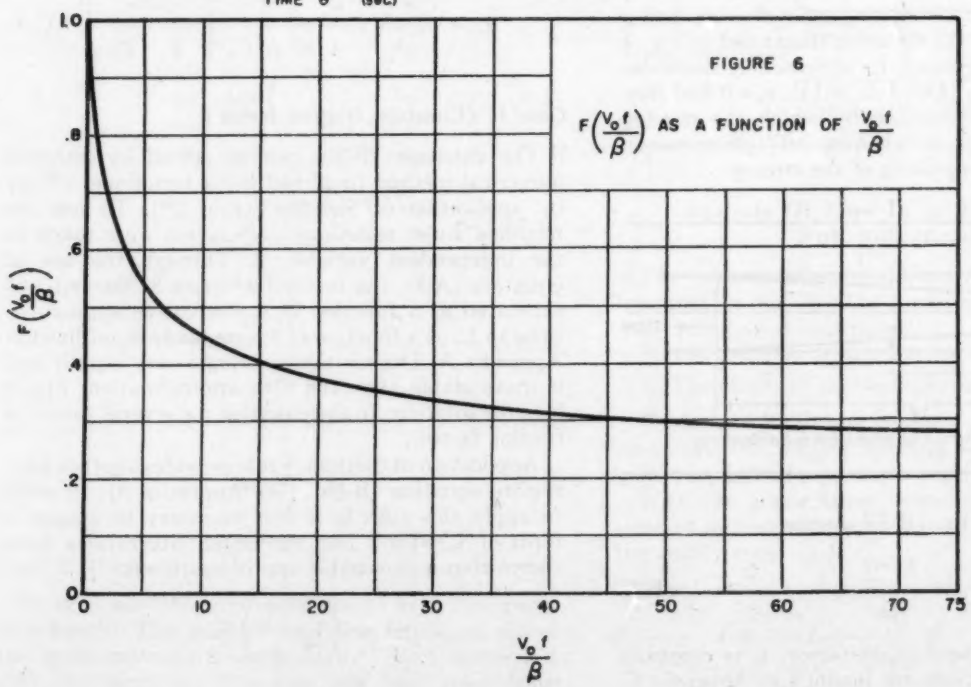


Fig. 6 is a plot of $F(V_o f/\beta)$ vs. $(V_o f/\beta)$, and Fig. 7 gives L_f as a function of V_o/β for various values of f. The table in Appendix B shows that these values calculated using Simpson's Rule are quite close to those determined by the modified Euler method. This provides justification for approximation (C-10).

A similar analysis for the case of θ-type heat transfer yields

$$L_f = \frac{V_o a^2}{\gamma^2} G\left(\frac{f V_o a}{\gamma^2}\right) = \frac{V_o a^2}{\gamma^2} G(D)$$

Fig. 8 is a plot of G(D) vs. D.

Effect Of Superheat On Fluidity Of A Pure Metal

Commercially, metals are never poured at their melting points. It becomes necessary thus to consider the effect of superheat on fluidity.

In order to treat analytically superheated metals, it is convenient to deal with the problem in two steps: first, the stage in which the metal in the mold cools off to the melting point (no solidification) and second, the process of solidification. Let the length attained by the metal in stage one be L_1 and the distance flowed after solidification has begun be L_2. To simplify the analysis, assume that the two stages are independent so that

$$L_f = L_1 + L_2 \qquad \text{(D-1)}$$

To solve for L_1, it is necessary to consider the equation of flow once again. (B-3) In stage one with no solidification, $V = V_o$ and $\gamma = a$. With f constant

$$V_e = V_o \left(1 + \frac{f}{2a} L_e\right)^{-1/2} \qquad \text{(D-2)}$$

$$\left(1 + \frac{f}{2a} L_e\right)^{1/2} dL_e = V_o \, d\theta \qquad \text{(D-3)}$$

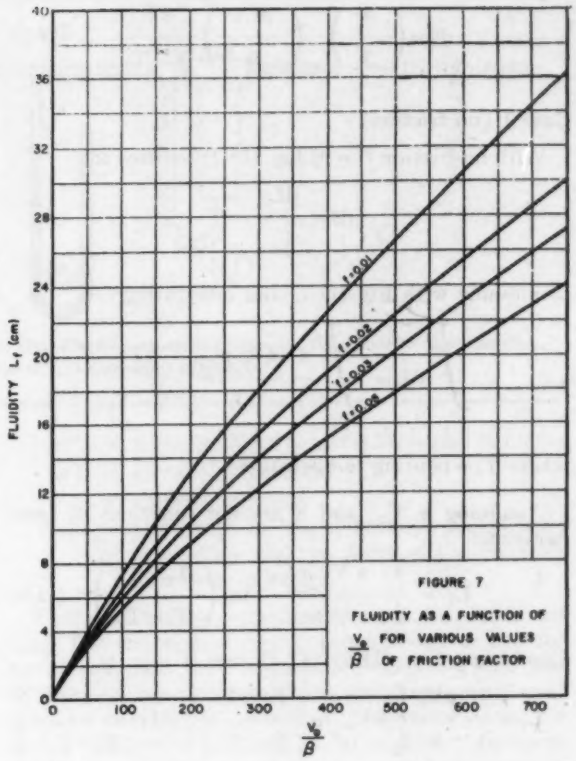


FIGURE 7

FLUIDITY AS A FUNCTION OF $\frac{V_o}{\beta}$ FOR VARIOUS VALUES OF FRICTION FACTOR

Integrating

$$\frac{4}{3}\frac{a}{f}\left[\left(1 + \frac{f}{2a} L_e\right)^{3/2} - 1\right] = V_o \theta \qquad \text{(D-4)}$$

Assuming that the thermal situation is governed by the h-type of heat transfer, a heat balance is performed for a differential element of the tube dL.

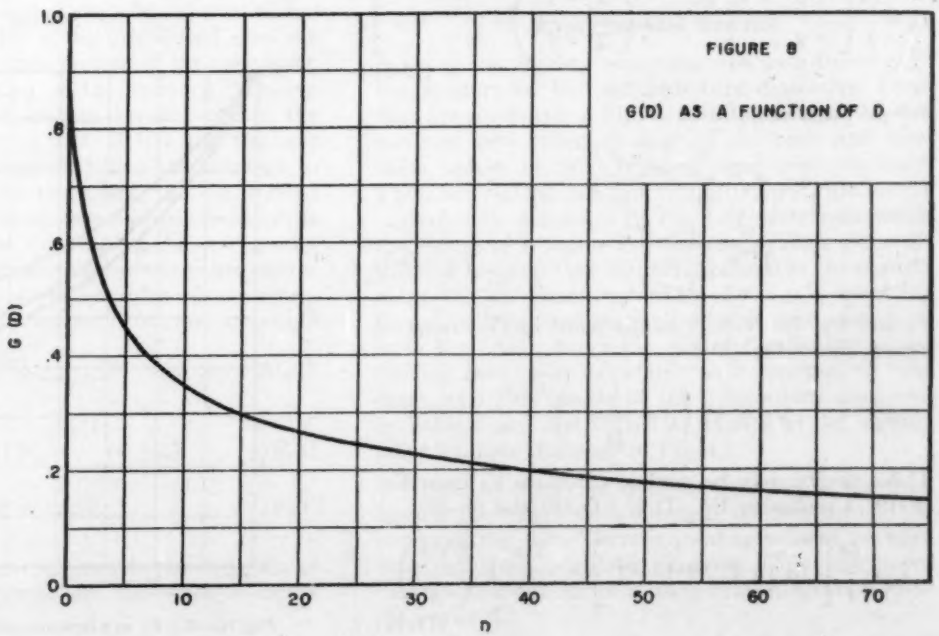


FIGURE 8

G(D) AS A FUNCTION OF D

$$d\theta = \left(\frac{a}{2}\right)\left(\frac{\rho C_p}{h}\right)\frac{dT}{T-T_r} \qquad \text{(D-5)}$$

Case I (no friction)

With no friction (f = 0) Eq. (D-2) reduces to:

$$d\theta = \frac{dL}{V_o}$$

Combining with Eq. (D-7) and integrating:

$$\int_0^{L_1} dL = \int_{T_o}^{T_s} \frac{1}{2}\,\frac{a\,V\,\rho C_p}{h}\,\frac{dT}{T-T_r}$$

where T_o = pouring temperature.

Assuming ρ, C_p, and h are not functions of temperature:

$$L_1 = \frac{1}{2}\,\frac{a\,V_o\,\rho\,C_p}{h}\,\ln\left(\frac{T_o-T_r}{T_s-T_r}\right) \qquad \text{(D-6)}$$

Combining Eqs. (D-6), (D-1), and (C-4):

$$L_f = \frac{1}{2}\,\frac{aV_o}{h}\left[\rho\,C_p\,\ln\frac{T_o-T_r}{T_s-T_r} + \rho H_f\,\frac{1}{T_s-T_r}\right] \qquad \text{(D-7)}$$

Case II (with friction factor constant)

Combining Eqs. (D-3) and (D-5):

$$\left(1+\frac{f}{2a}L_e\right)^{1/2} dL_e = \frac{1}{2}\,\frac{V_o\,a\,\rho C_p}{h}\,\frac{dT}{T-T_r}$$

Integrating

$$L_1 = \frac{2a}{f}\left[\left\{\frac{3}{8}\,\frac{f\,V_o\,\rho C}{h}\,\ln\left(\frac{T_o-T_r}{T_s-T_r}\right)+1\right\}^{2/3} - 1\right] \qquad \text{(D-8)}$$

For purposes of notation:

$$\alpha = \frac{3}{8}\,\frac{f\,V_o\,\rho\,C_p}{h}$$

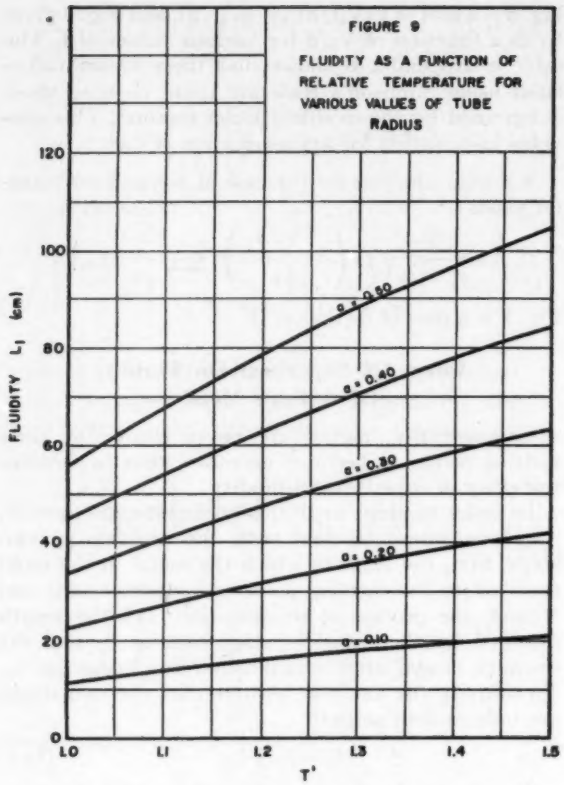
$$T' = \frac{T_o-T_r}{T_s-T_r} \qquad \text{(D-9)}$$

To find L_2, the velocity at the point L_1 (V_1) must be known. From Eq. (D-2):

$$V_1 = V_o\left(1+\frac{f}{2a}L_1\right)^{-1/2}$$

This velocity may be used to calculate L_2 from Eq. (C-19). Combining Eq. (D-1), (D-10) and (C-19):

$$L_f = \frac{2a}{f}\left[(\alpha\,\ln T'+1)^{2/3}-1\right] + \frac{V_1 a}{\beta}\,F\left(\frac{V_1 f}{\beta}\right) \qquad \text{(D-10)}$$



Eq. (D-12) is too involved to show directly the effects of the variables of interest. For this reason Eq. (D-10) was evaluated for various values of T', f, and a. The properties of the metal and of the test assumed were:

$$\frac{\rho H_f}{T_s-T_r} = .400 \text{ cal/cm}^3$$

$$\rho C_p = .800 \text{ cal/cm}^3\,^\circ\text{C}$$

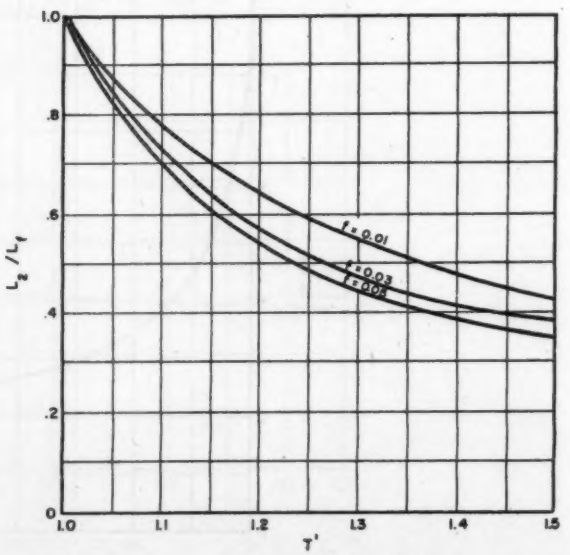


Fig. 10—L_2/L_f as a function of relative temperature.

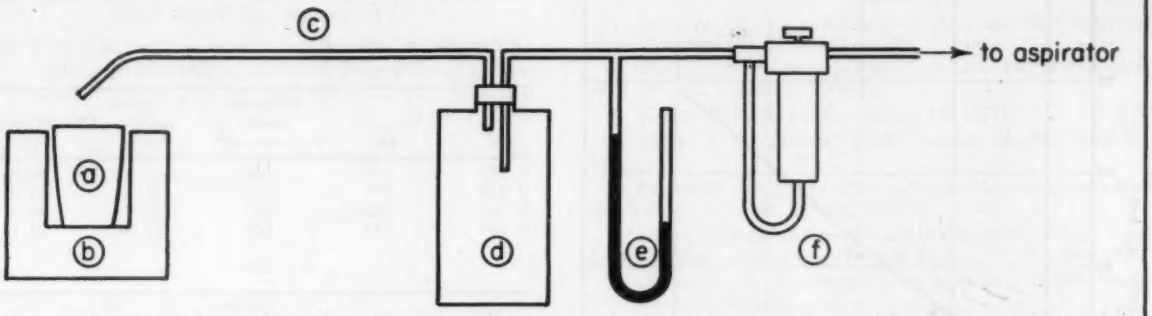


Fig. 11—Schematic diagram of equipment. (a) Crucible of metal; (b) Electric resistance furnace; (c) Fluidity test channel; (d) Pressure reservoir; (e) Manometer; (f) Cartesian manostat.

$$h = .25 \text{ cal/cm}^2 \text{ °C sec}$$

$$V_o = 250 \text{ cm/sec}$$

These are average values selected for the properties of most of the common metals and from experimental determinations. Fig. 9 shows L_f as a function of T′ for various values of a.

Fig. 10 shows L_2/L_f as a function of T′.

Experimental Procedure

In order to evaluate the relationships derived in the previous sections, it was necessary to devise a new fluidity test. The equations were derived assuming a constant and instantaneously-applied pressure head ($P_e - P_2$). It is obvious that this condition could not be met by the present fluidity test in which the application of pressure head is not only noninstantaneous but also varies from test to test if the pouring rate varies. A simple test that fulfills the requirements was devised (Fig. 11). The normal spiral channel was replaced by a pyrex tube (vycor or silica tubes may be used). The pressure head was provided not by a head of metal, but rather by an adjustable partial vacuum applied to one end of the tube. The other end was bent to allow metal to be introduced into the tube while keeping the main portion of the tube level. Instantaneous application of the vacuum (pressure head) was achieved by sealing the entrance of the tube with a thin wafer of wax. Before the test was run, the vacuum was applied. When the entrance to the tube was dipped into the metal, the wax melted and the pressure instantaneously applied. To keep the pressure constant in the tube during the test, a vacuum reservoir was provided close to the other end of the channel. The vacuum was provided by an aspirator, controlled by a Cartesian manostat, and measured with a mercury manometer.

The effective head of metal (ΔP_m) was determined as follows:

$$\Delta P_m = \Delta PH_g \frac{\rho H_g}{\rho_m} - \Delta Z \qquad \text{(E-1)}$$

$$V_o^2 = H = 2g\Delta P_m \qquad \text{(E-2)}$$

where:

ΔPH_g = Difference in pressure between inside of tube and atmosphere, measured in inches of mercury

ΔP_m = Effective head of metal measured in inches

ρH_g = Density of mercury

ρ_m = Density of metal

ΔZ = Difference in height between metal level in crucible and the level of the channel, measured in inches

The pure metals investigated were tin, lead and zinc. The criterion of purity was whether or not the metal solidified with a smooth interface. This was determined by numerous pour-out tests with cold graphite crucibles as the molds. The purity of the tin was at least 99.90 per cent, of the lead 99.95 per cent and of the zinc 99.9 per cent. The pyrex tubes used as molds were sized on the inside diameter to within 0.002 inch of their stated size. Pressure was controlled to within ±0.1 cm of mercury and the temperature measured to within 2°C. To determine L_e as a function of time, and the total freezing time, motion pictures were taken of the course of the fluidity test. A camera speed of 64 frames per second was used. The values of length were read in each frame on a standard microfilm reader.

Experimental Results

Using tin, fluidity was measured as a function of temperature for four different tube diameters. These data are presented in Fig. C-1 in Appendix C. Motion pictures were taken of most of the tests and from them values of final freezing time were obtained. These are plotted as a function of temperature for the various tube diameters in Fig. C-2. Extrapolation of the values of θ_f to the melting temperature gives the value of freezing time for metal poured at the melting point $\theta_{f(m.p.)}$. From the values of $\theta_{f\ (m.p.)}$, assuming h-type of heat transfer, values of h were calculated from Eq. (A-7). From Eq. (C-16) the fluidity at the melting point was calculated as a function of tube diameters. The results of the calculations are given in Table 1 and are plotted as fluidity at the melting point vs. tube diameter in Fig. 12.

Figure 13 shows for one test (conditions listed on figure) the length as a function of time during the course of the test. There is good agreement between observed and calculated curves. (The calculated curve was the result of a numerical integration as in Appendix A.)

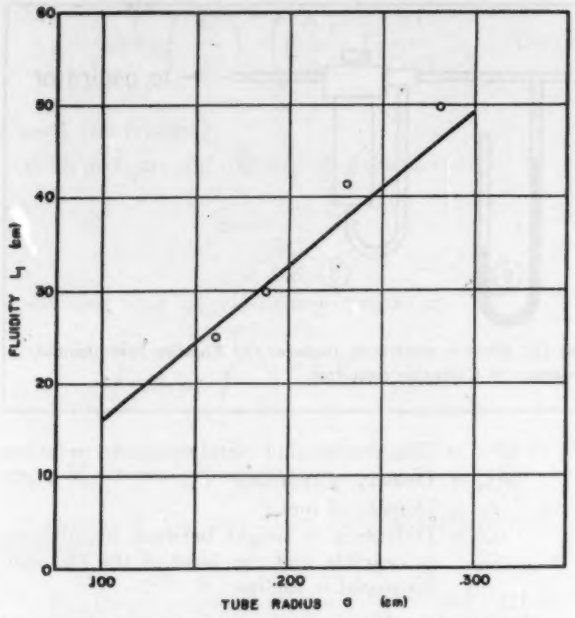


Fig. 12—Fluidity of tin at melting point as a function of tube radius.

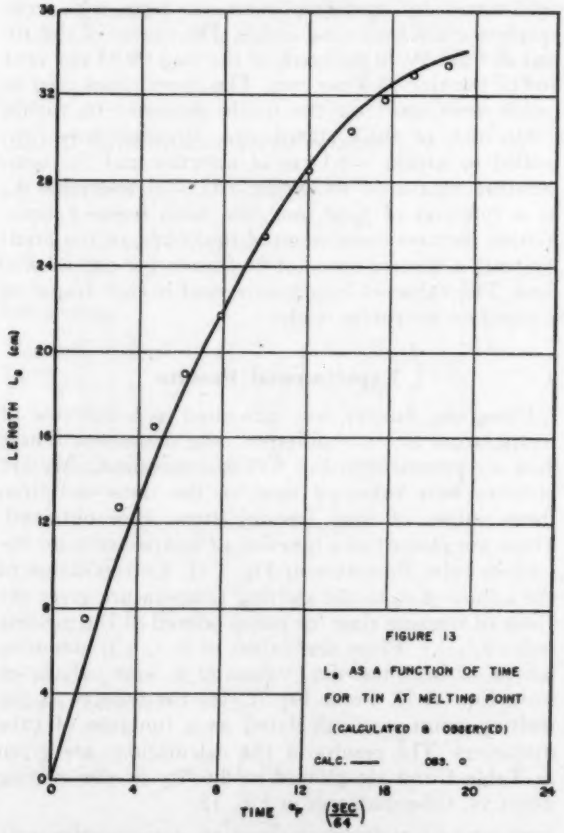


Table 2 presents calculated and observed values for the fluidity of zinc and lead at their melting points.

In order to evaluate Eq. (D-10) which gives fluidity as a function of temperature for h-type heat transfer, metal tubes were used for the test. Metallic tubes provide a greater chill than glass tubes (they have a better thermal conductivity) and thus make contact resistance a larger portion of the total resistance to heat transfer. These metal tubes were bent and sealed with wax as the glass tubes were. Final length was determined by cutting the tubes open. The value of h was calculated from the fluidity at the melting point (extrapolation) by Eq. (C-16). This value was used in Eq. (D-10) to determine fluidity as a function of temperature. The resulting calculated curve and the data are shown in Fig. 14.

TABLE 1—FLUIDITY OF TIN AT MELTING POINT AS A FUNCTION OF TUBE SIZE

$P_m = 10$ in.; $V_o = 222$ cm/sec
$P_{H_g} = 7.25$ in.; $f = 0.023$

a (cm)	θ_f(m.p.) (sec)	h (calculated) (cal/sec-cm^2)	L_f (calculated) (cm)	L_f (observed) (cm)
0.281	.453	.160	46.2	49.5
0.230	.391	.152	38.1	41.4
0.187	.308	.157	30.7	29.7
0.161	.247	.168	26.2	25.0
		avg. 0.159		

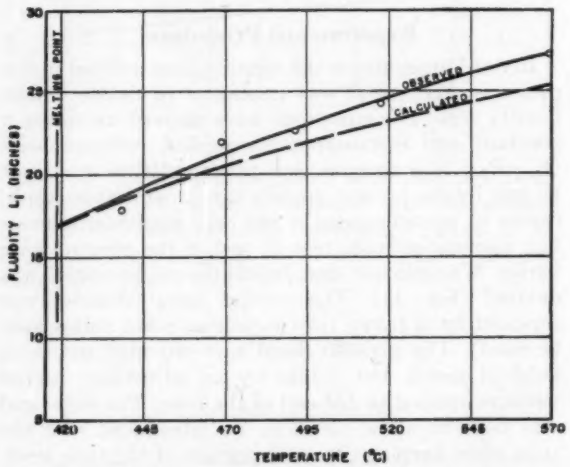


Fig. 14—Fluidity of zinc as a function of temperature in aluminum tubes (I.D., 0.152 in.)

Conclusions

From Analytical Approach

For pure metals, in cases in which the main resistance to heat flow is at the mold-metal interface (chill mold or molds of small diameter):

1. Fludity is a linear function of tube diameter.
2. Fludity at the melting point is a function of the dimensionless term $V_o\ (\rho H_f)/h(T_s - T_r)$ and of

TABLE 2

Metal	a	V_o	h	L_f (calculated) (cm)	L_f (observed) (cm)
Pb	.234	222	.098	31.7	28.9
Zn	.230	222	.0802	53.6	56.0
Zn	.155	289	.0802	44.0	42.5
Zn	.203	289	.0802	57.5	57.3
Zn	.240	289	.0802	68.5	71.2

its product with the friction factor $V_o f (\rho H_f)/h (T_s - T_r)$.

3. The effect of superheat on fluidity is related to the dimensionless product $fV_o\rho C_p/h$) and to the natural logarithm of the relative temperature $(T_o - T_r)/(T_s - T_r)$.
4. The effect of changes in viscosity is small. Since the flow is turbulent,

$$f = 0.185 \left(\frac{DV_o}{\mu}\right)^{-0.20}$$

$$\frac{\Delta f}{f} = 0.20 \left[\frac{\Delta \mu}{\mu}\right]$$

Values of f normally fall between 0.02 and 0.03. Using Fig. 9 and the above equation (for f = 0.025), a change of 50 per cent in viscosity ($\Delta\mu/\mu = 0.50$) amounts to an average of 2.6 per cent change in fluidity.

5. The portion of the total fluidity contributed by the metal after it has reached the melting point is considerable.

For pure metals at their melting point in cases where the main resistance to heat flow is in the mold itself (insulating molds or molds with large channels):

1. Fludity is a function of the dimensionless term

$$\frac{V_o a}{K'\rho' C_p'} \left(\frac{\rho H_f}{T_s - T_r}\right)^2$$

and of its product with the friction factor

$$\frac{V_o f a}{K'\rho' C_p'} \left(\frac{\rho H_f}{T_s - T_r}\right)^2$$

2. For different metals, the fluidity in an insulator such as a sand mold should vary roughly as the square of the volumetric heat of fusion divided by the difference between the melting point and the mold temperature $\left(\frac{\rho H_f}{T_s - T_r}\right)^2$.
3. Changes in viscosity have a small effect on fluidity.
4. Fluidity varies roughly as the square of channel diameter.

From Experimental Approach

For pure metals:

1. Assuming that the main resistance to heat flow is contact resistance, fluidity at the melting point can be successfully predicted in lower-melting metals in glass tube molds.
2. Fluidity is not zero at the melting point. This was determined both from extrapolation to and direct measurement at the melting point. Even when the metal is highly superheated, a substantial portion of the total fluidity is contributed by the metal after it has reached its melting point.

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APPENDIX A

Numerical Solution By Modified Euler Technique

Increments were taken in the independent variable θ. Using Eq. (A-8), values of N were calculated for various friction factors where $N = f(a^4 - r_o^4)/8r_o^4(a - r_o)$

Then

$$\left(\frac{dL_e}{d\theta}\right)^2 (1 + NL_e) = V_o^2$$

Values of L_{en} for the time θ_n (which equals n θ) is given by:

$$L_{e_n} = L_{e_{(n-1)}} + \left(\frac{dL_e}{d\theta}\right)_{(n-1)} \left[\theta_n - \theta_{(n-1)}\right]$$

A modification can be made by recalculating $(dL_e/d\theta)$ based on the average value of L_e during the interval θ to $\theta + \Delta\theta$.

Successive approximations can be made by using the values of $dL_e/d\theta$ determined in the previous approximation as follows. Let $L_e^{(k)}$ represent the kth approximation to L_e.

$$L_{e_n}^{(k)} = L_{e_{(n-1)}}^{(k)} + \left(\frac{dL_e}{d\theta}\right)_{\text{avg}}^{(k-1)} \left[\theta_n - \theta_{(n-1)}\right]$$

where

$$\left(\frac{dL_e}{d\theta}\right)_{\text{avg}}^{(k-1)} = \left[\frac{1}{2}\left(\frac{dL_e}{d\theta}\right)_{n}^{(k-1)} + \left(\frac{dL_e}{d\theta}\right)_{(m-1)}^{(k-1)}\right]$$

The solution converges rapidly as follows:

k	L_t
1	32.76
1 (modified)	28.57
2	29.57
3	29.62
4	29.60

APPENDIX B

Simpson's Rule Section

This system of solution divides the interval of evaluation (c to d) into 2n parts. With $n = d\text{-}c/2n$ and $y = g(x)$

$$\int_c^d g(x) = \frac{h}{3}\left[y_o + 4y_1 + 2y_2 + 4y_3 \ldots\ldots + 4y_{2n-1} + y_{2n}\right]$$

Using $2n = 4$

$$\int_c^d g(x) = \frac{d\text{-}c}{12}\left[y_o + 4y_{\frac{3c+d}{4}} + 2y_{\frac{c+d}{2}} + 4y_{\frac{c+3d}{4}} + y_d\right]$$

The assumption is made that $V_e = \left(1 + \frac{\theta}{\theta_t}\right) V_o$

Considering h-type heat transfer Eq. (A-8), Eq. (B-10) becomes:

$$\frac{dL_e}{dr_o} = -\frac{2V_o r_o}{a\beta}\left[1 + \frac{f}{16}\left(\frac{a^4 - r_o^4}{a - r_o}\right)\left(\frac{V_o}{a^2 r_o^4}\right)\right]^{1/2}$$

Applying Simpson's Rule:

$$L_t = 1/6\ \frac{V_o}{\beta}\left[g(a) + 4g\left(\frac{3a}{4}\right) + 2g\left(\frac{a}{2}\right) + 4g\left(\frac{a}{4}\right) + g(0)\right]$$

$$g(r_o) = \left[1 + \frac{f}{16}\left(\frac{a^4 - r_o^4}{a - r_o}\right)\frac{V_o}{a^2 r_o^4}\right]_{r_o}$$

$$L_t = 1/6\ \frac{V_o a}{\beta}\ 1 + \left(1 + 21.3\ \frac{V_o f}{\beta}\right)^{-1/2} + \left(1 + 1.76\ \frac{V_o f}{\beta}\right)^{-1/2} + \left(1 + .370\ \frac{V_o f}{\beta}\right)^{-1/2}$$

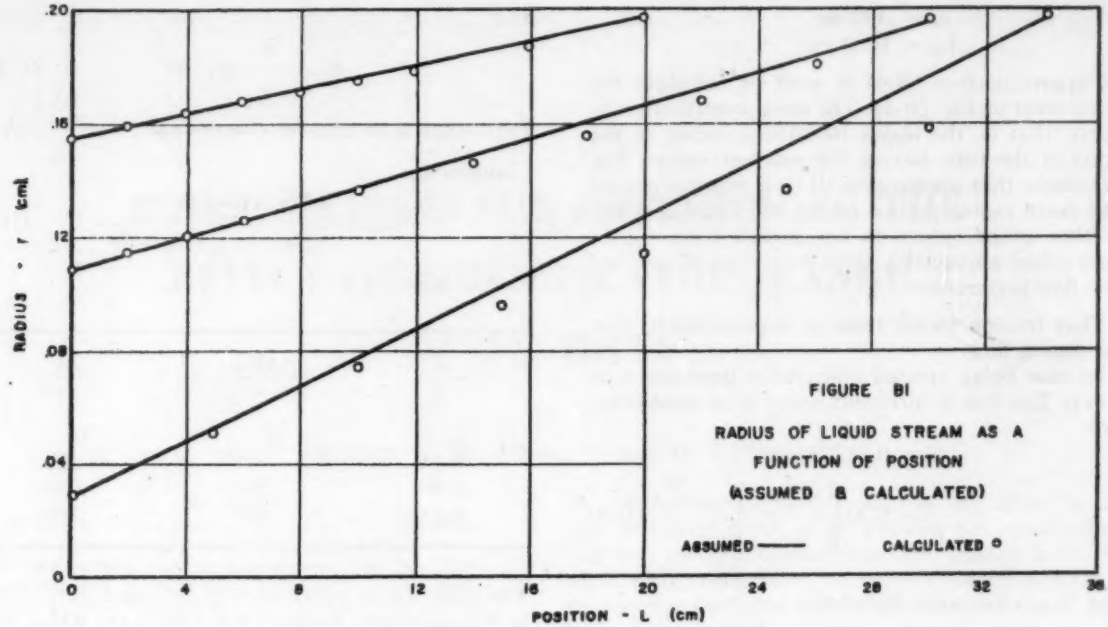


FIGURE B1

RADIUS OF LIQUID STREAM AS A FUNCTION OF POSITION (ASSUMED & CALCULATED)

APPENDIX C

Justification of Assumptions

I. That $r = r_o + \frac{a - r_o}{L_e} L$ (B-6)

The assumption was made that the liquid-solid interface was a conical surface as indicated by the equation above. From the numerical solutions and Eq. (A-8), values of r as a function of L were calculated at various stages in the freezing. The values used for the example were as follows:

$V_o = 37$ cm/sec
$\beta = 1.02$
$a = 0.197$
$f = 0.02$

In Fig. (B-1), these values are plotted along with the lines indicating the assumed values. For the values above

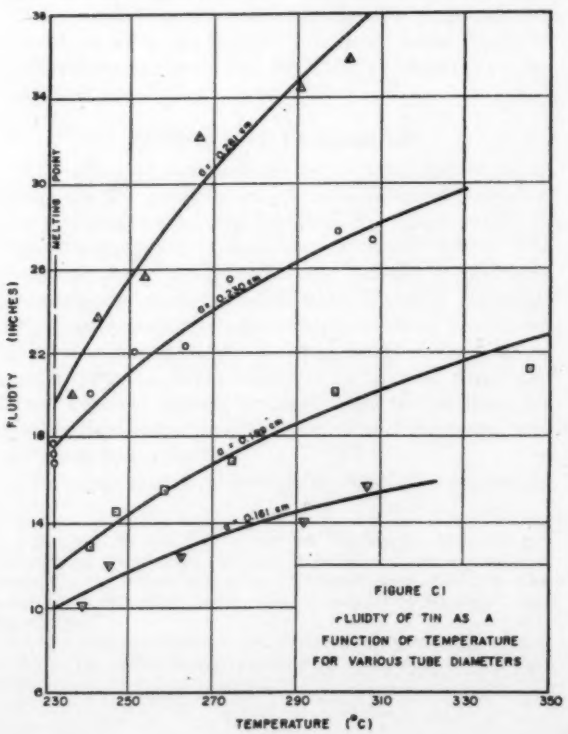


FIGURE C1

FLUIDTY OF TIN AS A FUNCTION OF TEMPERATURE FOR VARIOUS TUBE DIAMETERS

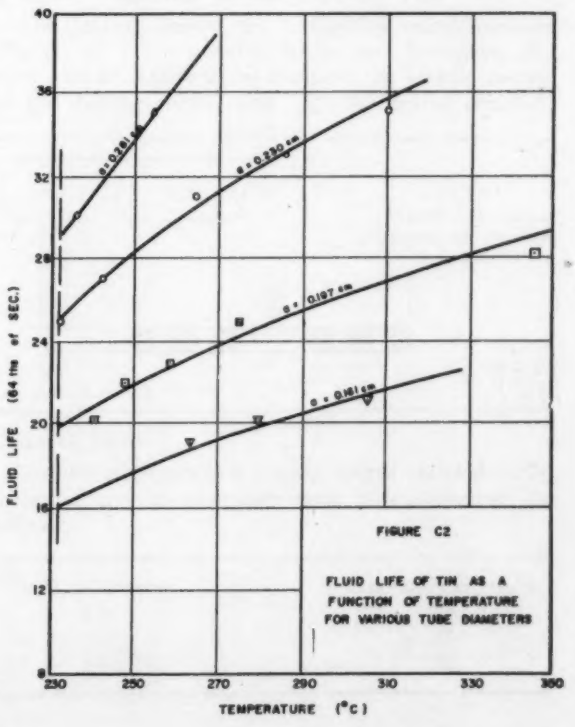


FIGURE C2

FLUID LIFE OF TIN AS A FUNCTION OF TEMPERATURE FOR VARIOUS TUBE DIAMETERS

$$\theta_f = .192 \text{ sec}$$
$$L_f = 34.21 \text{ cm}$$

This approximation (B-6) is used to calculate the friction term of Eq. (B-5). The main contribution to friction (that is, the major throttling) occurs in the portion of the tube having the smallest radius. Fig. (B-1) shows that assumption (B-6) is especially valid in the small radius portion of the tube. In Fig. (B-1) note that actual values do not deviate from the assumed values appreciably until more than 87 per cent of the flow is complete.

II. That friction factor remains approximately constant during flow.

The case being treated concerns a pure metal at the m.p. The flow is turbulent under most conditions, hence,

$$f = 0.185\, N_{Re}^{-0.20}$$

$$\frac{\Delta f}{f} = 0.20\left[\frac{\Delta N_{Re}}{N_{Re}}\right] \qquad \text{(II-1)}$$

where N_{Re} = Reynolds Number $= \dfrac{2V_r\rho}{\mu}$

Now

$$V = V_e\frac{a^2}{r^2}$$

and

$$V_e = \frac{V_o}{\theta_f}(\theta_f - \theta) \qquad \text{(C-13)}$$

For h-type heat transfer $r^2 = a^2 - a\beta\theta$ (A-8)

Combining,

$$N_{Re} = \frac{2a^2V_o\,\rho}{\mu\,\theta_f}\,\frac{\theta_f - \theta}{(a^2 - a\,\theta)^{1/2}} \qquad \text{(II-2)}$$

As θ changes:

$\frac{\theta}{\theta_f}$	$\frac{\Delta N_{Re}}{N_{Re}}$	$\frac{\Delta f}{f}$
0	0	0
0.25	.135	.027
0.50	.293	.059
0.75	.500	.100

The table above indicates that f changes very slowly as θ/θ_f increases. Even at $\theta/\theta_f = 0.75$, (at which time more than 86 per cent of flow is complete) the friction factor has changed only 10 per cent. Under most operating conditions this change in f will not appreciably affect the results.

A NEW METHOD FOR DETERMINING THE EFFECT OF SOLIDIFICATION RANGE ON FLUIDITY

By

D. V. Ragone,* C. M. Adams,** and H. F. Taylor***

ABSTRACT

A fluidity test in which glass tubing was substituted for the classic spiral was used to study fluidity across the lead-tin alloy system. Fluidity and fluid life (total time required to stop the liquid metal stream) vary inversely with solidification range. The pure metals and eutectic had substantial fluidity at the melting point but alloys with large solidification range depended upon superheat for most of their fluidity.

Fluidity is an empirical measure of the distance a liquid metal can flow in a channel before being stopped by solidification. This fluidity is affected by intrinsic properties of the metal such as heat of fusion, melting temperature, composition, viscosity, density, and heat capacity; and extrinsic properties such as gas content, surface oxide films, and insoluble impurities. The variables of the test equipment such as mold material, mold design, and pouring rate are also very important.

This work is concerned with the effect of composition of an alloy on fluidity and with a test designed to eliminate many of the variables of the test equipment that are difficult to control.

Fluidity and Composition

The effect of composition on fluidity has been investigated by pouring metal at a constant temperature interval above the liquidus for many compositions across binary and ternary systems.[20-25,10] The results of these indicate that the fluidity is inversely proportional to the solidification interval (liquidus temperature—solidus temperature). These results indicate that the same relationship holds in the case of ternary systems. With regard to alloys of iron, Andrews, Percival and Bottomley[1] have shown that this relationship holds true for the iron-carbon and the iron-phosphorus systems.

The explanation advanced for this behavior hinges on the manner of solidification of the alloy. In the case of a pure metal or an eutectic, the metal solidifies with smooth regular interfaces inward from the mold wall. Alloys with a solidification range solidify by forming dendrites and by the precipitation of particles from the moving stream thereby impairing the flow considerably. Both these effects tend to shorten the spiral length.

It is interesting to note that viscosity of alloys shows a minimum at the eutectic composition which may be used to explain, at least in part, the high fluidity of eutectic alloys.[36,42] There is some disagreement concerning the extent of the effect of viscosity on fluidity measurements. Desch[4] and Briggs[34] believe the effect to be minor while Bastein, Polyak[9] and Lips[35] indicate the viscosity is an important variable.

Variables of the Fluidity Test

The casting system for a fluidity spiral consists (Fig. 1) of (1) a pouring basin and downgate, (2) some sort of "cushion" to dissipate the kinetic energy of the falling metal, and (3) the spiral channel.

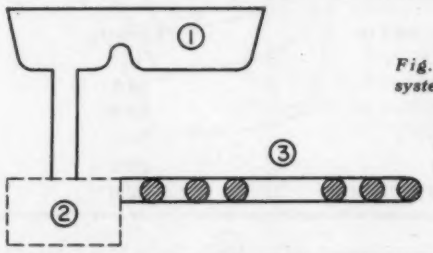


Fig. 1 — Casting system for fluidity spiral.

Channel Shape

Greene[8] indicates that casting length varies directly with the area to perimeter ratio of the channel as follows:

Area/perimeter	Average Length, in.
2.0	23
2.2	27
2.5	32

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Design of the "Cushion"

Three different designs for the cushion were tried by Andrews et al.[1] Under similar casting conditions the three different designs gave values as shown in Fig. 2. This is evidence that the results of different investigators cannot be evaluated together if they used different "cushion" design.

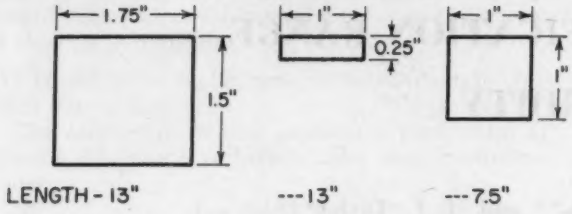


Fig. 2 — Variation of spiral length with "Cushion Design".

Molding Materials

Iron has been used[3,9,11,20] as a mold material for some low melting metals; however, it is more common to use a sand mold. One would expect that the length of the spiral would vary inversely as the "heat abstracting power" of the mold material. Cold iron molds thus would yield shorter spirals. Taylor et al[28] conducted tests which compared several types of sands. They found that green sand gave values which were slightly higher than dried sand molds. There was no noticeable effect of varying the grain size from AFS 63 to 123.

Pouring

Many investigators recognized that pouring speed was an important variable.[3,28,32,41] The length of the spiral varies directly with the pressure head applied;[9] hence it is not unreasonable to assume that the spiral length would vary with the rate of application of the pressure head. Taylor et al[28] have shown the following variation:

Time Required to Fill Mold, sec	Length, in.
1.0	30.5
1.7	28.0
2.2	26.0
3.0	25.0
4.0	22.0

These results are consistent with the observations by Courty[3] and Eunin[12] that the spiral is complete in about one second.

Since it is difficult to apply the pressure head instantaneously, investigators have concentrated their efforts on standardizing the application of pressure head. Evans[6] and Taylor[28] used specially designed pouring basins to accomplish this end, while Courty[3] used a fusible plug in a bottom-pouring system. Worthington[32] used a ladle designed to give constant pouring of metal over the lip. Worthington reports that much of the scatter that plagues fluidity data is eliminated by this constant pouring method.

Experimental Procedure

In the course of some previous work on the fluidity of pure metals, a simple fluidity test was used in order to apply the pressure head of metal immediately to the fluidity channel (Fig. 3). The normal spiral channel was replaced by a pyrex tube (vycor or silica tubes may be used). The pressure head was provided not by a head of metal, but rather by an adjustable partial vacuum applied to one end of the tube. The other end was bent to allow metal to be introduced into the tube while keeping the main portion of the tube level. Instantaneous application of the vacuum (pressure head) was achieved by sealing the entrance of the tube with a thin wafer of wax. Before the test was run, the vacuum was applied. When the entrance to the tube was dipped into the metal, the wax melted and the pressure instantaneously applied. To keep the pressure constant in the tube during the test, a vacuum reservoir was provided close to the other end of the channel. The vacuum was provided by an aspirator, controlled by a Cartesian manostat, and measured with a mercury manometer.

Effective head of metal (ΔP_m) was as follows:

$$\Delta P_m = \Delta PH_g \frac{\rho H_g}{\rho_m} - \Delta Z$$

where

ΔPH_g = Difference in pressure between inside of tube and atmosphere, measured in inches of mercury.
ΔP_m = Effective head of metal measured in inches.
ρH_g = Density of mercury.
ρ_m = Density of metal.
ΔZ = Difference in height between metal level in crucible and the level of the channel, measured in inches.

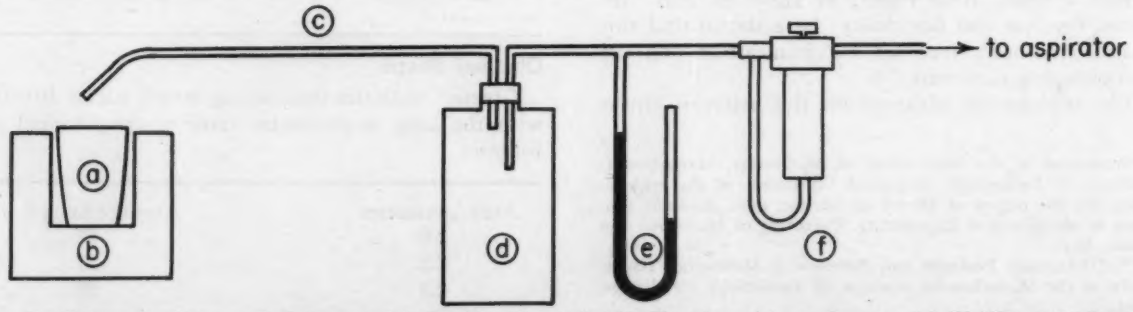


Fig. 3 — Schematic diagram of equipment; (a) Crucible of metal; (b) Electric resistance furnace; (c) Fluidity test channel; (d) Pressure reservoir; (e) Manometer; (f) Cartesian manostat.

TABLE 1 — SUMMARY OF DATA ON Pb - Sb SYSTEM

Alloy	Weight % Sn	Density at Liquidus (grams/cm³)	Viscosity at Liquidus (centipoise)	P_{Hg} for $P_m = 10$ in.	Liquidus °C	Fluidity, in.† Liquidus	25° S.H.**	50° S.H.	Fluid Life (64th of a sec.)
A	0	10.66	2.53	11.00	327.4	11.1	14.4	19.0	18
B	4.00	10.50	2.54	10.80	318	8.5	13.0	16.5	14
C	13.46	10.08	2.54	10.04	296	6.0	9.3	12.0	8
D	18.96	9.78	2.55	10.00	283	3.3	6.5	8.8	3
E	24.32	9.30	2.52	9.60	270	4.0	7.5	10.0	6
F	50.22	8.40	2.45	8.70	220	5.0	9.3	12.5	9
H[1]	62.09	8.17	2.33	8.37	183	22.0	24.5	27.0	38
I*	65.25	7.95	2.36	8.25	187	21.5	24.5	28.0	32
J	69.99	7.70	2.38	8.08	193	12.0	16.5	19.5	21
K*	97.40	7.04	2.00	7.28	228.6	10.0	14.0	16.3	12
L	100	6.98	1.96	7.25	231.9	17.3	22.3	26.0	25

[1] Analyzed for lead, tin content by difference.
*Average of two determinations: 62.04 and 62.13.
†Conditions of test: a = 0.238 cm; P_m = 10 inches.
**Superheat

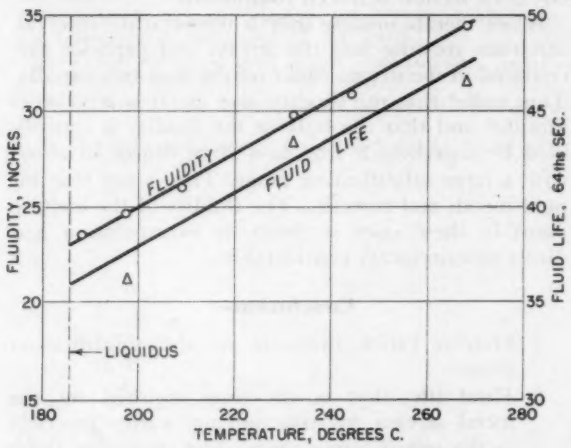


Fig. 4 — Fluidity and fluid life of 65.25% Sn alloy.

This test also enabled the progress of the metal stream in the tube to be recorded with a motion-picture camera. The pyrex tubes used as molds were sized on the inside diameter to within 0.002 inch of their stated size (0.187 inch). Pressure was controlled to within ± 0.1 cm of mercury and the temperature measured to within 2 C. To determine length of flow as a function of time, and the total freezing time, motion pictures were taken of the course of the fluidity test. A camera speed of 64 frames per second was used. The values of length were read in each frame on a standard microfilm reader. The purity of the tin used in these experiments was at least 99.9 per cent and that of the lead 99.95 per cent.

For each alloy listed in Table 1 a curve of fluidity as a function of temperature was prepared using the above technique (Example: Figure 4). The alloys were superheated about 150 C and as they cooled slowly in a resistance furnace fluidity measurements were made. The vacuum applied was adjusted to account for density differences in the alloy systems. These densities as well as values for viscosity were reported by H. J. Fisher[42] and are recorded in Table 1. The values of fluidity and fluid life at the liquidus and at 25 C and 50 C superheat were taken from these curves and plotted against composition in Fig. 5 and 6.

Discussion of Results

There are many advantages to using the fluidity test described. Among these are:

1. The pressure head is instantaneously applied, and is constant. Human factor in pouring is eliminated.
2. Little metal is used so that many tests can be made from one crucible of metal while it is cooling.

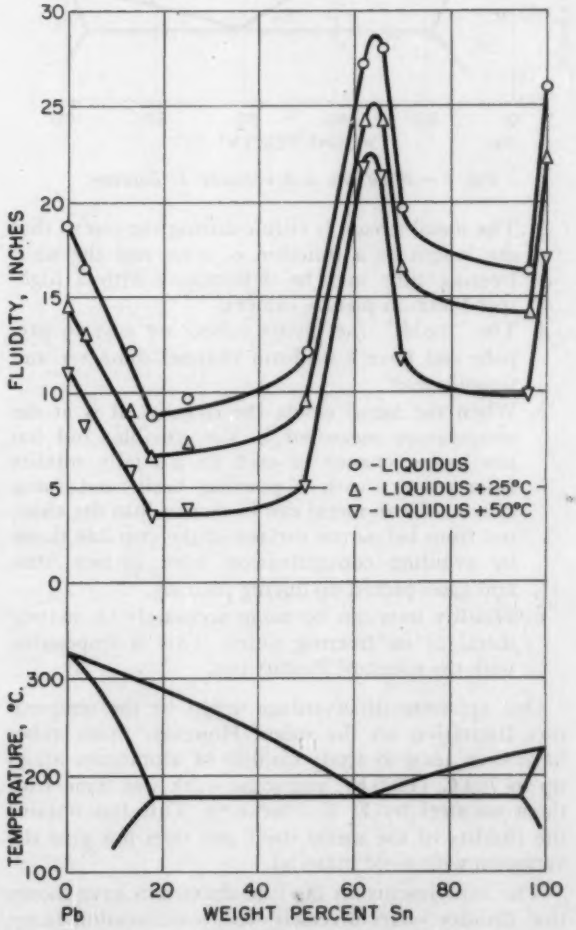


Fig. 5 — Fluidity of Pb-Sn System.

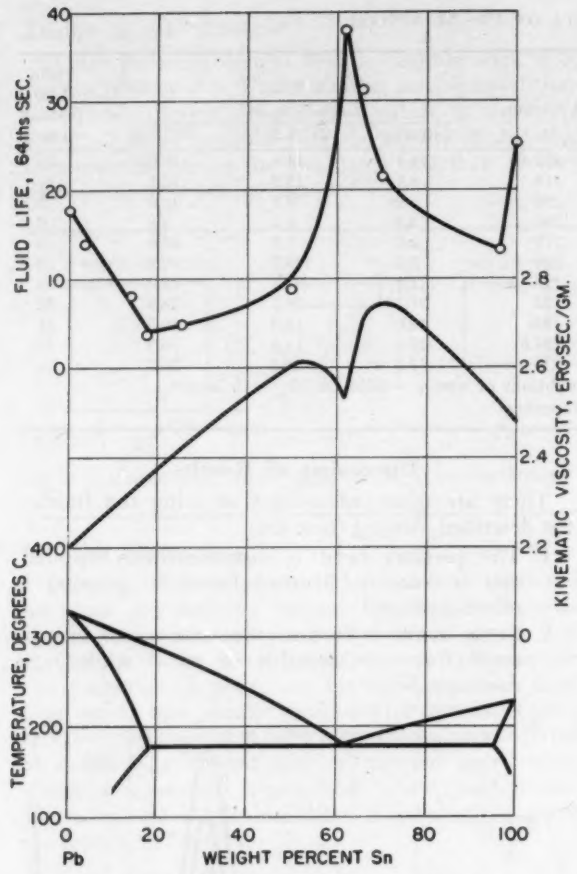


Fig. 6 — Fluid life and viscosity at liquidus.

3. The metal stream is visible during the test so that the length as a function of time and the total freezing time may be determined with a high-speed motion picture camera.
4. The "molds" (i.e. pyrex tubes) are easy to prepare and have a uniform channel diameter and smoothness.
5. When the metal enters the channel, it is at the temperature measured in the crucible, and has not had a chance to cool or partially solidify from contact with a pouring basin and sprue system. Clean metal can be drawn into the channel from below the surface of the crucible thereby avoiding contamination from surface films and gases picked up during pouring.
6. Fluidity tests can be made accurately on molten metal at its freezing point. This is impossible with the standard fluidity test.

One apparent disadvantage might be the temperature limitation on the tubes. However, pyrex tubes have been used to study fluidity of aluminum alloys up to 760 C (1400 F) and some work was done with them on steel by E. E. Hucke.[48] This test obtains the fluidity of the metal itself and does not give the variation with mold material.

The experiments on the lead-tin system have shown that fluidity varies inversely with solidification range both at the liquidus and in the liquid region (Fig. 5). Fluid life, the time required to stop the stream, varied in a similar manner (Fig. 6). This diagram also shows the variation of kinematic viscosity with composition. These viscosity variations may contribute to the gross effect, however, it is felt that they are only a minor factor. Previous work[47] has shown that changes of 50 per cent in the viscosity affect the fluidity a matter of 3 per cent. The maximum change in viscosity in this system is 25 per cent and the changes of fluidity on the order of hundreds of per cents. Since the fluid life varies inversely as the solidification range, it is logical to conclude that mode of solidification affects the fluid life and hence the fluidity of alloys. When metal solidifies with a smooth interface (at one temperature) as pure metals and eutectics do, the fluid life is high, for the channel will be choked off only when all the metal across one cross section is frozen completely.

When metals solidify over a temperature interval, dendrites growing into the stream and particles precipitated in the stream choke off the flow very rapidly. Thus belief that the fluidity of a metal is zero at its liquidus and that the bulk of the fluidity is contributed by superheat is true to a large degree in alloys with a large solidification range. This is not true for pure metals and eutectics. The fluidity of the melting point in these cases is (both by extrapolation and direct measurement) considerable.

Conclusions

1. Fluidity varies inversely as the solidification range.
2. Fluid life, that is, the time required for the metal stream to stop moving, varies inversely as the solidification range. The change in fluidity with solidification range is thus due to change of mode of solidification and not to changes in viscosity.
3. With alloys having a wide solidification range, the major portion of the fluidity is due to the superheat above the liquidus.
4. The fluidity test used in this work has many advantages over the standard fluidity test. This new test should be considered for practical foundry use.

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MECHANISM FOR PORE FORMATION IN SOLIDIFYING METALS

By

W. D. Walther,* C. M. Adams,** H. F. Taylor***

ABSTRACT

An analysis has been derived to predict the feeding range of cast bars, plates, and tapered bars. The analysis is proposed to present an insight into the origin of porosity in castings. Porosity may be a consequence of gas dissolved in molten metal and evolved during solidification; however, because of limitations in feeding range, porosity may develop in the near absence of gas in solution, or porosity may form as an interaction of feeding limitations and dissolved gas.

The results predicted by the analysis were checked experimentally. Very pure aluminum, magnesium, lead, tin, and Armco iron were cast to the shapes, long bars, large flat plates, and tapered bars. Some of the conclusions which verified predictions were (1) plates are more easily fed than bars, (2) a very slight taper is sufficient for feeding a bar of pure metal to complete soundness (taper of 0.03 in./in.), and (3) gas in solution limits and feeding range of a metal.

I. Introduction

Research in foundry metallurgy during the past decade has aided the foundryman to increase strength and acceptability of castings. Quantitative data and calculation have determined optimum riser size for a casting; [1, 2] also, in principle, optimum location and spacing between risers [3, 4] for many cases in the steel foundry have been determined.

In non-ferrous foundries the importance of gas porosity has been clearly established, and means for eliminating this source of trouble have been demonstrated.[5] Recently, a tool for quantitative estimation of gas content has been described, [6] which should further aid the foundryman. Recent researches [7, 8] have illustrated the importance of aiding the feeding conditions for all castings, in order to achieve maximum mechanical properties.

Some of these researches have treated the influence of feeding range on soundness, others have treated the influence of gas porosity. The interrelationship between gas content and feeding range have not been considered even qualitatively. However, under difficult conditions of feeding, porosity is more easily developed, either as interdendritic shrinkage, or as a combination of shrinkage and gas evolution. For example, in aluminum alloys having relatively long solidification ranges and high thermal conductivity, dispersed porosity or shrinkage occurs readily, especially when these alloys freeze under low temperature gradients.

On the other hand, in pure metals or alloys of narrow liquidus-solidus interval of freezing, dispersed porosity is most uncommon; rather, shrinkage is localized, either as "pipe" or as centerline shrinkage.

Fundamentally, both types of cavity formation, dispersed porosity (interdendritic porosity) and centerline shrinkage are related phenomena. They are usually considered apart because centerline shrinkage is associated with the geometry of the casting, while dispersed shrinkage is not related to shape. Both types of defects in castings may find origin in (a) gas evolution, (b) shrinkage in the absence of gas, or (c) as a combination of both shrinkage and gas evolution. To assess the relative importance of gas evolution and shrinkage as causes of unsoundness in a casting, something must be known of how a hole will form in a solidifying casting (other than that of simple gas evolution resulting from supersaturation). To verify analytical predictions quantitatively, pure metals were used. The advantage of using pure metals is that during solidification only a single liquid channel need be considered, and cavities formed are located along the thermal center or centerline of a casting.

A. Mechanism for Hole Formation in a Casting

Consider the long cylindrical casting and the riser shown in Fig. 1. The casting is long enough that during solidification, a constant temperature zone develops over a great portion of the length. The riser is of sufficient size to feed solidification shrinkage. If this long cylindrical mold is filled with a pure liquid metal such as iron, aluminum, or tin, a thin solid shell of metal will form soon after pouring. Within this shell is contained the remaining pure liquid metal. With metal of sufficient purity (and in the absence of

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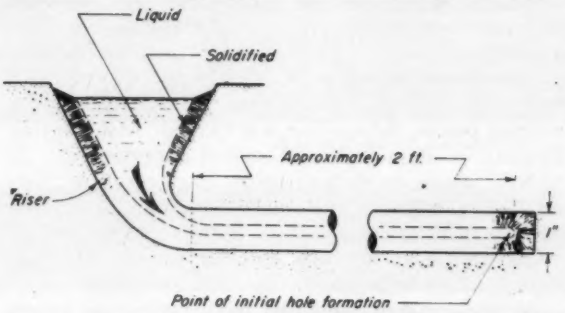


Fig. 1—Schematic representation of long cylinder casting with riser for investigating cavity formation along the centerline of pure metal cylinder castings.

undercooling), the liquid-solid interface will have a smooth cylindrical surface. As solidification proceeds a thinner and thinner liquid interior results, so that late in the process a long thin cylinder of liquid metal remains, contained within the already solidified shell.

The volumetric change on solidification requires a flow of liquid metal from the riser reservoir into the long thin liquid metal channel in the casting. Fluid velocity, acceleration of fluid particles, and frictional resistance to flow develop, and all result in a pressure drop along the length of the cylindrical casting. When the pressure drop becomes large enough, liquid metal ruptures and a cavity is formed in the casting.

This condition is determined mathematically by relating both heat flow and fluid flow principles. Qualitatively the factors may be stated as follows:

(1) Rate of heat loss determines how fast the casting will solidify. Heat loss from the casting into a sand mold has been determined independently of fluid flow considerations.

(2) Rate of heat loss determines solidification rate, and solidification rate and solidification shrinkage determine how much and how fast the casting requires feed metal from the riser.

(3) With increasing rate of solidification, as occurs during late stages of freezing (or in thin sections), a liquid metal particle must move down the remaining liquid channel from the riser into the casting at increased velocity, under increasing acceleration, and also, under increasing frictional resistance to flow. These combined factors cause a pressure drop, and the pressure drop increases as the liquid channel gets smaller. At a certain radius of remaining liquid channel a large enough pressure drop arises to produce a negative pressure or tensile force on the liquid. This tensile force or negative pressure causes rupture of the liquid metal. With gas in solution in the metal, a smaller pressure drop allows rupture of the liquid metal.

B. Derivation of Equation Governing Hole Formation

An equation governing hole formation in the cylindrical casting of Fig. 1 was derived, and is included in Appendix I. The equation developed relates pressure drop along the length of the casting to the radius of liquid channel at the time the first hole forms, and is as follows:

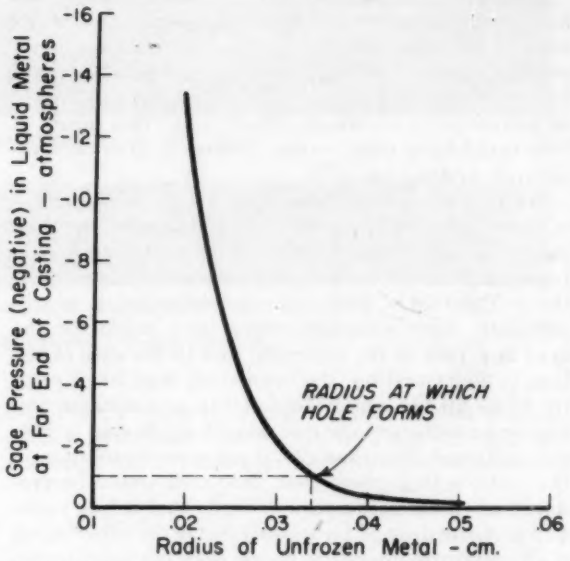


Fig. 2—Effect of Channel radius on pressure within lead cylinder casting. The graph shows the rapid pressure drop accompanying a change in amount of pure metal remaining liquid. Very large negative pressures result late in solidification.

$$\Delta P = \frac{\rho_e}{g} \cdot \frac{64\gamma\beta^4 L^2}{r^4} \left[\frac{1}{2} + \frac{\gamma f L}{3r} \right] \qquad (1)$$

where ΔP = pressure drop along the length of casting Fig. 2

ρ_e = weight density of liquid metal at its freezing point

γ = solidification shrinkage factor (3-6 per cent for most metals)

f = friction factor

L = length of liquid zone in casting

r = radius of liquid core in partially solid casting

β = constant (see Appendix I)

Since cavity size is related to pressure, and thus equilibrium solubility of gas, any gas in solution in the melt, such as hydrogen, allows a smaller pressure drop before rupture of liquid metal than is the case without gas in solution. Consequently, with gas in solution a larger cavity may be formed. However, at a very small radius of hole, a large negative pressure may develop, and the resultant tensile force on the liquid metal could cause a hole to form even in the absence of gas. This possible condition is apparent from a plot of pressure drop versus hole size, shown in Fig. 2. Figure 2 is a graphical representation of Equation (1).

Also of note in Equation (1) is that pressure drop is proportional to the square of the length of the liquid channel. Cavity formation will occur at a point in the liquid farthest from the riser (since pressure drop is greatest there), and the longer the liquid zone, the greater the hole size will be.

A numerical evaluation of Equation (1) for a given pure metal shows the velocity component to be almost negligible, and the acceleration and frictional resistance terms of similar magnitude, depending upon length of channel, radius of channel, and magnitude of solidification shrinkage. Practically, this means a hole could form even in the absence of frictional resistance to flow.

Some of the assumptions upon which this equation is based are: (1) No undercooling, (2) pure metal so that a smooth liquid-solid interface exists, and (3) a region of constant temperature over a long segment of the cylinder at a time when solidification is nearly complete. Approximations have been made for the heat flow part of the equation, and in the case of the long cylinder casting, the evaluation may be in error by 10 to 20 per cent. Most of the assumptions and approximations err on the side of predicting a hole size too small. For example, if gas were in solution in the metal a larger hole than predicted would be produced when the assumption is made that a hole forms at a pressure drop of 1 atmosphere. On the other hand, if a constant temperature region does not exist over a sufficiently long length of the cylinder casting, calculated hole size may be too large. Also, high purity metal, in the absence of any nucleating agents, may withstand very high tension (or negative pressure); in this event, prediction of hole size on the basis of breakdown at zero atmosphere would account for too large a cavity formation.

II. Experimental Evidence

A. Long Cylinder Casting

In order to verify size of cavity predicted by Equation (1), a simple experimental procedure was adopted. Pure aluminum (99.995%), pure magnesium (doubly distilled), pure tin, killed Armco iron, and pure lead were poured into the long, thin cylinder mold shown in Fig. 1. The long thin cylinder assured a constant temperature region over a major length of the tube. In two cases, as noted later, both pure aluminum and magnesium were poured into a pyrex glass tube liner (dried at red heat) so that gas pickup during pouring and filling the mold was minimized. Otherwise, pure metals were poured into silica sand molds bonded with clay, and fired to red heat prior to pouring. All molds were tilted at a 10-degree angle so that molten metal would flow uphill to minimize turbulence and air entrapment. Aluminum and magnesium were degassed with chlorine gas prior to pouring. Metal was checked qualitatively for gas content using a reduced pressure test described elsewhere.(9) All metals, except Armco iron, were poured 0 to 10°C above their respective melting points. Armco iron was "killed" with varying amounts of aluminum, and poured approximately 25 to 50°C above its melting point. Each resulting casting had 5/16 inch machined off the cope and drag side. Thus, a slice 3/8 inch thick was obtained from the center section of each cylindrical casting; this slice was radiographed to determine hole size and distribution.

A comparison of calculated and experimental values for size of cavity in the long cylinder casting of Fig. 1 is presented in Table 1. The distance from the riser at which the observed centerline cavity first forms is also tabulated in Table 1. It was for this distance that predicted radius was also calculated, assuming hole formation at a pressure drop of one atmosphere.

TABLE 1—RADIUS OF HOLE AT WHICH BREAKDOWN OCCURS IN CYLINDRICAL CASTING OF FIG. 1.

Cyl. Diam. = 1 In.

Metal	Predicted Radius (Assuming Breakdown) at Δ P = 1 atm., cm	Observed Radius, cm	Distance of Largest Cavity from Riser, cm
Mg	0.050	0.059	61
Al	0.028	0.0286	41
Fe	0.032	0.036	50
Sn	0.019	0.03	41
Pb	0.057	0.059	55

The agreement between observed and calculated radius of cavity is better than can reasonably be expected considering some of the assumptions and approximations made in deriving the equation. Perhaps, some compensation of errors develops to bring agreement so close. Experimental values of centerline cavity radius given in Table 1 were the smallest obtained for each metal. With less care exercised, hole size was larger than shown. A comparison of change in hole size for magnesium and aluminum is shown in Figs. 3 and 4. Figures 3a and 4a are radiographs of sections of aluminum and magnesium cylinder castings obtained when the respective metals were degassed and poured into baked silica sand molds, while Figs. 3b and 4b are from sections of aluminum and

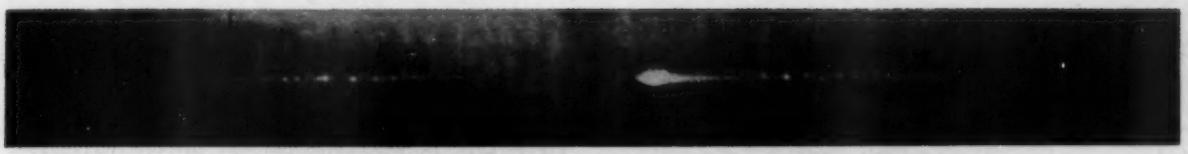


(a) High-purity magnesium degassed and poured into dry sand molds.



(b) High-purity magnesium degassed and poured into dry pyrex tube.

Fig. 3—Radiographs of sections from long cylinder magnesium castings.



(a) High-purity aluminum degassed and pured into dry sand mold.



(b) High-purity aluminum degassed and poured into dry pyrex tube.

Fig. 4—Radiographs of sections from long cylinder aluminum castings.

magnesium cylinder castings cast in pyrex tubes after degassing. The difference between Figs. 3a and 4a are quite marked; Figure 3a shows bubbles along the centerline formed due to gas evolution, and Fig. 3b shows a very fine cavity formed along the centerline, of radius very nearly predicted by Equation (1). The difference between Figs. 4a and 4b is not so marked, but the difference in hole size is still quite evident. This difference is due primarily to gas pickup from the baked silica mold. Figures 5a and 5b show a corresponding difference for two heats of Armco iron; sample 5b at a higher state of deoxidation. All these radiographs illustrate pictorially the effect of increased gas content on hole size in castings, results which demonstrate what the numerical evaluation of Equation (1) predicted.

B. Evaluation of an Infinite Plate Casting

A mathematical analysis for a plate casting, of such dimensions as to approximate an infinite plate, predicted that plate castings should feed much more easily than a cylinder casting. The derivation of the equation governing a plate casting is similar to that for a long cylinder; the equation derived is as follows:

$$\Delta P = \frac{\rho_e}{g} \cdot \frac{16\gamma\beta^4 L^2}{x^2 \dagger^2} \left[\frac{1}{2} + \frac{\gamma f L}{6x}\right] \qquad (2)$$

where x = ½ thickness of cavity formed in a plate casting

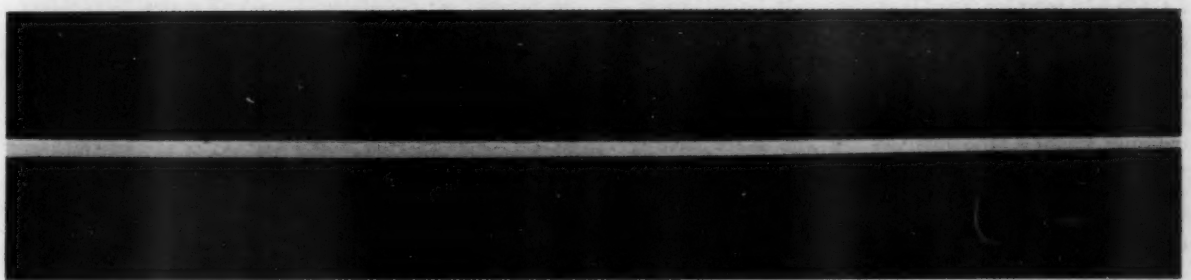
$\dagger$ = thickness of plate casting.

All other factors have been previously described. Table 2 gives predicted cavity size in a plate casting and shows a comparison with the cylinder casting, and also, with experimental values of hole size obtained with plate castings. Cavity size is 10 to 20 times smaller for a plate casting than for a cylinder whose diameter is equal to the plate thickness.

A radiograph was made of a pure magnesium plate casting 14 in. x 8 in. x ⅞ in. This radiograph, made with the x-ray beam normal to the center plane of the casting, ostensibly verified the calculated prediction



(a) Well deoxidized casting has hole size approximating theoretical.



(b) Incompletely deoxidized casting has too large a hole.

Fig. 5—Radiographs of sections from cylindrical Armco iron castings.

TABLE 2—TABULATION OF CALCULATED AND EXPERIMENTAL HOLE SIZE FOR PLATE CASTING OF FIG. 6. CALCULATED VALUE FOR HOLE SIZE OF CYLINDER CASTING IS GIVEN FOR COMPARISON. CYL. DIAM. = PLATE THICKNESS

Metal	L (cm)	Calculated Hole Size for Plate at $\Delta P = 1$ atm (cm)	Experimental Hole Size for Plate (cm)	Calculated Hole Size for Cylinder (cm)
Fe	25.4	0.0032	Too small for accurate measurement	0.047
Mg	25.4	0.0058	0.006	0.084
Sn	25.4	0.0019	Too small for measurement	0.032

that any cavity formed would be very small. In fact, if any cavity was formed it was beyond the sensitivity of the radiograph to record. Therefore ⅜-in. strips were cut from the middle of the plate (see Fig. 6), the strips machined smooth, and an exposure made with the strips turned 90 degrees so that the x-ray beam was parallel with the center plane of the plate. The exposure was made in this way to obtain maximum sensitivity. Nonetheless, the cavities were too small to show up clearly, and in order to estimate their thickness, magnification of the radiograph was necessary.

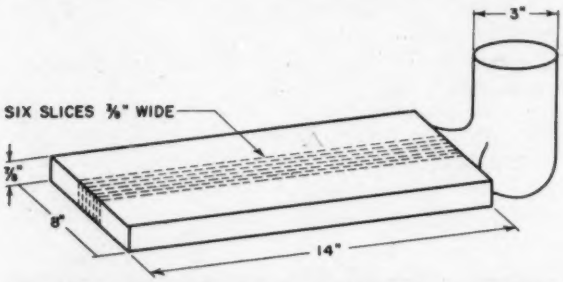


Fig. 6—Schematic drawing of plate casting used for evaluating the mathematical analysis. Slices were cut from the plate casting, as shown, for radiographic determination of any cavity formation.

Similar experiments were performed for tin and iron with similar results as obtained with magnesium. These results correlate well with the findings of Bishop [10] for steel plates and bars. Bishop found that adequate risers feed plates further than bars, or at a given distance hole size in a plate casting was much smaller than that of a bar. (A mathematical analysis for long square bars has been derived, and it shows, to a close approximation, the same results as a long cylinder). In contrast to the analysis for a cylinder casting, the analysis for the plate casting shows the friction component accounts for the major part of the resultant pressure drop.

C. Long Casting With a Shape the Frustrum of a Cone

Padding is a well-known device for eliminating centerline shrinkage in steel. To make a sound cylindrical casting, padding is added, giving a casting the shape of a frustrum of a cone. Padding promotes directional solidification from the end of a casting to the riser, and in effect, produces a tapered liquid region with decreasing amount of liquid as distance from the riser increases. An analysis which approximates this condition was derived. In the derivation, the assumption was made that the liquid region decreases in cross-section linearly from the riser to the end of the casting. The liquid region also defines the shape of a frustrum of a cone. (Actually, if external dimensions of a casting define a frustrum of a cone, the interior liquid region, during solidification, more nearly approximates a paraboloid). As before, the analysis is given in terms of pressure drop down the length of the cylinder:

$$\Delta P = \frac{\rho_e}{g} \cdot \frac{32\beta^4\gamma}{3k^2(r-kl)^2} + \frac{128\mu\beta^2\gamma}{3k^2(r-kl)^2} \qquad (3)$$

where k = taper coefficient

$(r-kl)$ = change in radius with length

μ = viscosity of the liquid

and all other terms have been previously defined.

Significantly, Equation (3) predicts a relatively small "k" value or taper is required for effective feeding of a long frustrum of a cone-shaped casting. Even so, the equation predicts a taper larger than is necessary. Equation (3) predicts a taper of about 0.03 inch per inch should be sufficient for a sound magnesium casting, and slightly less than 0.03 inch per inch for pure iron.

This prediction was tested experimentally by making a 20-inch long frustrum of a cone with a taper of 0.03 inch per inch and 1-inch diameter at the small end. Castings of Armco iron killed with 0.1 per cent aluminum and castings of pure magnesium were poured into baked sand molds. Thin slices ⅜ inch thick were obtained from the center and radiographed. As predicted, no centerline cavities were discernible in the radiographs.

Brinson and Dumas [11] determined empirically that a taper of about 0.1 inch per inch was necessary to make sound plate castings of medium carbon steel. For frustrum of a cone-shaped casting an even larger taper would be necessary to give complete soundness than for a plate. The discrepancy between their empirical determination and that found here lies in the higher purity metal used in this investigation. Underdeoxidized or "gassy" steel could account for some of the discrepancy, although introduction of dendrites along the centerline would cause much greater frictional resistance to flow, and appears the more plausible explanation.

III. Conclusions

A mechanism has been proposed to account for cavity formation in a solidifying casting. Experimental verification of the analysis has been obtained using pure aluminum, magnesium, iron, tin, and lead. In line with the empirical work of Bishop [10] on the feeding of steel castings, the analysis and results show that a casting the shape of a plate or slab is much easier to feed than a long cylinder or bar.

The analysis also gives an indication of the importance of gas in solution on the extent of centerline shrinkage and interdendritic shrinkage. With higher gas contents, the extent of both centerline and interdendritic shrinkage is greater.

Qualitatively, the analysis gives an insight into the feeding of castings. In particular, it allows a more complete understanding of the formation of interdendritic shrinkage in alloys of long solidification range. On the basis of the analysis, magnesium alloys should be more prone to interdendritic shrinkage than others. In practice, because of the more difficult gas problems, aluminum alloys exhibit interdendritic porosity to an even greater extent. The quantitative effect of different properties of metals, such as volume heat of solidification, density, and solidification shrinkage, on feeding of castings is given. A clearer understanding is obtained of the effect of frictional resistance on the tendency for porosity formation in castings.

The analysis may be applied to steel and other metal systems with narrow solidification range and relatively low thermal conductivity. This would entail an appropriate approximation for an increased friction factor and acceleration component with dendritically freezing alloys. A direct application to alloys of long solidification range and "mushy" type freezing would be more difficult.

Shape of the liquid region was found to be important in the feeding of castings. A long cylinder casting was found much more difficult to feed than a flat plate casting. A slight taper on the cylinder casting permitted complete feeding of solidification shrinkage. A much smaller taper would suffice for a plate casting.

APPENDIX

Derivation of Equation Governing Pressure Drop in a Long Cylinder Casting During Solidification

As described the derivation depends upon an analysis of heat flow from the solidifying casting, and an analysis of fluid flow of molten liquid along the channel within the partially solid casting. Equations governing heat flow are required to determine rate of solidification and, consequently, rate of flow of molten metal along the length of casting. The mass flow of molten metal along the length of casting arises from the volume change on solidification.

A. Heat Flow

Rate of heat flow into a sand mold at the mold-metal interface is given approximately by the following equation:[12]

$$q = \frac{K\Delta T_m}{\sqrt{\pi\alpha\theta}} \cdot 2\pi RL \qquad (a)$$

where q = rate of heat flow into sand mold
K = thermal conductivity of sand
α = thermal diffusivity of sand
ΔT_m^* = temperature of the melting point minus ambient temperature
R = radius of cylinder
L = length of liquid zone in cylinder.

Assumptions governing this equation have been treated elsewhere.[12] Freezing rate of the cylinder casting depends upon rate of heat flow from the casting:

$$q = -\rho_s \Delta H \cdot 2\pi rL \frac{dr}{d\theta} \qquad (b)$$

where ρ_s = weight density of solid metal
ΔH = heat of solidification
r = radius of liquid cylinder in partially solid casting
θ = time.

Equation Equations (a) and (b) to find linear solidification rate,

$$\frac{dr}{d\theta} = -\frac{\beta R}{\sqrt{\theta}-r} \qquad (c)$$

where $\beta = \dfrac{K\Delta T_m}{\rho_s \Delta H \sqrt{\pi\alpha}}$ = constant.

Integrating Equation (c) between limits, $\theta = o$, θ; $r = R$, r

$$r^2 = R^2 - 4R\beta\sqrt{\theta} \qquad (d)$$

Equation (d) gives radius of liquid channel contained within the casting of radius R as a function of time.

B. Fluid Flow of Metal in Partially Solid Casting

At some fixed location along the length of the solidifying cylinder casting, mass flow at that location is determined by solidification shrinkage of the volume of liquid metal beyond that location, or:

feed metal supply − feed metal demand

$$\pi r^2 \bar{V} = -\gamma \cdot 2\pi r(L-x)\frac{dr}{d\theta} \qquad (e)$$

where x = distance from riser
$\bar{V}$ = average velocity of metal at position x
γ = % shrinkage/100.

Combining Equations (c), (d), and (e),

$$\bar{V} = \frac{8(L-x)\gamma R^2\beta^2}{r^2(R^2-r^2)} \qquad (f)$$

With the aid of Equation (f), acceleration and friction forces at any position (X) in the cylinder casting may be determined.

$$\bar{\alpha} = \frac{d\bar{V}}{d\theta} = \bar{V}\left(\frac{a\bar{V}}{ax}\right)_r + \left(\frac{a\bar{V}}{ar}\right)_x \frac{dr}{d\theta} \qquad (g)$$

$\bar{\alpha}$ = acceleration of a liquid particle.

Combining Equations (c), (d), (f), and (g):

$$\bar{\alpha} = \frac{64\gamma R^4(L-x)}{r^4(R^2-r^2)^2}\left[\frac{R^2-2r^2}{R^2-r^2} - \gamma\right] \qquad (h)$$

The pressure gradient along the length of the cylinder casting due to acceleration and friction, from Newton's first law is as follows:

$$\Sigma F = \bar{M}\bar{a}, \text{ or}$$

$$-T \cdot 2\pi r dx - \pi r^2 dP = \frac{\rho_e}{g} \cdot \pi r^2 dx \cdot \bar{\alpha} \qquad \text{(i)}$$

where g = acceleration due to gravity

ρ_e = weight density of liquid metal

P = pressure

T = friction force per unit area

$$T = \frac{\rho_e}{2g} \cdot \bar{V} \cdot f$$

f = friction factor (numerical evaluation obtained from chart of friction factor versus Reynolds' number).

In addition to pressure drop from acceleration and friction terms, the velocity component adds a small pressure drop.

$$\Delta P_v = \frac{\rho_e}{2g} \cdot V_o^2$$

$$\Delta P_v = \frac{\rho_e}{g} \cdot \frac{32L^2\gamma^2\beta^4R^4}{r^4(R^2-r^2)^2} \qquad \text{(j)}$$

Equation (h) added to Equation (i) and integrating between limits $X=0$, and $X=L$ gives:

$$\Delta P_{\text{TOTAL}} = \frac{\rho_e}{g} \cdot \frac{64\gamma\beta^4R^4L^2}{r^4(R^2-r^2)^2}\left[\frac{R^2-2r^2}{2(R^2-r^2)} + \frac{\gamma fL}{3r}\right] \qquad \text{(k)}$$

Numerical evaluation shows $4 \ll R$, which simplifies Equation (k) to:

$$\Delta P_{\text{TOTAL}} = \frac{\rho_e}{g} \cdot \frac{64\gamma\beta^4L^2}{r^4}\left(\frac{1}{2} + \frac{\gamma fL}{3r}\right) \qquad \text{(l)}$$

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QUANTITATIVE EVALUATION OF THE SUSCEPTIBILITY OF VARIOUS ALLOYS TO SHRINKAGE DEFECTS

By

Richard A. Flinn*

Probably the greatest source of scrap in the production of quality castings is the occurrence of "shrinkage" defects resulting from the increase of metal density in passing from the liquid to the solid state. Quantitative measurement of the susceptibility of different metals to this defect is of great commercial importance as well as of theoretical interest as a physical phenomenon. In the discussion of this problem it is best to review the widespread incidence of the shrinkage defects first and then proceed to the underlying causes.

Definitions, Review of the Literature

If an adequate source of liquid metal is not closely available to furnish additional metal to a freezing region of a casting, one of the following principal defects will be produced:

1. Open centerline shrinkage or riser shrinkage
2. Flocculent, centerline shrinkage
3. Interdendritic shrinkage
4. Surface shrinkage ("sinks", "dishing").

All four of these defects are clearly attributable to change in density. The surface contours of the casting freeze and trap a volume of liquid metal which is fated to yield a smaller volume of solid metal. If the liquid freezes with a smooth advancing plane front toward the thermal center of the casting, an open shrinkage spot will be produced. If a good deal of dendrite growth precedes the advancing general front, flocculent shrinkage will be obtained. If crystallization is predominantly random and little related to position, interdendritic microshrinkage will occur. Finally, if the initial cope surface whch is produced upon freezing is weak it will collapse toward the shrinkage area because of atmospheric pressure and gravitational effects.

It should perhaps be pointed out at this time that it will be demonstrated, contrary to the general belief, that all four of these defects can be produced in one metal although in practice the defects are more prominent in one or two of the categories for a given metal.

In addition to the preceding principal defects, liquid to solid shrinkage can be an important secondary factor in producing other defects, notably hot tears and microporosity.

One of the principal sources of hot tears is the strain resulting from contraction of already solidified metal dendrites just above and at the solidus. Ackerlind, Bishop and Pellini[1] have demonstrated that unless liquid metal is available to fill in the interdendritic ruptures, hot tearing tendency is exaggerated. In a recent national survey of hot tear castings shrinkage was considered a contributory factor in 15 per cent of the cases.[2]

Microporosity is the term generally used to describe the presence of scattered amounts of gas porosity. When liquid pools of metal are trapped within a casting without access to a riser, the voids that are formed have only the vapor pressure of the cast metal within them, not atmospheric pressure. This is usually a very low value and promotes ebullition of any dissolved gases in the metal. The total void volume is therefore the sum of the shrinkage volume and evolved gas volume. As a corollary it should be pointed out that with very strong directional solidification both of these defects may be eliminated.

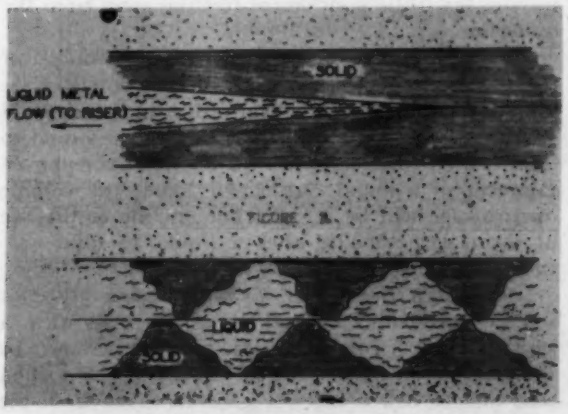


Fig. 1 — Feeding of castings during solidification.

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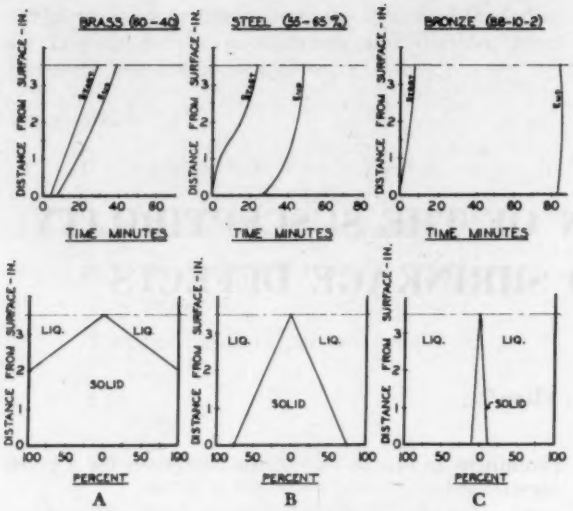


Fig. 2 — Freezing of sand-cast alloys[3].

Quantitative Treatment of Shrinkage

To avoid shrinkage it has long been recognized that two conditions must be present:

1. An adequate supply of liquid metal must be available at the time needed. This is obtained by having an adequately sized riser with a sufficiently slow cooling rate and is discussed in great detail elsewhere.

2. A satisfactory feeding channel must remain open to the regions demanding liquid metal. This is the most difficult portion of the feeding problem and empirical data are available only for a few analyses. The nature of the problem is illustrated by Fig. 1. When a smooth freezing front is obtained, it presents little impedance to the flow of liquid metal (Fig. 1a) and only a slight thermal gradient is sufficient to insure satisfactory feeding. By contrast, a jagged dendritic front (Fig. 1b) in which metal is actually freezing at the centerline before other regions of the same section are completely frozen, presents severe impedance to liquid metal flow. In addition, to obtain a sound casting in this case, the metal must penetrate not only along the centerline but also sideways to reach the positions of freezing away from the centerline.

Feeding distance, therefore, becomes a problem of the impedance to liquid flow at the centerline. It should be noted parenthetically that the *amount* of feed metal needed, a simple function of the liquid to solid density difference, has little effect upon the attainment of a sound casting. Of course the volume of shrinkage will be affected but not the occurrence of shrinkage, provided that a liquid to solid *contraction* is involved.

It has been the custom to rank certain alloys as severely dendritic and others as freezing with a smooth front. A comparison of this type is shown by the freezing wave curves of 60-40 brass, 0.6 carbon steel and 88-10-2 bronze, (Figure 2a, taken from the work of Brandt, Bishop and Pellini[3]).

If the structure within these sand castings is examined at the time crystals begin to form at the centerline in each case, the situation shown in Fig. 1b will exist. In the 60-40 brass the casting will be solid to within one-half inch of the centerline, whereas in both the steel and 88-10-2 some liquid will still remain at the *surface.**

If the wide variety of alloys studied by the NRL group[3] is surveyed in this manner, the data of Table 1 will be obtained.

TABLE 1

Material	Thickness of Casting Completely Solid When Freezing Begins at Center. (Total Distance, Surface to Centerline = 3.5 in.)	Centerline Resistance Factor (Center Line Freezing Time/ Casting Freezing Time) × 100
Copper (99.8)	3.2	<1
Lead (99.)	2.5	17
60-40 Brass	2.2	26
12% Cr Steel	0.5	38
18-8 Steel (0.2% C)	0.5	35
0.6% C Cast Steel	0 + 10% liquid at surface	54
Monel	0 + 50% liquid at surface	64
Al_4Cu	0 + 90% liquid at surface	96
88-10-2	0 + 90% liquid at surface	95
Al_8Mg	0 + 90% liquid at surface	91

As another measure of the tendency to block off the feeding channel, a "centerline resistance factor" has been calculated. It is reasoned that the degree of blocking for any situation will depend on the relative amount of time that dendrite growth takes place at the centerline or (centerline freezing time/total freezing time) x 100.

Thus, in the case of the 60-40 brass centerline freezing is ten seconds while the entire casting takes 38 seconds. Thus, for 26 per cent of the freezing time, dendrite growth was occurring at the centerline. In steel for 64 per cent of the time, centerline dendrite growth prevailed while for the 88-10-2 bronze the figure was 95 per cent. There is little doubt that a markedly steeper thermal gradient throughout the latter casting is required for adequate feeding as indicated by the centerline resistance factor and the illustrations of Fig. 2.

A further application of the resistance factor as a quantitative basis for comparison is illustrated in Table 2 and Fig. 3 in dealing with the same alloys but in chilled molds. The freezing curves, Fig. 3, indicate that the feeding conditions should be much improved. Not only is the 60-40 brass almost completely frozen (3.2 of 3.5 in.) before centerline crys-

*For the purposes of illustration the amount of liquid remaining at the surface has been taken as proportional to the percent of time required to complete solidification. The derivation of the exact amount depends upon the phase diagram, the correlation of freezing time with percentage of precipitate and undercooling effects.

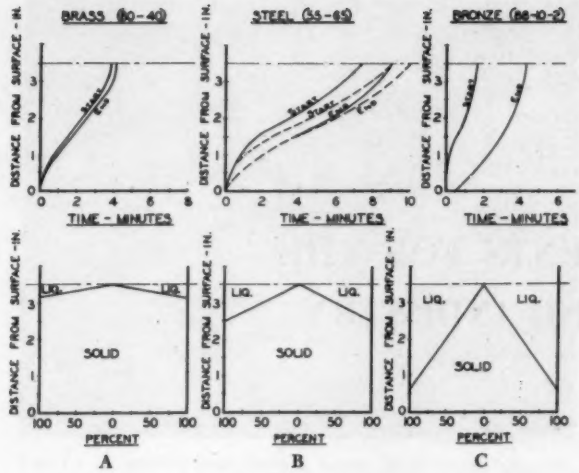


Fig. 3 — Freezing of chilled alloys[3].

TABLE 2

Material	Thickness of Casting Completely Solid When Freezing Begins at Center. (Total Distance, Surface to Centerline = 3.5 in.)	Centerline Resistance Factor (Center Line Freezing Time/ Casting Freezing Time) × 100
Copper (99.8)	3.2	3.1
60-40 Brass	3.2	2.5
12% Cr Steel	3.0	9.3
18-8 (0.2% C) Steel	2.7	9.4
Lead 99%	2.6	17
0.6 C Steel	2.5	25
Monel	2.2	19
88-10-2 Bronze	0.6	63
Al 4.5 Cu	0 + 25% liquid at surface[1]	46[2]
Al 8 Mg	0 + 50% liquid at surface	65

[1] See text for explanation

[2] Shows anomalous freezing. See (3). Derived from Brandt, Bishop and Pellini (3).

tals appear, but the 0.6 carbon steel approximates the condition of the sand cast 60-40 brass and the feeding of the 88-10-2 resembles the sandcast 0.6 carbon steel.

A summary of the resistance factors, Table 2, for these chilled molds shows the same trend. The factor for the 0.60 carbon steel is now of the same order as the sand-cast 60-40 brass and the 88-10-2 approximates that of the sand-cast 0.60 C steel. The relation of resistance to physical properties is summarized in Table 2.

Since the centerline resistance factor appears to be related to feeding distance, it should be possible to derive a curve of the type shown diagrammatically in Fig. 4. From data of this type it might be possible to ascertain the feeding distance for a new alloy from a knowledge of its resistance factor in the desired mold material. This is far simpler to obtain than feeding distance data.

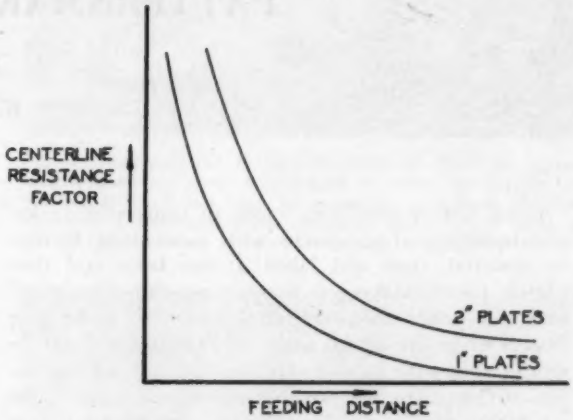


Fig. 4 — Correlation of feeding distance with resistance factor (schematic).

Conclusions

1. The centerline resistance factor (centerline freezing time/total freezing time) provides a quantitative index of the susceptibility to shrinkage (feeding difficulty) for a given alloy to mold material combination.

2. The resistance factor indicates the ranking of metal compositions and mold effects in accord with general experience in feeding.

3. When feeding distance data become available for a variety of alloy-mold material combinations, it may be possible to develop a resistance factor-feeding distance relationship.

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NEW DEVELOPMENTS FOR THE PATTERNMAKING INDUSTRY

By

M. K. Young*

Great strides have been made in using plastics for matchplates and coreboxes with substantial savings in material, time and labor. It has been said that plastic patternmaking is making possible new products that might formerly have died on the drawing boards or in the design stage. So notwithstanding the stories about the plastic corebox that did not do the job or the plastic pattern that failed to produce 2,000 or 3,000 metal castings, plastic patternmaking is a tried and proven method. Plastic patternmaking methods and techniques appear to have merit in reducing cost.

Fabrication of plastic matchplates and coreboxes presents many of the problems that are experienced in other industries where casting and molding materials are used. In the foundry, the metal casting is only as accurate as the sand mold in which it was cast. By the same token, plastic matchplates, die patterns, etc., are only as accurate as the molds in which the resin is poured or laminated (see Figs. 1-12). Therefore, it is extremely important that the proper mold-making materials and techniques be applied. It is equally important that the proper plastic materials be used.

During the early stages of plastic patternmaking, the phenolic plastics were used. While these materials helped to bring about this new concept in patternmaking, the materials did leave something to be desired. In the course of aging, it was found that the phenolic plastics were not dimensionally stable. In addition to after-shrinkage, they were found to crack or craze in service. As time went on, the phenolic plastics were improved.

Then other materials, such as the polyester plastics, were evaluated and found to have merit as a patternmaking material. However, due to the high initial shrinkage, close tolerances could not be held without allowance for this shrinkage. Notwithstanding the high shrinkage of the polyester it is still being used for some applications.

In 1950, the epoxy resins were introduced as a patternmaking material. The epoxies are ideally suited for plastic patternmaking since their low shrinkage provides very accurate reproduction of master patterns or models. They also have good resistance to heat, high compressive strength, good adhesion, and they cure or harden very easily at room temperature. Plastic patterns can be made with these resins by casting or laminating (see Figs. 1-12).

The epoxy resins are used in conjunction with fillers such as iron oxide, aluminum oxide and powdered aluminum. These fillers provide additional hardness and strength for patterns and coreboxes. In addition, the greater resiliency of the plastic surface provides much better resistance to wear. Epoxy plastic coreboxes have been reported to outlast aluminum coreboxes in sand blowing operations. When epoxy plastic patterns or coreboxes become cracked or worn, repair is accomplished quickly by the addition of patching resin made from the same material as the corebox or pattern.

Plastic resins have found extensive use in fabrication of plastic stamping dies, jigs and fixtures in the automotive and aircraft industries. These fields have opened new opportunities for the patternmaker who can handle plastics.

Materials Suitable for Short Runs

These new materials are ideally suited for short-run production patterns which would have been prohibitively expensive if conventional wood patterns were used. Even in extended production runs, some epoxy resin patterns have proven that they can do the job. Although this new concept for patternmaking is in the development stage, the use of epoxy resins for plastic patterns offers one of the finest opportunities the patternmaker has had in many years.

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Fig. 1 — A surface coat resin is mixed with proper quantity of hardener and applied with paint brush. It is flowed on freely and never brushed out unless fine detail is being reproduced. First brush coat is applied approximately 1/32 in. in thickness.



Fig. 2 — Another mix is used to build up 1/8-in. to 3/16-in. surface coat thickness. As required to make the pattern or corebox, a third mix may be necessary to achieve desired thickness. After surface coat becomes tacky (45-50 min) a thin coating of laminating resin is applied.



Fig. 3 — Precut glass cloth is pressed into laminating resin. Wrinkles are smoothed out with back of patternmaker's hand.



Fig. 4 — Cloth is pulled and worked into design of the pattern, model or mold to avoid wrinkles and resin pockets.

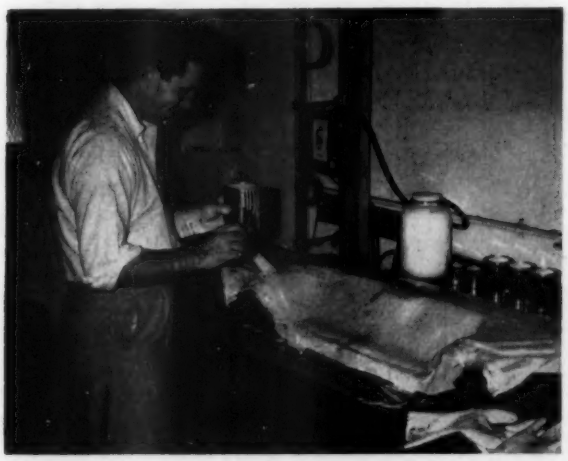
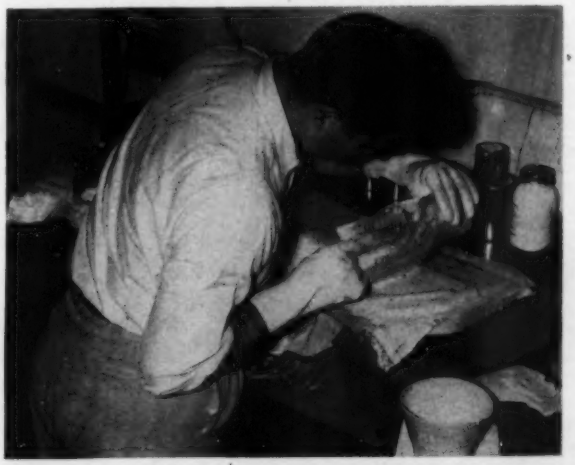


Fig. 5 (Above) — A coat of laminating mix is applied for each cloth lamination. Fig. 6 (Right) — Then another sheet of cloth is applied. Cloth is cut oversize so laminate will



cover entire pattern. This excess material can be removed with a knife or bandsaw after the laminating operation.



Fig. 7 (Above) — Glass tape 3 in. in width can also be used to make lay-up. Joints should overlap. Fig. 8 (Right) — Metal tubing is fastened to the laminate with putty mix. Tubing provides rigidity and facilitates handling of check fixture. Putty mix is built up with spatula to desired thickness and trowled in corners to form fillets which provide greater strength.





Fig. 9 — Where intricate shapes are to be reproduced the surface coat resin is brushed into the small cavities and fine detail of the mold surface.



Fig. 10 — After the surface coat has cured to touch dry or tacky stage, laminating resin is applied, followed by pre-cut glass cloth as shown here.



Fig. 11 — Expanded metal is then imbedded into laminating mix before it hardens. This provides additional strength and serves as a mechanical bond, for plaster cement backup.



Fig. 12 — A flask or frame is positioned around the mold. Gypsum cement is poured as backup material where light weight is not a factor.

ANALYSIS OF FACTORS AFFECTING SURFACE FINISH OF GRAY IRON CASTINGS

By

R. L. Yard* and D. C. Ekey**

Introduction

Casting finish is vital to the foundry industry. Its importance can be attributed to the fact that functional requirements or machining specifications of a part may require a smooth as-cast surface. Then too, castings are often prominently displayed as components of end products (or are end products in themselves) necessitating a good finish for appearance sake.

The processes of die casting, investment molding and shell molding provide the attainment of a smoother surface as an inherent advantage over the typical green sand molding process. While such improved surface characteristics tend to lower machining costs, the capital involved in these processes frequently necessitates relatively high production rates to warrant their use. Despite these newer developments the largest portion of all castings is produced in green-sand molds.

Economy, versatility and rapid production account for the wide use of green-sand molding processes. Therefore, it would be advantageous to improve the surface finish of green sand castings. Such improvement would provide a wider application of green-sand molding instead of the more costly methods now in use.

Ideas have been advanced for establishing standards for as-cast surfaces,[1,3] but the investigation of the factors affecting surface finish have been meager. Fairfield and MacConachie[7] found that pouring temperature, moisture, sand preparation and the amount of ramming of the sand in the flask affect the resulting surface finish. In their study they used, as an index of casting finish, the loss in weight by buffing the test casting to a smooth finish.

A study by Ekey and Goldress,[5] employing a statistical analysis technique, indicates that sand fineness and metal pressure have a significant effect on surface roughness, while there is no reason to suspect that wood flour (a sand additive) has any effect on surface roughness. Neither of the aforementioned investigations showed the effect of a combination of two or more variables on surface finish (interaction of variables).

The object of the experiment was to determine what interactions of the variables (sand fineness, mold hardness and static metal pressure) might be significant in their relation to surface finish.

With this objective in mind it was decided to use a statistical inference technique known as the analysis of variance. This technique employs randomized blocks and enables one to test the significance of the interaction of variables.

The variables used in studying the surface finish of gray iron castings were sand fineness, static metal pressure and mold hardness. The sand fineness levels are gradations from coarse to fine sand. The metal pressure and mold hardness levels tend to be extreme values for these variables under the imposed conditions. The variables, shown in Table 1, were chosen after careful consideration of previous investigations and theoretical assumptions based on experience and study in foundry technology. Cast iron was chosen since it is the major cast alloy, comprising approximately 80 per cent (by weight) of all casting production.

TABLE 1 — VALUES OF EACH VARIABLE

Sand Fineness		
Level	Type of Sand	AFS Fineness Number
I	New Jersey Silica 2-14	33
II	New Jersey Silica 60-1	62
III	New Jersey Silica 100-1	96
IV	New Jersey Silica 140-1	124

Metal Level	Pressure (measured from parting line), In.	Mold Level	Hardness Reading
1	4	A	50
2	16	B	85

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The physical model chosen to fulfill the object of the experiment consisted of two pairs of cast rectangular blocks, 13 in. x 4 in. x 3 in. Eight green-sand test specimens, prepared by ramming sands of varying grain fineness to two degrees of hardness and placed in core prints of baked sand molds, formed part of the sides of each block (Fig. 4). These sand specimens contained the measured variables except for that of metal pressure. Metal pressure was varied by the additon of a down sprue extension to one of the pairs of molds for each replication. Thus, close control of these variables was possible. Surfaces of the cast blocks, formed by the end of the sand cores, were then measured with respect to surface roughness. Statistical analysis (of the roughness indices obtained) indicated whether the interactions of the variables, or whether the main factors (or both), were significant with respect to the casting surface finish of gray iron.

It was felt that the benefits of this research would be of maximum assistance to the producers of gray iron castings.

Equipment

The equipment necessary for the experiment had to be designed and fabricated prior to conducting it. An aluminum matchplate was made of the 13 in. x 4 in. x 3 in. block with core prints to accommodate eight standard 2 in. x 2 in. diameter sand test specimens (Fig. 1). The matchplate insured uniform location of the component parts of the mold. To withstand the metal pressure and to facilitate handling, the molds were made of core sand, (No. 140-vented) hand rammed, baked for 4 hr at 450 F.

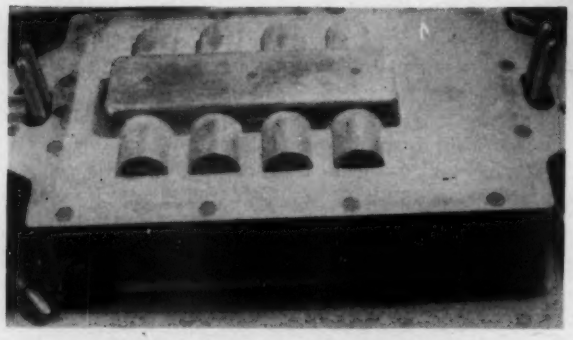


Fig. 1 — Photo of matchplate.

The down sprue extensions (Fig. 5) were made by ramming core sand around a 1-in. dowel rod in a piece of 12 in. x 5 in. diameter stovepipe. After drawing the dowel rod these extensions were baked in a core oven.

This physical model was designed to meet the requirements of the mathematical model (i.e., replication, randomization, pressure differential and placement of units of material).

Preparation of Test Specimens

All sand used in the specimens was obtained from foundry suppliers. Each of the four types of washed sand was combined with 4 per cent western bentonite. The resultant mix contained 5½ per cent moisture. Processing was done in 25-lb batches with a laboratory-type muller using a three minute mixing cycle. The resulting mixture was then stored in sealed jars to prevent loss of moisture from the sands preparatory to their use. The mechanical properties of green permeability and green compressive strength for each of the mixtures were determined in the laboratory and are presented in Table 2.

TABLE 2 — PROPERTIES OF THE SAND MIXTURES

Sand	Green Permeability No.	Green Compressive Strength
2-14	312.50	3.8
60-1	116.00	3.4
100-1	63.25	3.4
140-1	38.25	4.1



Fig. 2 — Completed green-sand specimen.



(Right)
Fig. 3 — Photo of specimen rammer.

Standard equipment was employed to make the sand test specimens (Fig. 3). Trial and error methods yielded the proper weight of a given sand to obtain a 2-in. specimen at a given mold hardness. (Fig. 2). The hardness was controlled by the number of rams on the sand specimen. The weights of sand needed for each specimen was determined prior to the day on which the experiment was performed. This procedure made the rapid and accurate preparation of the specimens possible. As the specimens were stripped from the specimen tube, they were placed, bottom up, on a board designed to hold enough specimens for one replication of the test.[16]

Sixteen sand test specimens (one replication) were randomly placed, eight per mold, in two test molds. The specimens were placed in the core prints

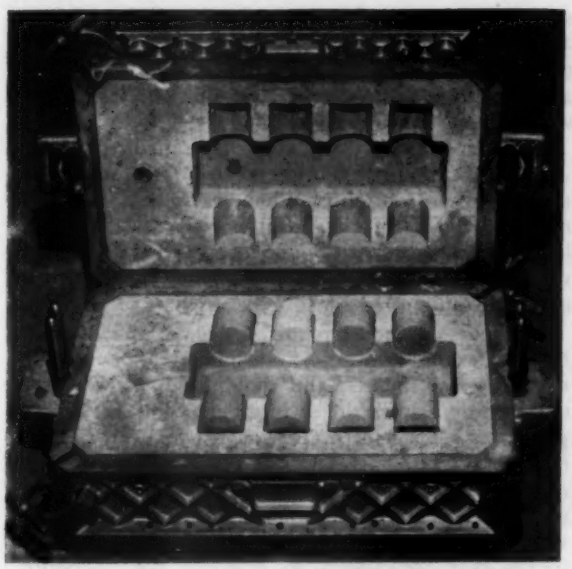


Fig. 4 — Open mold with green sand test specimens in position.

with the bottom end (i.e., the end away from the rammer) forming the test surface (Fig. 4). A talc sealer was supplied to the parting line, molds were closed, flasks removed, and slip jackets and weights positioned. A 12 in. down sprue extension was added to one mold of each replication. A mixture of fire clay, heap sand and water was used at the base of the extension as a seal (Fig. 5).

Lip pour ladles of 100-lb iron capacity were utilized in pouring the molds from a 2,000-lb cupola heat. The metal was allowed to cool in the ladles until an optical pyrometer reading indicated the pouring temperature to be between 2680-2690 F. Analysis of the gray iron alloy used showed 3 per cent carbon, 0.10 per cent sulphur and 2 per cent silicon.

On the day subsequent to pouring, the molds were shaken out and the gates removed. The resulting 34-lb castings were then cleaned with the aid of a laboratory type sand blast machine.

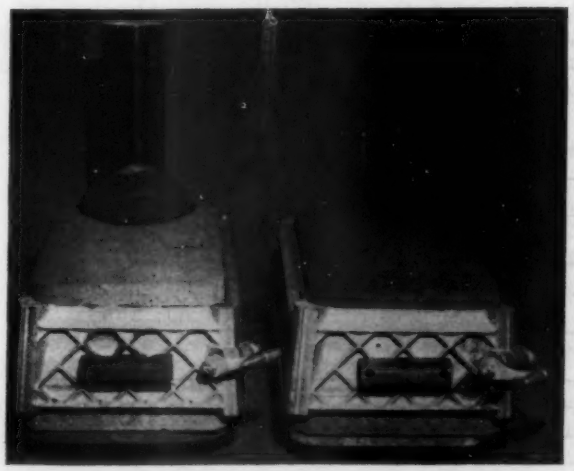


Fig. 5 — Photo of completed molds.

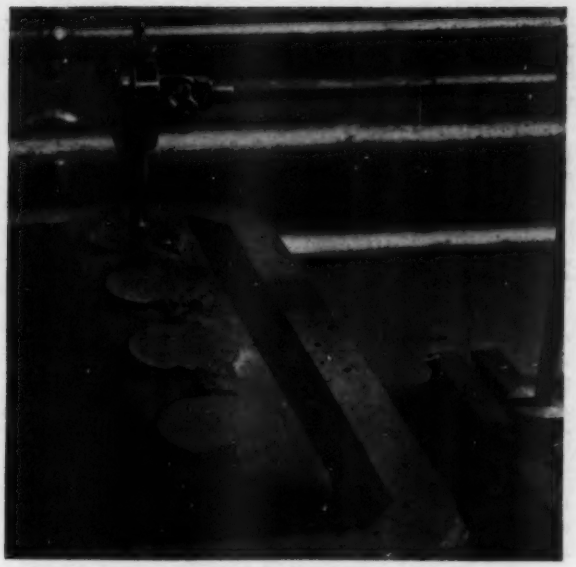


Fig. 6 — Roughness measurement set-up.

Measurement of Surface Roughness

Basically, surface roughness is the variation of the actual surface from a hypothetical mean surface line. This variation was recorded by means of a dial indicator. The castings were placed on edge in a sand bed so that the test surface was parallel to a surface plate (Fig. 6). Care was taken to insure that the tip of the indicator was perpendicular to the test surface. Four readings, chosen at random within a ¼-in. diameter circle, constituted a subgroup. Twenty random subgroups comprised a sample. Data for each sample are presented in Appendix A.

A good index of surface roughness is the average of the ranges for each subgroup provided the ranges are in statistical control. Therefore, statistical control limits were determined and out-of-control points eliminated before calculating the roughness indices. By this procedure it was possible to eliminate the out-of-control values and recalculate new control limits until all of the remaining values were in control.[11] The choice of 20 subgroups of four was based entirely on statistical theory.[11]

Penetration was encountered under two combinations of treatment conditions which utilized a 2-14 sand and the higher metal pressure (see Fig. 7 and

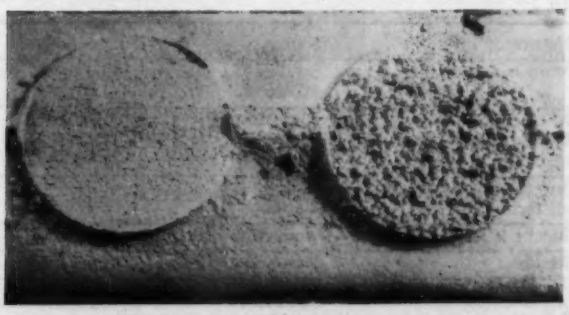


Fig. 7 — Comparative roughness of two of the surfaces.

TABLE 3 — ROUGHNESS INDICES

Sand Fineness		2-14		60		100		140	
Mold Hardness		50	85	50	85	50	85	50	85
Pressure 4 in.									
		8.570	6.563	6.975	4.893	4.795	3.085	3.905	3.526
		(1)	(2)	(3)	(4)	(5)	(6)	(7)	(8)
		13.325	7.374	6.058	3.765	4.460	3.505	4.605	4.195
		(9)	(10)	(11)	(12)	(13)	(14)	(15)	(16)
Pressure 16 in.									
		*157.812	*50.000	9.575	6.065	5.485	6.150	8.635	5.090
		(17)	(18)	(19)	(20)	(21)	(22)	(23)	(24)
		*189.062	*54.688	14.044	6.815	5.600	6.070	6.820	4.845
		(25)	(26)	(27)	(28)	(29)	(30)	(31)	(32)

* Denotes a penetrated condition.
(1) (2) (- -) Denotes sample number.

Table 3). It is generally recognized that penetration results in an adhering mass of sand and metal on the casting.[16] This being the case, dial indicator methods would not be measuring the true metal surface but the surface of the sand and metal mass clinging to the casting surface. Since the casting is of little value to the foundryman unless the penetration is removed, it was decided to use the average depth of penetration as a measure of roughness. It was felt that this measure would be comparable to the average range used for non-penetrated surfaces. To measure the depth of penetration, the blocks were cut through the center of the penetrated surface with a power hacksaw. The depth of penetration could then be determined directly from the cross section.[10] Table 3 contains the roughness indices obtained in the experiment.

Mathematical Model

Statistical methods permit the study of several factors simultaneously with a minimum expenditure of time, effort and money. These multivariate designs offer two advantages over experiments involving only single factors. Firstly, greater efficiency results because these factors are evaluated with a smaller number of observations than would otherwise be necessary,[12] and with the same degree of precision. Secondly, more comprehensive analysis of the data is possible in that not only main effects, but also their interactions can be evaluated. Interaction means that the variables are producing effects together which would not be produced separately. Also, any conclusion has a wider inductive basis when inferred from an experiment in which the quantities of other ingredients have been varied, than it would have in any amount of experimentation in which these ingredients would have been kept constant.[8]

The experiment was designed to utilize the advantage of a statistical inference technique known as the analysis of variance. When, as in this case, one of the variables of classification is essentially a replication, i.e., a repetition of the experiment under various environmental conditions, the experiment is called a randomized block experiment.

The physical model required the use of two molds for each complete replication. The treatment conditions of sand fineness and mold hardness were randomly assigned within each mold. Thus the appropriate error term for evaluation of conditions involving these treatments was the residual variability as represented in Table 4. Because two molds were necessary for each replication, differences among molds were confounded with differences due to height. This means that the ability to evaluate the "pressure" main effect was lost. Since previous research with aluminum has shown pressure to be significant with respect to casting surface finish,[5] the increased precision gained by confounding warrants the loss of this information. Confounding is a solution to the problem of heterogeneity of materials.[8]

Analysis of the Data

The F test, which is employed in the analysis of variance for testing the significance of departures from equality of means, was used to evaluate the results. The ratio of the interaction variance or main effect variance to the error variance is distributed as the F statistic.[3] Thus tests of significance are possible.

Duncan[4] points out that if an interaction is not significant, it is then an estimate of the residual error and is combined with the stated residual error forming an improved estimate of the residual variance. It is necessary first to test the lowest value of the highest order interactions followed by each higher value in the same order interactions. Successively lower order interactions and main effects are treated in the same manner using the best estimate of the error term in each case. If an interaction is significant, it is not valid to test a lower order interaction, or a main effect, related to the significant interaction against the best estimate of the error term. However,

TABLE 4 — ANALYSIS OF VARIANCE

Nature of Effect	Source	Sum of Squares	D.F.	F ratio*	$F_{.95}$	Significant
Main factors	Sand	18,265.482	3	1.186	9.28	No
	Hardness	2,330.686	1	1.952	4.32	No
Interactions between pairs of factors	S × H	5,524.634	3	1.695	3.16	No
	S × P	15,400.362	3	4.298	3.07	Yes
	H × P	1,822.918	1	1.748	4.45	No
Interaction of all factors	S × H × P	5,045.583	3	1.856	3.34	No
Blocks	Among Blocks	6,345.672	3			
Residual	Residual	12,685.309	14			
Total		67,421.646	31			

* F ratio calculated using best estimate of error term.

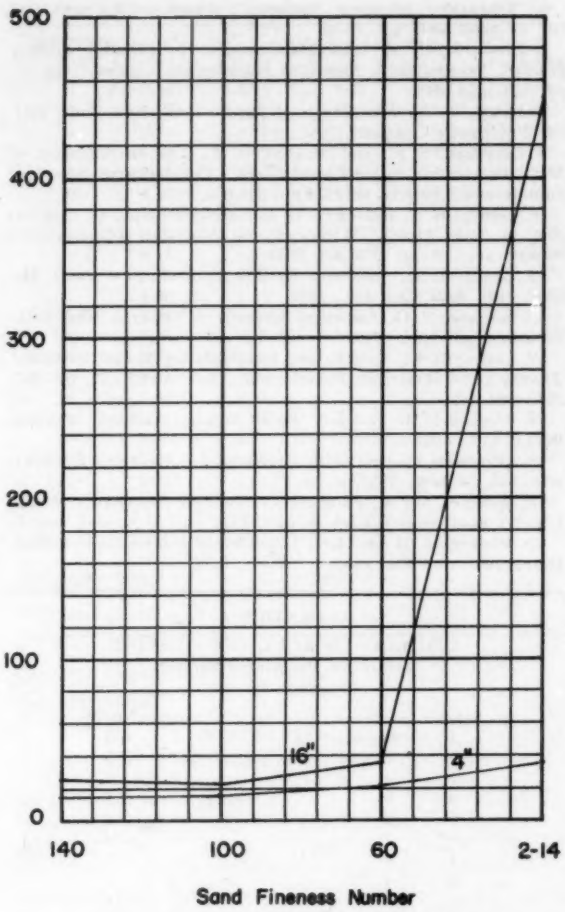


Fig. 8 — Graphical representation of the significant interactions between sand fineness and metal pressure.

the related lower order interaction or main effect can be compared to the significant interaction.

Similar discussions of this procedure may be found in Dixon and Massey,[3] Brownlee[2] and Moroney.[14]

It can be seen from Table 4 that the second order interaction of all the factors was not significant but the first order interaction between sand fineness and metal pressure was significant, when one employed the foregoing procedures, at the 5 per cent level of significance. (The 5 per cent level is considered appropriate for experimentation of this type).

In general, several factors may cause an interaction to show significance when tested:

1. Although there was no interaction, we obtained a value which we declare significant. This situation of non-interaction will occur 5 per cent of the time on the average if we use a 5 per cent level of significance;
2. The experimental materials of the subgroups were not randomly drawn;
3. Another uncontrolled and unrecognized factor was of sufficient importance to be included in the experiment; and
4. The interaction was in fact significant.

The first factor was included as a basic criterion in the mathematical design of the experiment. The method of evaluation is acceptable, unless equivocal results, based on experience and study in foundry technology, are believed to exist. The second possibility was avoided by the design of the experiment. From a technological point of view the third factor was rejected and attention was directed to the fourth.

The interaction between sand fineness and metal pressure was accepted as significant in this experiment.

Interpretation of the Results

The graphical representation of the interaction between sand fineness and metal pressure[14] is illustrated in Fig. 8. One will note that the finer sands are not subject to roughness in extremes of pressure, but the coarser sands do produce rougher finishes in conjunction with increasing metal pressure.

Examination of cold shuts, penetrations, veinings, or mis-runs gives evidence of the high surface tension of molten metal (1800 to 2000 dynes per centimeter for steel as compared to 81 dynes per centimeter for water at 20 C.). The direct effect of the true surface tension of many metals (gray iron probably included) is almost overshadowed by the effect of surface films, the influence of which is equivalent to that of a greatly increased surface tension[10].

At low metal pressures this condition could be predominant. Preventing the metal from being forced between the sand grains to any great extent, such surface tensions might form a relatively smooth surface in the presence of coarse sand. Rounded projections on the metal surface also tend to support this theory. With a sufficient increase in pressure this apparent surface tension could be overcome and the metal might be forced between the sand grains[10]. The size of the sand grains then becomes important to the resulting surface finish.

The findings of Gertsman and Murton on core sand[9] support the findings of this investigation: that is, hardness (within the range considered) has little effect on roughness and penetration.

Conclusions

Based upon the statistical analysis of the experiment, using a 5 per cent level of significance, the following conclusions were reached for the range of factors investigated:

1. Sand fineness and metal pressure interact to influence casting surface finish;
2. Increasing metal pressure and decreasing sand fineness increase surface roughness;
3. The mold hardness of sand due to ramming does not significantly influence casting surface finish;
4. Sand fineness, compared with the significant interaction is not a factor contributing to casting surface finish;
5. Sand fineness, mold hardness and metal pressure (second order interaction) do not interact with respect to casting surface finish; and

6. The first order interactions of sand fineness and mold hardness, and mold hardness and metal pressure were not significant with respect to casting surface finish.

Suggestions for Further Study

The experiment showed the interaction of sand fineness and metal pressure to be significant. The next step might be a detailed study of this interaction with more levels of the two variables or other studies not necessarily statistical in nature. The graph of the interaction indicates that the finer sands have a smaller tendency to interact with the metal pressure and suggests that future efforts be concentrated on the range of coarser sands.

Determination of the critical values of the variables for penetration would be of practical use to the foundry industry. Likewise, the problem of stages in the development of a penetrated surface invites another line of investigation. Such inquiry would involve provisions for bleeding the mold subsequent to pouring after various time intervals have elapsed.

Another possibility suggested is a consideration of the relationships between surface tension and metal pressure, a study which might be carried out with the investigation of the filling of very thin sections.

Summary

The object of this experiment was to study the effect of the interactions of the variables (sand fineness, mold hardness and metal pressure) on the surface finish of gray cast iron. This information is of practical value to the foundry industry. To accomplish the objective a statistical inference technique known as the analysis of variance was used. Analysis showed that an interaction of sand fineness and metal pressure had a significant effect on casting surface finish. Remaining interactions and main effects studied were not significant with respect to the dependent variable within the range studied.

Acknowledgments

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APPENDIX A

SUMMARY OF ROUGHNESS VALUES

Values for Penetrated Surfaces

(values in 1/32 in.)

Sample No. 17		Sample No. 25	
3	6	7	6
5	7	7	5
7	4	3	6
5	5	7	7
5	6	8	8
4	4	7	4
4	6	6	7
6	4	6	6
5	5	7	4
6	4	7	3
	101		121

Roughness index = $\frac{(101)\ (1000)}{(20)\ (32)} = 157.812$

Roughness index = $\frac{(121)\ (1000)}{(20)\ (32)} = 189.062$

Sample No. 18		Sample No. 26	
2	2	2	1
2	1	2	2
1	2	1	2
1	1	1	3
2	2	1	1
1	2	2	2
1	2	1	2
2	1	1	1
2	2	3	2
1	2	2	3
	32		35

Roughness index = $\frac{(32)\ (1000)}{(20)\ (32)} = 50.000$

Roughness index = $\frac{(35)\ (1000)}{(20)\ (32)} = 54.688$

See reproduction of Typical Data Sheet — Sample No. 2 on next page.

APPENDIX B

CALCULATIONS

SUMS OF SQUARES CALCULATIONS:

Correction Factor = $\frac{(626.355)^2}{32} = 12{,}260.018$

Hardness Main Effect	
50	85
449.726	176.629

TYPICAL DATA SHEET – SAMPLE NO. 2

Subgroup No.	X	X	X	X	Range
1	6.56	6.18	6.45	6.53	0.38
2	5.89	5.86	5.76	5.85	0.13
3	4.99	5.37	5.38	5.30	0.39
4	4.97	4.83	5.17	4.68	0.49
5	5.87	5.56	6.80	5.78	1.24
6	5.57	4.97	6.41	5.38	1.44
7	7.70	7.43	9.62	7.48	*2.19
8	8.07	7.78	7.63	7.52	0.55
9	8.17	8.20	8.28	8.52	0.35
10	9.35	8.87	8.89	9.18	0.48
11	8.09	7.33	7.14	7.66	0.94
12	7.00	7.01	6.81	6.35	0.66
13	7.16	6.37	6.26	7.09	0.90
14	8.66	8.65	8.57	9.04	0.47
15	8.67	9.09	8.56	9.17	0.61
16	7.99	8.57	7.86	8.38	0.71
17	10.24	10.23	10.46	10.11	0.35
18	8.68	8.74	8.44	7.88	0.86
19	7.24	6.98	7.17	7.32	0.84
20	5.77	5.41	5.21	5.09	0.68
				Total	12.47

Upper control limit = (2.28) (0.733) = 1.671
Lower control limit = (0.00) (0.733) = 0.000

$$\text{Roughness index} = \frac{(12.47)\ (10)}{(19)} = 6.563$$

* Denotes out of control point.

Each entry is the sum of 16 roughness indices.

$$\frac{\text{Sum of (entries)}^2}{16} = 14{,}590.704$$
$$-12{,}260.355 \text{ Correction Factor}$$
$$2{,}330.666$$

Sand Fineness Main Effect

2-14	60	100	140
487.394	58.190	39.150	41.621

Each entry is the sum of 8 roughness indices.

$$\frac{\text{Sum of (entries)}^2}{8} = 30{,}525.500$$
$$-12.260.018 \text{ Correction Factor}$$
$$18{,}265.482$$

Pressure Main Effect

4 in.	16 in.
89.599	536.756

Each entry is the sum of 16 roughness indices.

$$\frac{\text{Sum of (entries)}^2}{16} = 18{,}508.436$$
$$-12{,}260.018 \text{ Correction Factor}$$
$$6{,}248.418$$

Hardness × Pressure Interaction

50		85	
4 in.	16 in.	4 in.	16 in.
52.693	397.033	36.906	139.723

Each entry is the sum of 8 roughness indices.

$$\frac{\text{Sum of (entries)}^2}{8} = 22{,}662.040$$
$$-12{,}260.018 \text{ Correction Factor}$$
$$-\ 2{,}330.686 \text{ Hardness Main Effect}$$
$$-\ 6{,}248.418 \text{ Pressure Main Effect}$$
$$1{,}822.918$$

Sand Fineness × Pressure Interaction

2-14		60		100		140	
4 in.	16 in.	4 in.	16 in.	4 in.	16 in.	4 in.	16 in.
35.832	451.562	21.691	36.499	15.845	23.305	16.231	25.093

Each entry is the sum of 4 roughness indices.

$$\frac{\text{Sum of (entries)}^2}{4} = 52{,}174.280$$
$$-12{,}260.018 \text{ Correction Factor}$$
$$-18{,}265.482 \text{ Sand Fineness Main Effect}$$
$$-\ 6{,}248.418 \text{ Pressure Main Effect}$$
$$15{,}400.362$$

Sand Fineness × Hardness Interaction

2-14		60		100		140	
50	85	50	85	50	85	50	85
368.769	118.625	36.652	21.538	20.340	18.810	23.965	17.656

Each entry is the sum of 4 roughness indices.

$$\frac{\text{Sum of (entries)}^2}{4} = 38{,}380.820$$
$$-12{,}260.018 \text{ Correction Factor}$$
$$-18{,}265.482 \text{ Sand Fineness Main Effect}$$
$$-\ 2{,}330.686 \text{ Hardness Main Effect}$$
$$5{,}524.634$$

Sand Fineness × Hardness × Pressure Interaction

2-14		60		100		140	
50	85	50	85	50	85	50	85
4 in.							
21.895	13.937	13.033	8.658	9.255	6.590	8.510	7.721
16 in.							
346.874	104.688	23.619	12.880	11.085	12.220	15.455	9.935

Each entry is the sum of 2 roughness indices.

$$\frac{\text{Sum of (entries)}^2}{2} = 66{,}897.110$$
$$-12{,}260.018 \text{ Correction Factor}$$
$$-18{,}265.482 \text{ Sand Fineness Main Effect}$$
$$-\ 2{,}330.686 \text{ Hardness Main Effect}$$
$$-\ 6{,}248.418 \text{ Pressure Main Effect}$$
$$-\ 1{,}822.918 \text{ H} \times \text{P Interaction}$$
$$-15{,}400.362 \text{ SF} \times \text{P Interaction}$$
$$-\ 5{,}524.634 \text{ SF} \times \text{H Interaction}$$
$$5{,}045.583$$

Total and Among Blocks Sums of Squares

2-14		60		100		140	
50	85	50	85	50	85	50	85
4 in.							
8.570	6.563	6.975	4.893	4.795	3.085	3.905	3.526
13.325	7.374	6.058	3.765	4.460	3.505	4.605	4.195
16 in.							
157.812	50.000	9.575	6.065	5.485	6.105	8.635	5.090
189.062	54.688	14.044	6.815	5.600	6.070	6.820	4.845

Each entry is a roughness index.

Sum of (entries)^2 = 67,421.646

$$\frac{\text{Sum of (sum of blocks)}^2}{4} = 18{,}605.690$$
$$-12{,}260.018 \text{ Correction Factor}$$
$$6{,}345.672$$

F RATIO CALCULATIONS:

For the second order interaction (SF × H × P):

$$\frac{12{,}685.309}{14 \text{ D.F.}} = 906.094 \qquad \frac{5{,}045.583}{3 \text{ D.F.}} = 1{,}681.861$$

$$F = \frac{1{,}681.861}{906.094} = 1.856$$

Since the critical value of $F_{.95}$ is 3.34, the interaction is not significant.

Pool the non-significant interaction with the error term:

$$12{,}685.309 + 5{,}045.583 = 17{,}730.892$$

$$\frac{17{,}730.892}{17 \text{ D.F.}} = 1{,}042.994$$

For the first order interaction (H × P):

$$\frac{1{,}822.918}{1 \text{ D.F.}} = 1{,}822.918 \qquad F = \frac{1{,}822.918}{1{,}042.994} = 1.748$$

Since the critical value of $F_{.95}$ is 4.45, the interaction is not significant.

Pool the non-significant interaction with the error term:

$$\begin{array}{r} 17{,}730.892 \\ 1{,}822.918 \\ \hline 19{,}553.800 \end{array} \qquad \frac{19{,}553.800}{18 \text{ D.F.}} = 1{,}086.322$$

For the first order interaction (SF × H):

$$\frac{5{,}524.634}{3 \text{ D.F.}} = 1{,}841.545 \qquad F = \frac{1{,}841.545}{1{,}086.322} = 1.695$$

Since the critical value of $F_{.95}$ is 3.16, the interaction is not significant.

Pool the non-significant interaction with the error term:

$$\begin{array}{r} 19{,}553.800 \\ 5{,}524.634 \\ \hline 25{,}078.434 \end{array} \qquad \frac{25{,}078.434}{21 \text{ D.F.}} = 1{,}194.211$$

For the first order interaction (SF × P):

$$\frac{15{,}400.362}{3 \text{ D.F.}} = 5{,}133.454 \qquad F = \frac{5{,}133.454}{1{,}194.211} = 4.298$$

Since the critical value of $F_{.95}$ is 3.07, the interaction is significant.

For the Sand Main Effect:

$$\frac{18{,}265.482}{3 \text{ D.F.}} = 6{,}088.494 \qquad F = \frac{6{,}088.494}{5{,}133.454} = 1.186$$

Since the critical value of $F_{.95}$ is 9.28, the Sand Fineness Main Effect is not significant.

For the Hardness Main Effect:

$$\frac{2{,}330.685}{1 \text{ D.F.}} = 2{,}330.686 \qquad F = \frac{2{,}330.686}{1{,}194.211} = 1.952$$

Since the critical value of $F_{.95}$ is 4.32, the Hardness Main Effect is not significant.

STATISTICAL TECHNIQUES FOR CLASSIFYING FOUNDRY SANDS

By

J. M. Leaman* and D. C. Ekey**

Introduction to the Problem

The foundry industry must control size and distribution of the sand grains it uses in mold and core making. This control, or lack of control, affects both casting and mold characteristics. Casting characteristics associated with sand properties are surface finish and casting defects such as blows, buckles, burnt-on sand, cuts or washes, drops, surface cracks, penetration, pinholes, rat-tails, and scabs. Sand distribution and fineness is considered a major factor in mold and core properties such as strength, permeability, and refractoriness.

As an aid to the control of sand grain size and the dispersion of sizes, the foundry industry needs a more efficient and accurate method of classifying and describing the sand distribution. Classification of the sand is necessary for purchasing specifications, quality control, investigations, testing, and research concerning sand factors. Currently used formulas and techniques for classifying sands are considered inadequate for communication and operational procedures, according to articles by Sanders,[11] Morey,[6] and Booth.[2]

Where several sands are blended in a foundry operation, the grain distribution is susceptible to change by the introduction of core sand, new molding sand, reclaimed system sand, or other special additives. Many minor changes in sand distribution are of little practical significance. However, a vast majority of these changes do result in improper working of the sand, resulting in adverse mold conditions and defective castings. The development and use of an accurate method of describing the sand distribution to maintain predetermined parameters on the sand would provide a more rational approach to sand control with a resulting efficiency of operation, sounder bases for decisions, and a minimum of defective product.

The most commonly used and standard method of classifying sand is the AFS grain fineness number. To classify sand according to this method, a representative sample is first washed by the standard AFS clay washing method[4] to remove the clay size particles which are 20 microns and finer. The remaining sand grains are then dried and passed over the standard sieve series (Table 1). The amount of sand, by weight, retained on each screen is then

TABLE 1 — STANDARD SIEVE SERIES

U.S. No.	Microns	Tyler Screen Scale Equivalent	
		Opening, in.	Mesh
6	3360	.131	6
12	1680	.065	10
20	840	.0328	20
30	590	.0232	28
40	420	.0164	35
50	297	.0116	48
70	210	.0082	65
100	149	.0058	100
140	105	.0041	150
200	74	.0029	200
270	53	.0021	270

TABLE 2 — CALCULATION OF AFS GRAIN FINENESS NUMBER

Sieve Number	Sand Retained		Multiplier	Product
	Weight	Percent		
6	None	.0	3	0
12	None	.0	5	0
20	None	.0	10	0
30	None	.0	20	0
40	.20	.4	30	12
50	.65	1.3	40	52
70	1.20	2.4	50	120
100	2.25	4.5	70	315
140	8.55	17.1	100	1710
200	11.05	22.1	140	3094
270	10.85	21.8	200	4360
Pan	9.35	18.6	300	5580
Totals	44.10	88.2		15243
Clay	5.9	11.8		

$$\text{Grain Fineness Number} = \frac{\text{Total Product}}{\text{Percent Sand Substance}} = \frac{15243}{88.2} = 173$$

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A thesis submitted in partial fulfillment of the requirements for the Degree of Master of Science.

TABLE 3 — ANOTHER AFS NO. 173 SAND

Sieve Number	Sand Retained		Multiplier	Product
	Weight	Percent		
6	None	.0	3	0
12	None	.0	5	0
20	None	.0	10	0
30	None	.0	20	0
40	None	.0	30	0
50	1.80	3.6	40	144
70	14.10	28.2	50	1410
100	1.10	2.2	70	154
140	.30	.6	100	60
200	1.35	2.7	140	378
270	10.90	21.8	200	4360
Pan	14.55	29.1	300	8730
Totals	44.10	88.2		15236
Clay	5.9	11.8		
Grain Fineness Number	$= \frac{\text{Total Product}}{\text{Percent Sand Substance}} = \frac{15236}{88.2} = 173$			

determined. Each sieve has a designated constant multiplier associated with a multiplicand which is the per cent by weight of sand on a given screen. The products thus obtained are then summed and this total product is divided by the per cent total sand grain. Thus one arrives at an average number which is identified as the AFS grain fineness number. An example of such calculations may be seen in Table 2.

By comparing Table 2 with Table 3 one major disadvantage and/or limitation of the AFS number is observed. Two sands with entirely different distributions (thus different physical properties) may have the same AFS grain fineness number. Thus the present method does not give accurate or reliable estimates of a sand's distribution. Moreover, the AFS number does not describe in any reliable way the dispersion or the skewness of the distribution. By a graphic comparison of the sands of Table 2 and Table 3, Fig. 1 illustrates this fact rather forcefully. Graph A has one peak (unimodal) and approximates a normal curve, while graph B has two very distinct peaks (bimodal), yet the AFS numbers are identical for both distributions.

Numerous practical men in the foundry industry recognize the inadequacy of this classification and certain techniques have been substituted in actual foundry practice. The cumulative curve was proposed as a substitute for the AFS grain fineness number

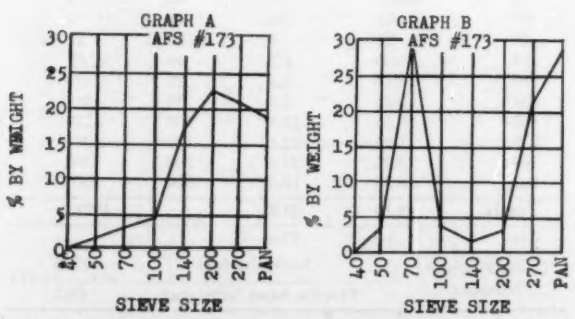


Fig. 1 — Two samples having the same AFS Grain Fineness Numbers but entirely different distributions.

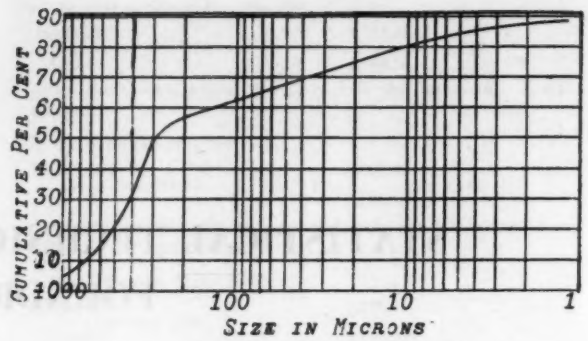


Fig. 2 — Cumulative curve classification.

in 1945 by R. E. Morey and H. F. Taylor.[8] Figure 2 shows this cumulative curve technique.

Data for the cumulative curve are obtained according to traditional sand testing procedures for separating the different-sized sand grains, but the clay region below 20 microns is also included in the cumulative curve classification.[6] The cumulative curve technique may be useful for research when more accurate classification of the finer clay size particles is desired, but the method would be of little value in foundry production and other uses requiring quantitative data.

The clay separation methods are cumbersome, expensive, and cannot be performed with equipment used to separate and classify sand grains. This makes the system less practical for general foundry use. Even if it were an accepted method, it would not be readily adopted to communication or mathematical treatment. A graph of the cumulative curve is the only way to designate or describe the nature of the sand-clay distribution.

Another substitute method for sand classification is the visual analysis technique as proposed by C. A. Sanders.[10] A series of test tubes is used to represent the different screen sizes as shown in Fig. 3. As proposed, the method could be used in the laboratory by placing the sand in a series of test tubes or by graphic representation (Fig. 3) of the per cent retained on each sieve. With either method a bar chart type of representation is obtained.

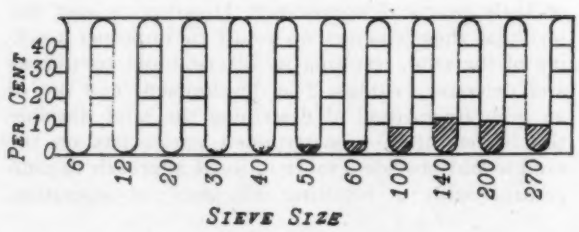


Fig. 3 — Test tube classification.

The use of laboratory test tubes is seriously questioned because it compares the sand by volume using round bottom test tubes. As can be seen by Fig. 4, the volume, by use of round bottom test tubes, is decreased by one-third for heights equal to the radius of the test tubes. Below this height errors are in-

creased, but above this height the error decreases percentage-wise as the height increases. However, this discrepancy can be corrected by use of flat bottom test tubes. The pictorial representation provided by this method can be misleading and is not readily described except by graphic presentation, which in itself presents definite limitations in communication and ease of presentation.

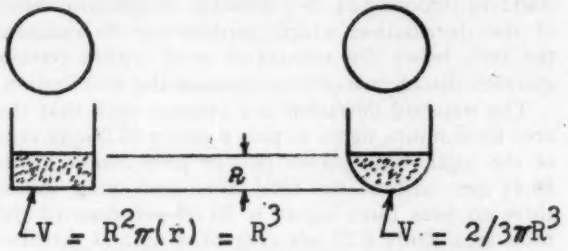


Fig. 4 — Comparison of volumes of test tubes.

In some instances, the AFS grain fineness number has been used along with a statement of the number of screens over which certain percents of the sand were spread (often 90 per cent).[7] This certainly is more informative than the plain AFS grain fineness number, but is merely a way of "Patching" and inaccurate and unreliable method. The popularity of these improvised techniques to describe sand attest to the inadequacy of the AFS grain fineness number.

The objective of the investigation is not merely one of criticism; rather, it is an honest attempt to evaluate present methods, and more important, to develop a method of sand classification which eliminates the above-mentioned inadequacies. The investigation is limited to sand size particles, and is not concerned with clay size particles.

Objectives of a New Method of Classifying Foundry Sands

In view of the evidence which indicates that the AFS grain fineness number and substitute methods of classifying sand do not adequately define the characteristics of foundry sand, and prior to the development of a new method, a set of objectives or qualifications of an "ideal" classification system is necessary.

(1) A new classification system should accurately describe the sand distribution and average size of sand grains (it will be shown that these two parameters are of prime importance in foundry practice).

(2) It should incorporate as much of the present foundry laboratory testing equipment as possible. This is imperative since the cost of new equipment and the development of new testing skills could make a new system unacceptable to the practical foundrymen.

(3) A new system should be capable of presentation in a brief form subject to ease of communication and interpretation. The "ideal" classification method should be equally well understood by different forms of communication, such as bar graphs, numbers, and cumulative curves.

(4) Another qualification of this "ideal" system is that it be simple enough for laymen to know what is meant by the statistics, without necessarily understanding the basic theories of obtaining the classification.

(5) It should also be practicable for both industrial and research uses. That is, it must be reliable and accurate for research purposes and yet simple enough for industrial acceptance.

(6) In addition, a good system will describe a sand in such a manner that other mold and/or casting characteristics can be correlated with it.

With these objectives of an "ideal" classification system in mind, it seems possible that a new classifcation can be developed, which will incorporate many of these objectives and be a great improvement over the present AFS grain fineness number and substitute methods.

Proposed Method of Classification of Sand

A study of the various methods of classifying foundry sands indicates that statistical methods can approach the ideal system of classification. By statistical methods, a system of numbers describing the central tendency, the distribution of the sand grains over the various screens (dispersion), and the skewness of the distribution is proposed.

There are various methods of measuring central tendency, each of which has a more appropriate application in certain situations. The arithmetic mean,[1] $\bar{x}$, is by far, the most common measure of central tendency and is more appropriate in our problem. It can be readily used with other statistics for purposes of tests of significance.

Of the various measures of dispersion, our problem of selecting suitable parameters seems to be a choice between variance and standard deviation. The standard deviation[1] or sigma, δ, will be used because it is easier to comprehend and is a more realistic approximation to the actual distribution. Besides this, it lends itself to the application of control chart procedures, which is an asset in controlling sand properties.

It is known that the standard deviation yields a distribution of sample sigmas which is skewed, and the sample sigmas are biased estimates of the universe sigma; but by the use of a constant[5] related to sample size, it is possible to obtain an unbiased estimate of the population standard deviation. However, as shown on page 685, this constant can be neglected without harmful effects. The standard deviation may also be more readily correlated with other mold conditions — a factor which is of prime importance in research. In a later section this point will be expanded and tests performed.

The most useful measure of lack of symmetry or skewness of a frequency distribution is the skewness factor,[1] K. This measure of skewness is a pure number and may be either positive or negative. For a symmetrical distribution, K is zero. In general, for non-symmetrical distributions, K is negative if the long tail of the distribution extends to the left, the negative direction on the scale of measurement, and

is positive if the long tail extends to the right, the positive direction on the scale of measurement (Fig. 5).

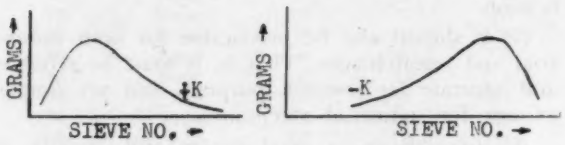


Fig. 5 — Curves showing skewness.

The proposed system for sand classification utilizes a series of three numbers, $\bar{x}/\delta/K$. Following is the method of calculation[1] of the new classification system for the distribution represented in Table 2.

U.S. Sieve Series	Cell (X)	Grams (f)	fX	fX^2	fX^3
6	1	—	—	—	—
12	2	—	—	—	—
20	3	—	—	—	—
30	4	—	—	—	—
40	5	.20	1.00	5.00	25.00
50	6	.65	3.90	23.40	140.40
70	7	1.20	8.40	58.80	411.60
100	8	2.25	18.00	144.00	1,152.00
140	9	8.55	76.95	692.55	6,232.95
200	10	11.05	110.50	1,105.00	11,050.00
270	11	10.90	119.90	1,318.90	14,507.90
Pan	12	9.30	111.60	1,339.20	16,070.40
		44.10	450.25	4,686.85	49,590.25

$$\bar{X} = \frac{\Sigma fX}{\Sigma f} = \frac{450.25}{44.10} = 10.2$$

$$= \frac{\sqrt{(\Sigma f)\ (\Sigma fX^2) - (\Sigma fX)^2}}{44.10} = \frac{\sqrt{(44.1)\ (4686.85) - (450.25)^2}}{44.10}$$

$$= 1.43$$

$$K = \frac{(\Sigma f)^2\ (\Sigma fX^3) - 3\Sigma f\ (\Sigma fX)\ (\Sigma f X^2) - 2\ (\Sigma f X)^3}{(\Sigma f)^3\ (\delta)^3}$$

$$= \frac{(44.1)^2\ (49590.25) - 3(44.1)\ (450.25)\ (4686.85) - 2(350.25)^3}{(44.1)^3\ (1.43)^3}$$

$$= -0.74$$

The proposed system would classify this sand as 10.2/1.43/–0.74 corresponding to $\bar{X}/\delta/K$ respectively. The corresponding number for the distribution of Table 3 is 9.73/2.26/–0.38.

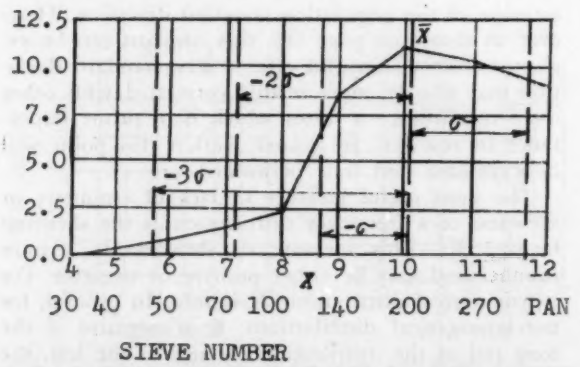


Fig. 6 — Relation of the new classification to the distribution.

A discussion of the meaning of the number of the sand of Table 2 is facilitated by an examination of Fig. 6. The arithmetic mean or common average is indicated by 10.2. This means that the average value of the sand grain sizes is two tenths of the way from the 200 screen between the 200 and 270 screens. The second number, 1.43, is a measure of the dispersion of the distribution. In a normal distribution the total area under the normal curve is one. The standard deviation, δ, is a measure of the dispersion of the distribution which permits one to estimate the area, hence the amount of sand, within certain specified distances from the mean of the distribution.

The standard deviation is a measure such that the area from minus sigma to plus sigma is 68.26 per cent of the total, from minus two to plus two sigma is 95.44 per cent of the total area, and from minus three to plus three sigma is 99.73 per cent of the total area. Only 0.27 per cent of a normal distribution falls, on the average, beyond plus and minus three standard deviations. Between plus and minus sigma of the distribution of Fig. 6, 70.7 per cent of the area falls, and 95.7 per cent falls between plus and minus two sigma.

An examination of Fig. 6 shows that a very small portion of the area falls beyond minus three sigma, but on the positive side of the arithmetic mean the area does not exist much beyond one standard deviation because of the discrete nature of the distribution as shown by the negative skewness factor.

The skewness, K, is a measure of the lopsidedness of the sand distribution and indicates the extent of variation from normality. The skewness factor of minus 0.74 for the distribution describes the fact that the tail of the distribution is skewed to the left. In general, skewness factors up to 0.50 are relatively minor and those above 0.75 can be considered indicative of extreme skewness.

The possible functions of sand distributions on the sieves are discrete in that a constant area system is being dealt with. In practice sands from No. 70 to No. 140 AFS fineness are most common and there is little chance of having extremely skewed distributions. However, if extreme distributions are encountered the skewness factor will describe them.

By use of cells (1,2,3,—) we have coded, in effect, the class intervals provided by the sieves, and assigned them values of one to twelve for ease of computation. One should recognize the non-uniformity of cell intervals based on sieve sizes (Table 1). This lack of uniform cell sizes does not invalidate the statistical methods, since the sand distributions are normal when using the weight retained on each sieve as the observed frequency.

In statistical work the cell intervals for a constant-cause system may be adjusted, as a method of transformation, to obtain normality for certain non-normal populations. This point is further clarified by Moroney,* "It is not essential, though often convenient, to have all the cells the same width." Other methods of transformation such as inverse sine, logarithms, reciprocals, and square root are also used to

*Page 69 of *Facts from Figures*[9]

obtain normality for some non-normal distributions.

The pan is included in the classificaton because it is used at present by sand suppliers and foundries when separating particle sizes, although material retained on the pan is not considered as sand according to Williams.[12] "In lieu of a classification it has become customary to accept as sand grain, those particle sizes which range from those passing through a No. 6-mesh screen and those retained on a No. 270-mesh screen." A breakdown of particle size classfication is illustrated by Fig. 7.

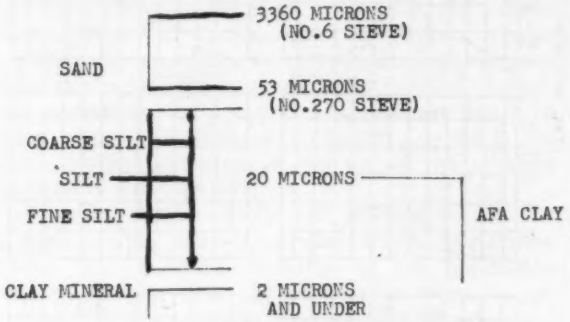


Fig. 7 — Particle classification[12].

Although AFS prescribes the per cent by weight as the characteristic to be used in classifying sand, other sand characteristics were investigated. The most applicable of the other measurements seemed to be the number of grains retained on a given sieve. The number of grains is particularly interesting because the surface area of grains can be correlated with it, and this is important with respect to the bonding surface for binders.

However, use of the number of grains retained on a sieve is impractical for two main reasons. First of all, the number of grains on a particular sieve cannot be readily counted, although it can be estimated approximately from the retained weight. This, of course, suggests the use of the weight as the characteristic to be measured. Secondly, the distribution of the number of sand grains retained is extremely skewed toward the finer sieves, thus making it practically impossible to derive a practical mathematical function to relate the distribution in numbers. Figure 8 shows the relation between sieve size and the number of grains per unit volume.

The weight seems to be the most suitable characteristic on which to classify the sand. Since it is readily obtained according to present testing procedures, no change in equipment is necessary. However, a more important reason for using the weight is the fact that it approximates a normal distribution for a great majority of foundry sands, thus permitting the use of common parameters used in mathematical statistics. When compared with particle size, the weight of sand grains can be viewed as a transformation[3] that gives a more normal approximation to the data.

Twenty-four graphs of different sands, chosen at random from more than a hundred studied, are presented (Fig. 9). These graphs illustrate visually

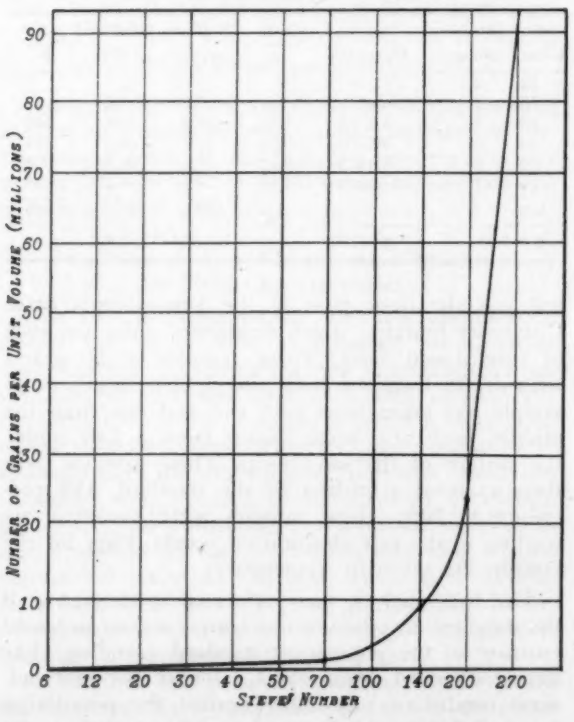


Fig. 8 — Relation of sieve size to number of grains per unit volume.

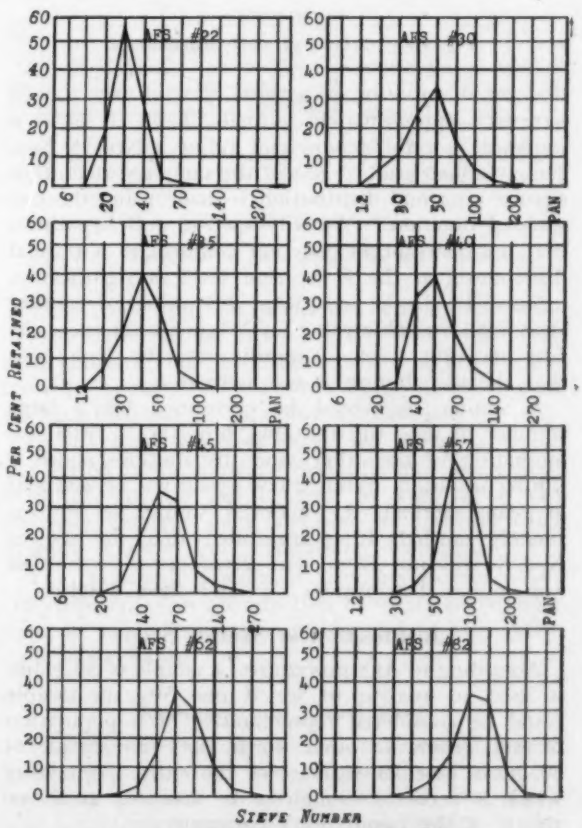


Fig. 9A — Graphs of sand distributions.

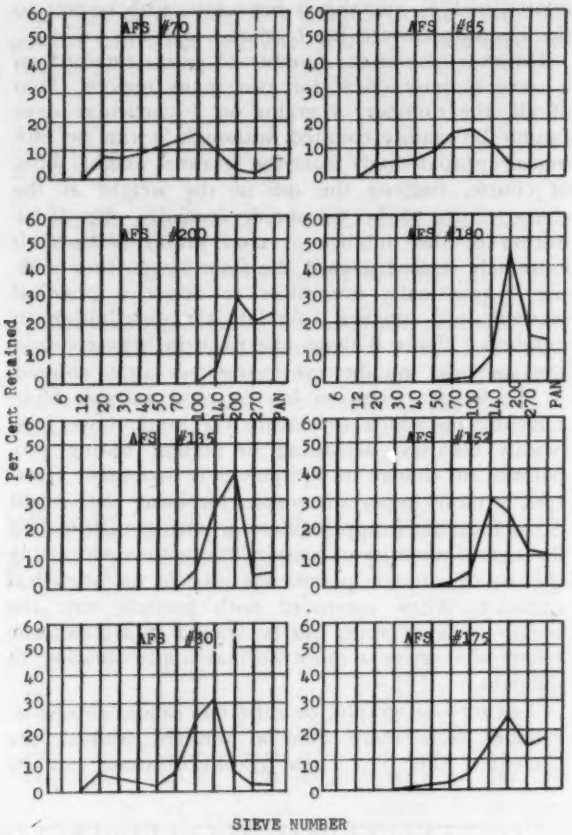


Fig. 9B — Graphs of sand distributions.

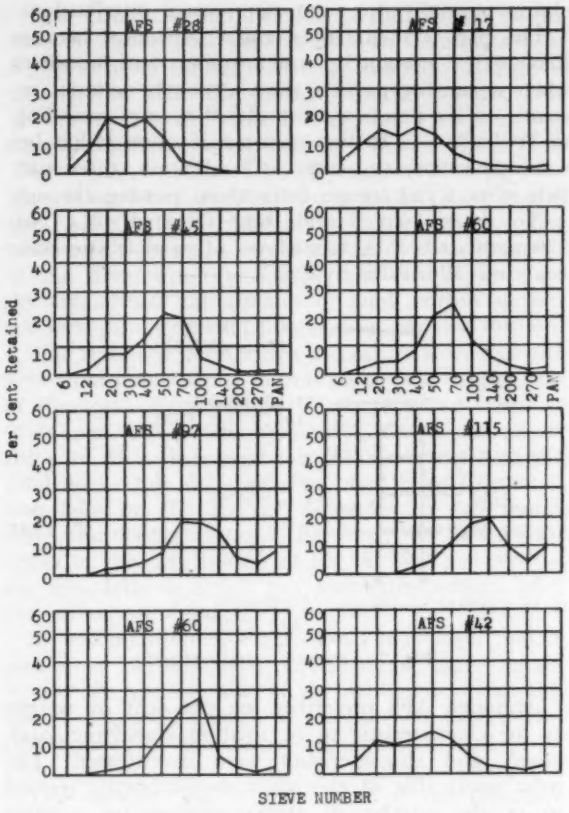


Fig. 9C — Graphs of sand distributions.

that the distribution of weights of sand over the 12 screens is approximately normal. Table 4 shows a statistical test for goodness of fit of a No. 25 New Jersey washed sand. A test of the sample distribution against a normal distribution is made using the chi-squared function.[3] We will test the null hypothesis that the distribution does not conform to a normal distribution at the 5 per cent level of significance.

Since 1.648, the calculated X^2 value, is smaller than 5.99, the table value for $X^2_{.95}$, we reject, at the five per cent level of significance, the hypothesis that the distribution is not normal.

A similar test could be performed for a large number of sand distributions; however, a visual examination of the other sand distributions shows a similar situation. While a few would not be accepted as being normal, the majority would be approximately normal. It is concluded that the use of X/δ/K will not entail any great error because of non-normal distributions on sand sieve weights.

Justification of Sample Size

According to AFS procedures, a sample of 50 grams is used to analyze to see if one 50-gram sample could be considered representative of a population of sand grains. In other words, does one sample of 50 grams give an estimate of the entire population which is accurate enough to be accepted as an estimate of the population's parameters?

The population of sand grains used to run the test was the floor sand of the Pennsylvania State University foundry, which represents about ten tons of conditioned sand. Three samples of 50 grams were drawn from the sand pile of 25-ft length. One sample was taken from each end and one from the middle, each one being drawn from a foot under the surface of the sand heap. These samples were then analyzed according to the standard AFS procedures.[4] Next these samples were classified according to the new clasification system. Data for the samples are given in Appendix A.

The first analysis was performed to determine if the standard deviation of one sample was an unbiased estimate of the population standard deviation. The largest standard deviation, δ_2, (hence the most adverse condition) was tested against the population standard deviation δ_0, which was estimated from the average of the three samples.

TABLE 4 — "GOODNESS OF FIT"

Screen	Mid-point	Interval	$\frac{X-\bar{X}}{\delta}$	F-Theoretical	Observed (F)	$\frac{(f-F)^2}{F}$
12	2					
20	3	Below 3.5	—	7.4	8.0	.049
30	4	3.5-4.5	−1.45	23.1	19.6	.530
40	5	4.5-5.5	−0.51	35.7	39.4	.384
50	6	5.5-6.5	0.42	24.9	26.4	.090
70	7	Above 6.5	1.35	8.9	6.6	.595
100	8			100.0	100.0	1.648

$\bar{x} = 5.05$ $\delta = 1.07$ $X^2\ .95\ (2) = 5.99$

Hypothesis: $\delta_0 = \delta_2$
Level of significance: One per cent.
Statistic:

$$X^2/DF = \frac{\delta 2^2}{\delta 0^2} X^2.005/49; \frac{\delta 2^2}{\delta 0^2} X^2995/49$$

$$\frac{\delta 2^2}{\delta 0^2} = \frac{(2.3)^2}{(2.26)^2} = 1.036 > .555; \ 1.036 < 1.60$$

Conclusion: At the one per cent level of significance there is no reason to suspect $\delta 0$ and $\delta 2$ are significantly different.

From this test we conclude that the sample standard deviation is an unbiased estimate of the population standard deviation. Since a 50-gram sample standard deviation can be representative of the population standard deviation without a constant depending on sample size and since this constant would be close to unity (0.985) for a 50-gram sample, it has been decided to eliminate the use of the constant mentioned on page 681.

A second test was performed to judge whether or not the sample means could estimate the population mean. The standard deviation of the population was used along with the population mean, u_0, estimated as the average of the three samples.

Hypothesis: $u_0 = x_2$
Level of significance: one per cent.

Statistic: $$Z = \frac{x_2 - u_0}{\delta o / \sqrt{N}} < Z_{\frac{1}{2}A}; \frac{x_2 - u_0}{\delta o / \sqrt{N}} > Z_{1-\frac{1}{2}A}$$

$$\frac{x_2 - u_0}{\delta_0 / \sqrt{N}} = \frac{7.81 - 7.90}{2.26 / \sqrt{50}} = -0.281 > -2.56; \ -0.281 < 2.56$$

Conclusion: At the one per cent level of significance there is no reason to suspect u_0 and x_2 are significantly different.

From the test we conclude that the sample mean of a 50-gram sample is an unbiased estimate of the population mean. From these two tests we conclude that a 50-gram sample is adequate to estimate the parameters of a sand heap or system.

Relation of New Classification System to Mold Characteristics

One of the objectives of the new classification system was to be able to correlate sand and/or mold characteristics with the new classification number. Several tests were conducted with synthetically constructed sand distributions to see if a correlation existed between permeability and green compressive strength, and the classification numbers. Sand distributions were synthetically constructed by collecting sands passed through the standard sieve series and weighing out the distribution according to predetermined values of the new classification system. Data from these tests are listed in Appendix B. Recorded values for compressive strength and permeability are the average of two readings.

In the first test, both the standard deviation and the skewness factor were held constant as the mean was varied. Results of the test are illustrated in Fig. 10. From the graph it is seen that the permeability decreases with increasing mean or decreasing grain size, and green compressive strength has no significant correlation to the mean.

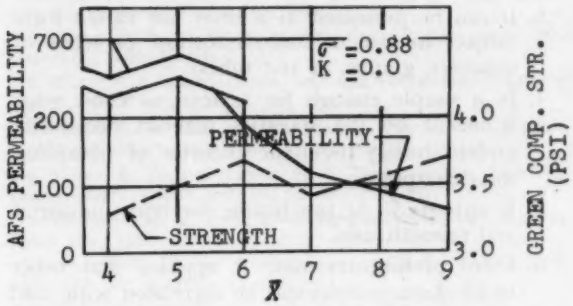


Fig. 10 — Variation of sand properties with arithmetic mean.

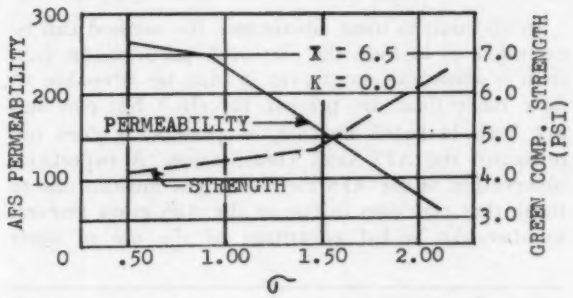


Fig. 11 — Variation of sand properties with standard deviation.

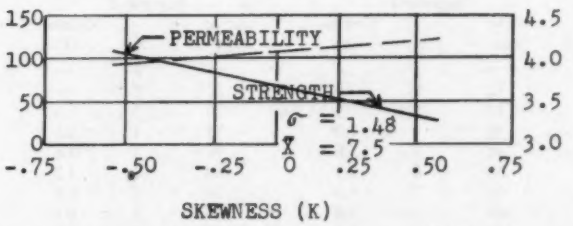


Fig. 12 — Variation of sand properties with skewness.

The second test was performed varying the standard deviation as the mean and skewness factor were held constant. Fig. 11 illustrates the results of the test. It appears that permeability decreases and strength increases as dispersion increases.

In the final test performed, the skewness factor was varied as the standard deviation and mean were held constant. As can be seen by Fig. 12, as the skewness changes from negative to positive, the strength increases but permeability decreases.

The three tests performed were not extensive but are proposed as an introduction to what can be done with the new classification system and to help verify its practical values.

Summary and Conclusion

The objectives of an "ideal" classification system for foundry sand have been presented and a system has been proposed which approximates the ideal. The objectives now become the advantages of the proposed method.

1. It accurately describes the sand distribution, average grain size, and skewness of the distribution.
2. It makes use of present-day testing equipment.
3. It can be presented in a brief but varied form subject to ease of communication (a series of numbers, graphs, or test tubes).
4. It is simple enough for laymen to know what is meant by the statistics without necessarily understanding the basic theories of obtaining the description.
5. It appears to be practicable for both industrial and research uses.
6. From preliminary tests it appears that other mold characteristics can be correlated with, and evaluated by, the description.
7. It is reliable and accurate.

In addition to these advantages, the method can be extended to include the clay-sized particles. In fact, from a statistical standpoint it may be advisable to have more than the present 12 cells,[5] but clay has not been included because, at present, it does not influence the AFS sand classification. A superficial observation of the AFS method might mislead one to think that clay does influence the AFS grain fineness number. An added advantage of the use of these statistics is the fact that control chart procedures of statistical quality control can more readily and accurately be applied to maintain the sand distribution within predetermined limits.

Recommendations for Future Study

In the course of this investigation, two main problems for future study evolved. One of these problems involves the number of cells used in the classification. Although the use of 12 cells is statistically sound,[9] experiments could be performed to determine whether it is advisable to use more cells, possibly by subdividing the pan size and also dividing the clay particles.

The other problem is related to the correlation between sand characteristics and the new classification parameters. Since the results of tests performed showed that correlations exist, future tests could be designed to further test these correlations with other mold characteristics and possibly even be extended to include interactions among variables.

APPENDIX A

Sample 1

Sieve Size	X	Grams (F)
6	1	—
12	2	—
20	3	.9
30	4	2.3
40	5	3.0
50	6	5.3
70	7	6.8
100	8	9.6
140	9	6.4
200	10	3.5
270	11	1.6
Pan	12	4.7
		44.1

Clay = 5.9 Grams
$\bar{x}_1 = 7.93$
$\delta_1 = 2.25$

Sample 2

Sieve Size	X	Grams (F)
6	1	—
12	2	—
20	3	1.4
30	4	2.5
40	5	3.6
50	6	4.7
70	7	6.3
100	8	9.8
140	9	6.6
200	10	3.5
270	11	1.6
Pan	12	4.2
		44.2

Clay = 5.6 Grams
$\bar{x}_2 = 7.81$
$\delta_2 = 2.30$

Sample 3

Sieve Size	X	Grams (F)
6	1	—
12	2	—
20	3	1.5
30	4	1.7
40	5	2.4
50	6	5.3
70	7	6.6
100	8	10.4
140	9	7.0
200	10	3.6
270	11	1.6
Pan	12	4.5
		44.6

Clay = 5.4 Grams
$\bar{x}_3 = 7.96$
$\delta_3 = 2.33$

$$\delta_o = \frac{\Sigma\delta}{N} = \frac{2.25+2.30-2.25}{3}$$

$$= 2.26$$

$$u_o = \frac{\Sigma\bar{x}}{N} = \frac{7.93+7.81-7.96}{3}$$

$$= 7.90$$

APPENDIX B

δ and K constant; $\delta = 0.88$, $K = 0$

Sample 1 Sieve Size	Grams	Sample 2 Sieve Size	Grams	Sample 3 Sieve Size	Grams	Sample 4 Sieve Size	Grams
12	1.5	30	1.5	40	1.5	70	1.5
20	13.5	40	13.5	50	13.5	100	13.5
30	20.0	50	20.0	70	20.0	140	20.0
40	13.5	70	13.5	100	13.5	200	13.5
50	1.5	100	1.5	140	1.5	270	1.5
$\bar{x}_1 = 4.0$		$\bar{x}_2 = 6.0$		$\bar{x}_3 = 7.0$		$\bar{x}_4 = 9.0$	
Perm = 700		Perm = 173		Perm = 125		Perm = 46	
C. Str = 3.3		C. Str = 3.6		C. Str = 3.4		C. Str = 3.6	

$\bar{x}$ and K constant; $\bar{x} = 6.5$, $K = 0$

Sample 1 Sieve Size	Grams	Sample 2 Sieve Size	Grams	Sample 3 Sieve Size	Grams	Sample 4 Sieve Size	Grams
						12	1.25
						20	2.5
				30	5.0	30	5.0
		40	7.5	40	7.5	40	7.5
50	25.0	50	17.5	50	12.5	50	8.75
70	25.0	70	17.5	70	12.5	70	8.75
		100	7.5	100	7.5	100	7.5
				140	5.0	140	5.0
						200	2.5
						270	1.25
$\delta_1 = 0.50$		$\delta_2 = 0.92$		$\delta_3 = 1.43$		$\delta_4 = 2.06$	
Perm = 270		Perm = 250		Perm = 173		Perm = 112	
C. Str = 4.3		C. Str = 4.4		C. Str = 4.6		C. Str = 6.4	

$\bar{x}$ and constant; $\bar{x} = 7.5$, $\delta = 1.48$

Sample 1 Sieve Size	Grams	Sample 2 Sieve Size	Grams
50	12.5	30	2.5
70	12.5	40	5.0
100	10.0	50	7.5
140	7.5	70	10.0
200	5.0	100	12.5
270	2.5	140	12.5
$K_1 = 0.53$		$K_2 = -0.53$	
Perm = 83		Perm = 105	
C. Str = 4.3		C. Str = 3.9	

Acknowledgments

The senior author expresses his sincere appreciation to Professor David C. Ekey whose guidance was most helpful. To Roger L. Yard the author is particularly indebted for assistance, and to Professor Clarence E. Bullinger and Dr. Clifton A. Anderson for their guidance and suggestions. The author also wishes to express his appreciation to the Foundry Educational Foundation, and to the committee on scholarship and awards whose financial assistance made his graduate work possible. Appreciation is also expressed for ideas and information from B. H. Booth, F. S. Brewster, L. Dill, R. E. Morey, and J. S. Schumacher who took time to give thought to the subject.

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METALLOGRAPHY FOR THE FOUNDRYMAN

GOOD METALLOGRAPHIC PRACTICE

By

A. M. Montgomery*

Polished and etched metal specimens were first examined microscopically in the last half of the nineteenth century by Sorby, Martens, Osmond, Le-Chatelier and others. Since that time, hundreds of scientific papers have been written describing the microscopic appearance of metals ranging from lithium to the transuranium elements.

In this *Symposium*, we are concerned with perhaps less than ten of the 74 metallic elements and particularly with the cast condition of these metals and their alloys. The specific details concerning the preparation and examination of metallographic specimens have been explained in many publications and will be further examined by the other authors in this *Symposium*. In the evolution of these metallographic procedures, however, certain general practices have been developed which are applicable to all metals and alloys. The following discussion will enumerate some of these in the order in which they would be used.

Sample Selection

The nature of the solidification process prevents attainment of complete microstructural homogeneity in castings. Variation of section size, different molding materials, the use of chills and risers, and the location of gates all contribute to structural variations. In sampling castings for metallographic examination, consideration must be given these factors. Thus the structure of a casting adjacent to a chill or under a riser would be respectively finer and coarser than the average structure of the casting. A sufficient number of specimens should be chosen to reflect these variations. Once the variation has been determined the number of samples can be reduced. In this connection, however, it is better to take too many specimens than to take too few, thereby allowing some important feature to escape notice.

Preliminary macroscopic examination, particularly under a low power binocular stereoscopic microscope, is frequently invaluable in the selection of metallographic specimens from a casting. Surface shrinkage, reaction between metal and molding material, and coarse grains can often be detected macroscopically before cutting into a casting. The examination of fractures using a microscope of low power can reveal the presence of porosity, dross, inclusions, coarse grains, coarse intermetallic compounds, and the source and course of fatigue fractures. Macroscopic examination of machined and etched sections from castings will frequently reveal porosity, grain size and segregation. In short, preliminary macroscopic examination can often reveal local characteristics not readily visible to the naked eye that should be further explored and confirmed by metallographic examination.

If some local condition, such as shrinkage, a specific surface condition, or the nucleus of a fatigue crack, is the object of examination, it is important that a reference sample be removed from a region known to be free from the local condition and representative of the average structure of the casting. In this way, a comparison can be made between the local structure that is segregated, porous or drossy and the representative structure. A comparison can be made more effectively between two adjacent metallographic specimens on the same microscope than between a single sample and a standard micrograph or an image retained by memory.

Sample Removal

A cardinal rule in the removal of metallographic samples from a casting is that the removal should result in no distortion or change in the microstructure of the sample (Fig. 1). With ferrous or other hard alloy castings, it may be necessary to use abrasive cut-off wheels, but they should be liberally flooded with water to prevent overheating that could alter the structure. Likewise cutting torches may be necessary in the removal of specimens, but the consequent heating effect should be considered and the heat affected zone removed by other means. In some cases, the width of the kerf removed by a hack saw may be excessive and the use of a fine jeweler's saw may be necessary. To sample very large castings

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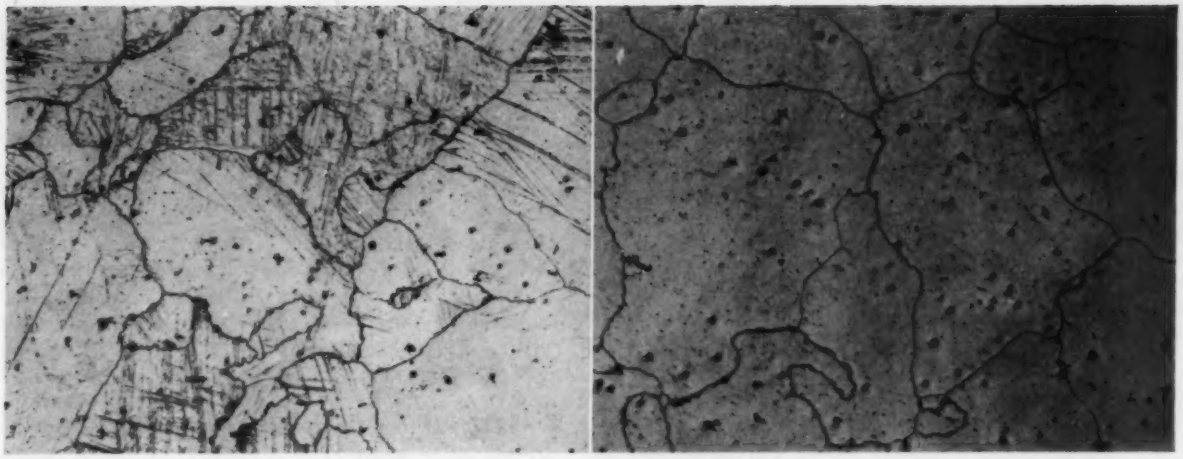


Fig. 1 (A - Left) — Strain lines in magnesium-base solution heat treated alloy resulting from hammering during sample removal; (B - Right) — Same specimen as in (A) but removed from casting by sawing. Etch — Glycol. Mag. — 100X.

that are not readily movable, a trepanning drill may be a useful tool.

The proper identification of specimens is particularly important when several samples are to be removed from a single casting. Several waterproof ink marking pens satisfactory for this purpose are available. Vibrating marking pencils and steel stencils are also useful. A sketch of the casting or a photograph showing the location of samples is a very desirable aid, particularly in the analysis of a fractured casting.

Removal of the sample from the casting should be followed by a cleaning operation to remove any oil or water-soluble coating or deposit that might interfere with subsequent polishing or etching. Generally oil or grease may be removed by washing in gasoline or trichlorethylene and water soluble deposits may be removed by scrubbing with a stiff brush under flowing water.

Mounting of Samples

Specimens that are too small to hold during grinding and polishing or specimens with surface or edge features to be studied are usually mounted in plastic. Such mounts have generally supplanted metal frame or pack mounting because of the greater convenience involved. Filled bakelite and lucite are the two plastics most generally used. Both require the application of heat and pressure during the mounting process. Several plastics are available that do not require additional heat or pressure. These mounting types are particularly convenient where many samples are to be prepared at one time and a production line technique can be employed.

The edges of mounted hard metal specimens may become excessively rounded because of the faster polishing rate of adjacent supporting plastic. A preliminary electrodeposit, such as pure iron, on ferrous samples, the inclusion of metal shot around the specimen in the mount or wrapping with metal foil, all serve to maintain flat edges on the sample.

Grinding of Samples

At this stage of specimen preparation, the avoidance of any practice that might cause distortion of structure should be emphasized again. Particularly, care should be taken to properly cool specimens during the grinding operation. If grinding is a hand operation, a negligible amount of heat is generated. On the other hand, if the grinding papers are mounted on a wheel or an abrasive wheel is used, the specimen should not be allowed to become hot to the touch. One relatively new development in grinding is the water resistant abrasive paper. These papers permit the use of water as both coolant and lubricant instead of kerosene. In addition to its obvious advantages, water is a more efficient coolant. Grinding on a specific abrasive should be continued until the previous surface marks have been completely removed and should be prolonged for an equal period of time thereafter to insure the removal of distorted metal. The amount of metal removed during the subsequent polishing operation is much less than in the grinding procedure; therefore, adequate grinding is essential.

Polishing of Samples

The polishing operation is a continuation of the metal removal initiated by grinding but on a much finer and slower scale. Because finer abrasives are used and scratch-free surfaces are being produced, good housekeeping is mandatory at this stage to prevent the carry-over of coarser abrasive particles and the introduction of foreign, hard particles which could produce scratches on the specimen of a depth that would not be removed without regrinding. Before proceeding from one polishing stage to the next, it is a good practice to carefully wash the prior abrasive from the specimen and from the hands with soap and water. Certain abrasives such as magnesium oxide can react in time with the air to form coarse, hard, scratchy particles so that frequent cleaning of excess abrasive from the polishing cloth and wheel are recommended. Air-borne dust from open windows can lodge on wet polishing wheels and cause scratches. The wheels should be kept wet and covered when not in use.

The polishing operation has an optimum duration.

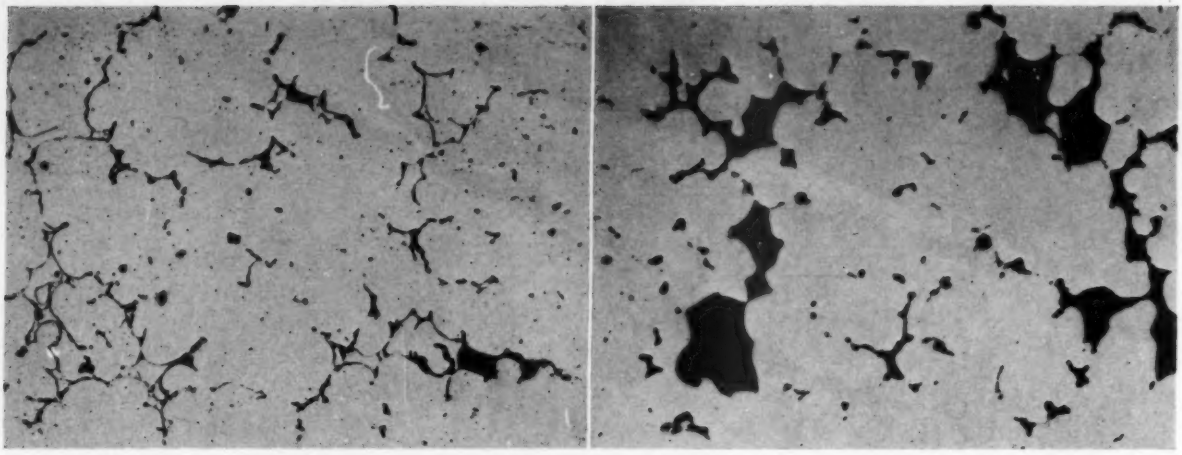


Fig. 2 (A - Left) — Porosity in aluminum-magnesium alloy as revealed by insufficient polish; (B - Right) — Same area as in (A) after adequate polish. Etch — None. Mag. — 100X.

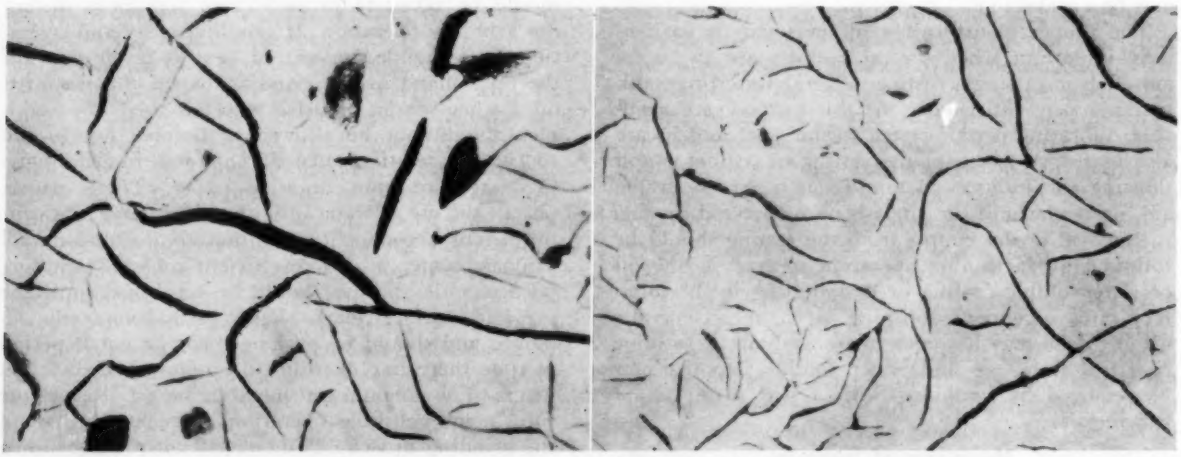


Fig. 3 (A - Left) — Gray cast iron polished too long showing exaggerated graphite flakes; (B - Right) — Same as (A) but properly polished. Etch — None. Mag. — 500X.

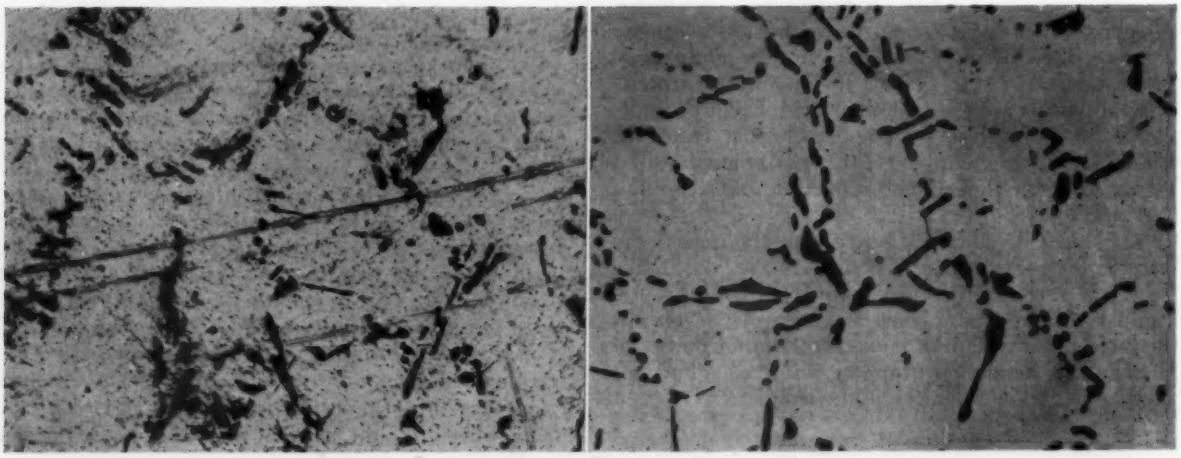


Fig. 4 (A - Left) — Poor mechanical polish on an Al-Si-Mg alloy showing fine debris embedded in solid solution matrix; (B - Right) — Same material as (A) but electrolytically polished. Note clean matrix. Etch — None. Mag. — 250X.

If the sample is not polished long enough, certain details such as porosity may not be revealed (Fig. 2). On the other hand if polishing is excessively prolonged, exaggerated relief of certain microscopic features may result (Fig. 3). Only practice can reveal the optimum polishing time to the metallographer.

One method that can be used to attain a flow-free surface is electrolytic polishing. Specimens prepared in this manner may occasionally be difficult to interpret because of the relief of microstructural constituents resulting from variable polishing rates. However, the degree of polish of the matrix solid solution attained by the electrolytic method is such that it can be used as a referee method in judging the adequacy of a mechanical polish. What may appear as finely divided precipitate in an improperly mechanically polished specimen, may by its removal upon electrolytic polishing be revealed to be embedded abrasive or constituent debris (Fig. 4).

Following polishing, the specimen, while still wet, should be washed in a stream of hot water or alcohol and blown dry to prevent water stains from forming.

Etching of Samples

Etching of a polished metallographic sample serves two purposes. It removes metal distorted by the polishing operation and it delineates or colors certain features of the microstructure. The degree of etching may vary, depending on the subsequent method of examination. Macroscopic examination generally requires deep etching. Microscopic examination requires less etching with only a mild etch necessary for the highest magnification to prevent impairment of resolution of fine details.

Photomicrography

A microstructure, once developed by polishing and etching, can be examined and interpreted by the metallographer but in order that others distant in time or space can interpret the structure, it is necessary to make a photomicrograph. As in the selection of the metallographic specimen, the choice of a representative field for photomicrography must be made with care, because variations in structure will be encountered in the same sample, particularly if high magnification is used. If a specific, local feature is to be emphasized, it is generally good practice to photograph the feature in a series of micrographs of successively higher magnification to show its relationship to the rest of the microstructure. In any case, a good rule to follow is to use the lowest magnification that resolves all of the pertinent detail.

Although most photomicrographs are produced using bright field illumination, the use of dark field, oblique, and polarized light illumination is frequently of considerable value in the examination and photography of such materials as transparent or translucent inclusions, sand grains, and various types of anodic or organic coatings.

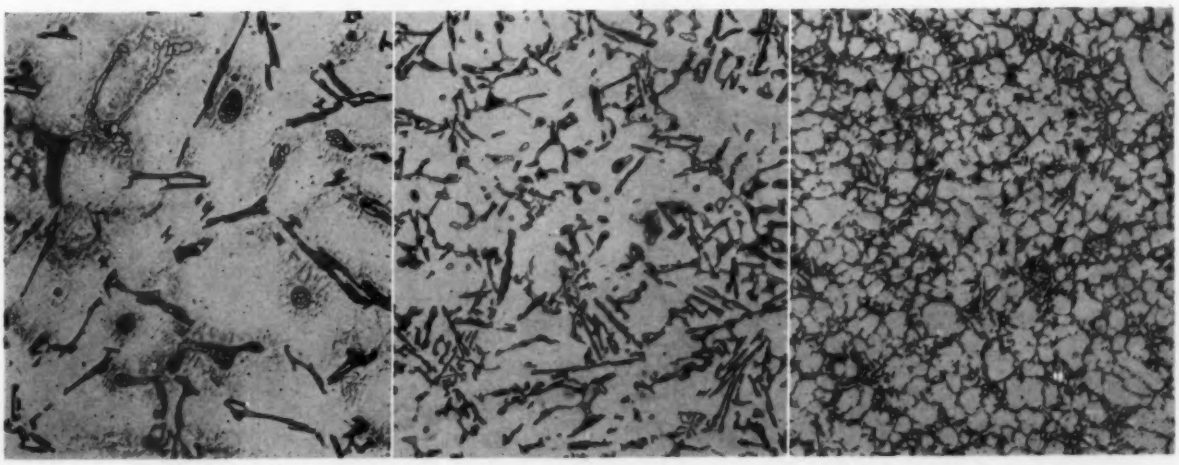
Metallographic plates afford the best negative material for photomicrography. However, plates present a storage problem since they are bulky and fragile. One or two manufacturers of photographic materials produce the metallographic emulsion on a film base which is reasonably satisfactory film for routine photomicrography.

To obtain the best photographic results, the recommendations of the microscope and photographic materials manufacturers should be closely followed.

Interpretation of Photomicrographs

The interpretation stage of metallography is the most important operation, for unless proper interpretation can be made, the most careful sample preparation is worthless. When careful sample preparation is combined with intelligent interpretation, a wealth of information can be obtained. No substitute can be found for experience in the interpretation of microstructures, and an intimate knowledge of the metallurgical processes that have produced the sample is invaluable in this interpretation.

Metallographic examination can enable the foundryman to determine the approximate rate of solidification of a particular section and the method of casting used. As solidification rate increases, the mi-



Sand Cast | Permanent Mold Cast | Die Cast

Fig. 5 — Sand-, permanent mold-, and die-cast structures of an Al-Si-Cu alloy showing refinement with increasing solidification rate. Etch — 0.5% HF. Mag. — 250X.

crostructure becomes finer; sand castings, permanent mold castings and die castings show successively finer structures (Fig. 5). In some alloys, the approximate chemical composition can be judged from the microstructure. The type and effects of various heat treatments can be judged. The analysis of casting defects is greatly facilitated by metallographic examination. The investigation of service failures would be seriously handicapped without metallography.

The foundryman who has hitherto not used metallography will find that the microscope can become an indispensable tool comparable to chemical analysis and mechanical testing in the improvement of his product. It is hoped that the material presented in this *Symposium* will arouse the interest of foundrymen in that direction.

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METALLOGRAPHY OF ALUMINUM ALLOYS

By

L. F. Mondolfo*

Use of the microscope in the aluminum foundry can give valuable information, some of which cannot be obtained by any other means. To get the most benefit from a metallographic investigation the study should be started by a macro-examination. From it not only information can be gathered as to grain size, porosity, segregation, etc., but help can be obtained on the selection of representative samples for micro-examination.

Once the samples have been prepared the examination should be started on the as-polished specimen at low magnification. This gives a general view of the specimen, which then can be followed by more detailed examinations at high magnifications and on etched specimens. The following examples will illustrate the recommended technique and the results that can be obtained.

355 Alloy Sand Castings

The first example covers an investigation on 355 alloy sand castings which were examined both as-cast and heat treated, to determine not only the quality of the castings, but also the quality of the operations in the foundry producing the castings. In Fig. 1 is shown a typical section at low magnification. As it can be seen there is some gas porosity, otherwise the distribution and size of the constituents are normal. A check of the quality of the ingot used (Fig. 2) indicates that the gas was present in the ingot. At higher magnification (Fig. 3) the structure of the casting is quite different from the normal (Fig. 4). In the casting there is no Mg_2Si visible and a mottling of the matrix is present that, when resolved at high magnification (Fig. 5), shows a Widmanstatten pattern caused by a precipitate. Both these abnormalities point to a very slow cooling of the casting, as to be obtained in castings poured in dry sand and from an excessively high temperature. Also visible in Fig. 3 are many long needles of β (FeSi) which are indicative of a high iron content in the alloy. In Fig. 6 is shown a needle of the iron silicon compound. The etching has revealed a cored structure, in which the center is δ (FeSi) and the outer rim β (FeSi). Such a structure can appear only if the δ (FeSi) formed first during solidification and then reacted peritectically with the liquid to produce β (FeSi). The δ compound can form from the liquid only if the iron content is above 2 per cent. Although it is evident that such a high iron content

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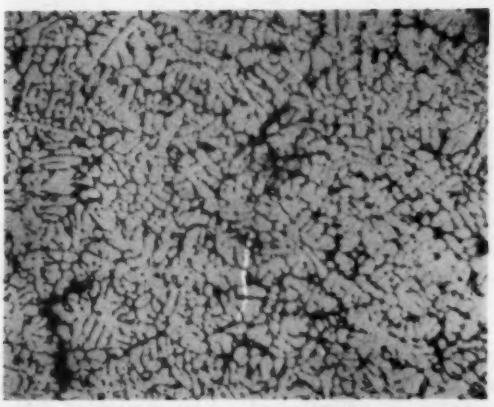


Fig. 1 — Alloy 355 as-cast. Unetched. Black is gas porosity. Mag.— 50X.

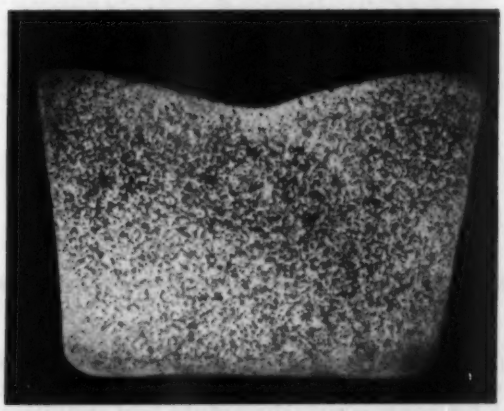


Fig. 2 — Alloy 355 ingot, half size. Black dots are gas porosity. Tucker's etch.

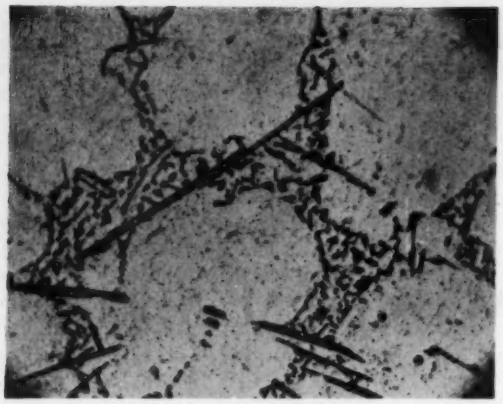


Fig. 3 — Same as Fig. 1 (Alloy 355 as-cast). Unetched. Long needles of β (FeSi); dark, Si; light globules, $CuAl_2$. Mag. — 250X.

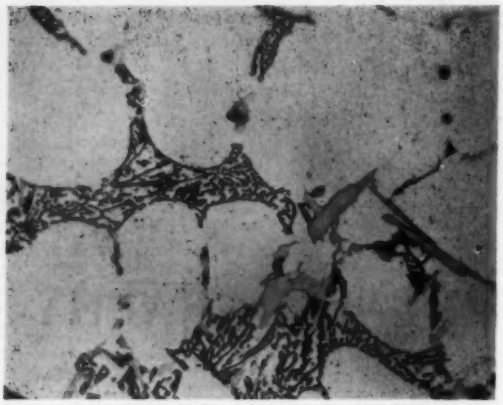


Fig. 4 — Normal structure of Alloy 355 as-cast. Unetched. Dark gray, Si; black Mg_2Si; outlined gray β (FeSi); outlined light, $CuAl_2$; very light, AlFeMgSi.



Fig. 5 — Same as Fig. 3 (Alloy 355 as-cast). Unetched. Widmanstatten pattern of precipitate. Mag. — 1000X.



Fig. 6 — Same as Fig. 3 (Alloy 355 as-cast). Keller's etch. Light δ (FeSi) surrounded by β (FeSi) etched. Dark gray, Si. Mag. — 1000X.

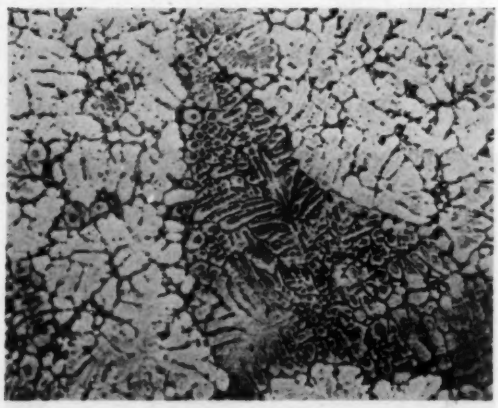


Fig. 7 — Alloy 355-T4; presence of some coring. Keller's etch. Mag. — 50X.



Fig. 8 — Alloy 355 as-cast; heavy coring. Keller's etch. Mag. — 50X.

in the alloy is localized, still a high average iron content is necessary to produce such a segregation.

A low-power examination of the heat treated structure (Fig. 7) shows that, although the heat treatment has greatly reduced the coring in the as-cast structure (Fig. 8) it has not eliminated it completely. The amount of coring still in the casting does not affect the mechanical properties to any great extent; however it is indicative of a heat treatment deficient either as to temperature or time. Figure 9 immediately excludes too low a temperature; the hole visible at the top indicates a small amount of eutectic melting.

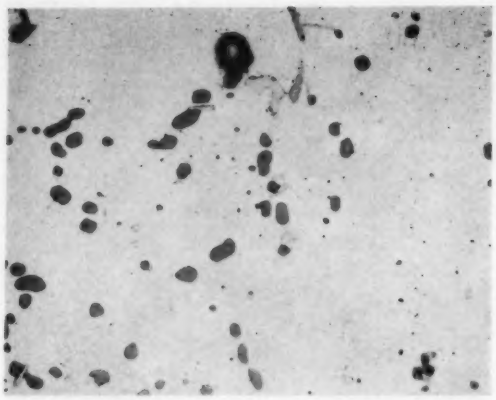


Fig. 9 — Same as Fig. 7. Unetched. Black spot of eutectic melting, excessive rounding of Si and Fe compounds, reduced amount of Si. Mag. — 250X.

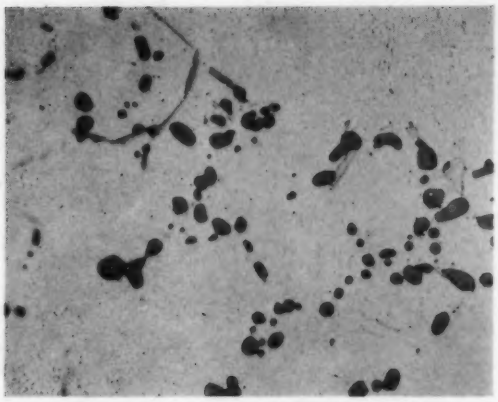


Fig. 10 — Alloy 355, properly heat treated. Unetched. Mag. — 250X.



Fig. 11 — Leaky sand casting, machined surface showing porosity. Unetched. Actual size.

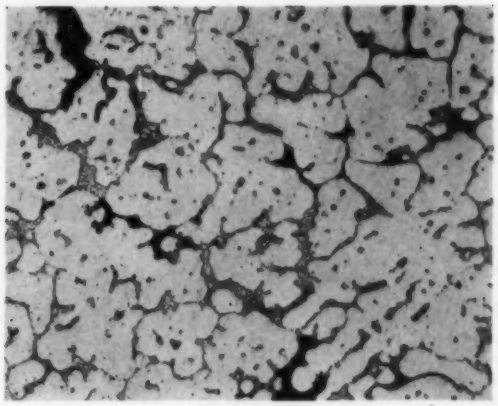


Fig. 12 — Same as Fig. 11; porosity at grain boundaries. Unetched. Mag. — 50X.

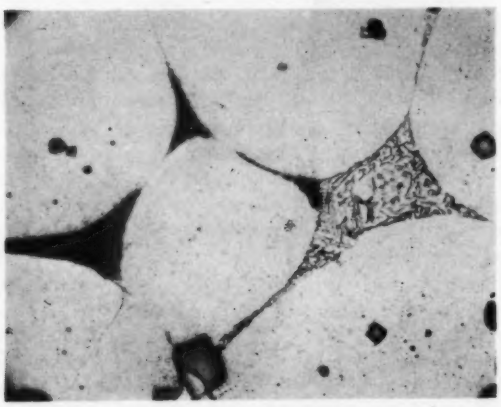


Fig. 13 — Same as Fig. 11; porosity at grain boundaries very finely divided of eutectic. Unetched. Mag. — 250X.

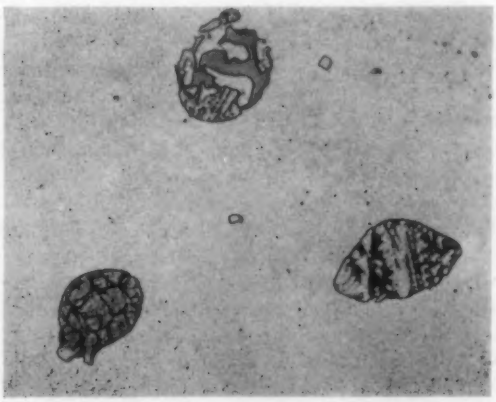


Fig. 14 — Same as Fig. 11; "rosettes" typical of eutectic melting. Unetched. Mag. — 1000X.

The high temperature of heat treatment is also confirmed by a comparison with a properly treated sample (Fig. 10) which shows more silicon particles and less rounding of the constituents.

Assuming that the samples examined are truly representative it can be said that the quality of the castings and their heat treatment are such that the castings probably meet specifications. The ingot is gassed and has a very high iron content. The foundry operations are somewhat lax. The control of pouring temperature is poor, but the gating and risering technique are good, since little or no shrinkage was found.

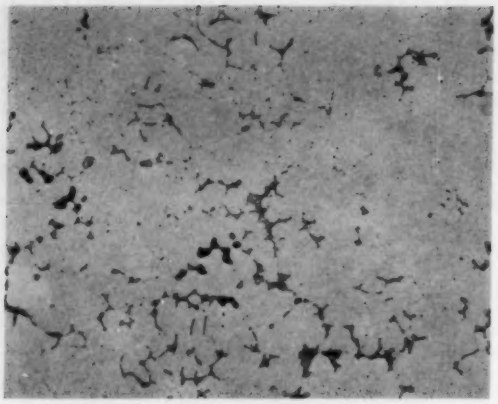


Fig. 15 — Casting in 195 alloy. The black areas are porosity. Unetched. Mag. — 50X.



Fig. 16 — Same as Fig. 15; α(FeSi) as "chinese script"; $CuAl_2$, light; dark globules, Si. Unetched. Mag. — 250X.

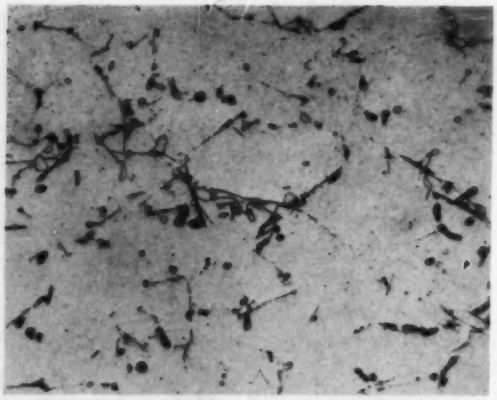


Fig. 17 — Alloy B195. Unetched. Dark globules, Si; needles, β (FeSi); light, $CuAl_2$. Mag. — 250X.

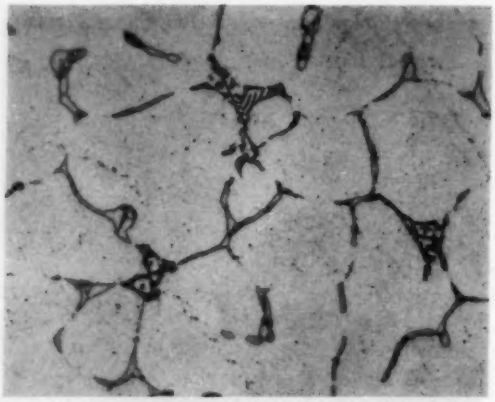


Fig. 18 — Alloy 195 as-cast. Unetched. Dark "chinese script", α(FeSi); gray dots, Si; light, $CuAl_2$. Mag. — 250X.

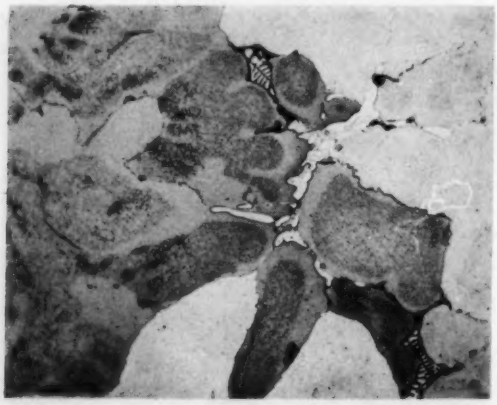


Fig. 19 — Alloy 195-T4, improperly treated. Keller's etch. Notice coring around grain boundaries. Mag. — 250X.

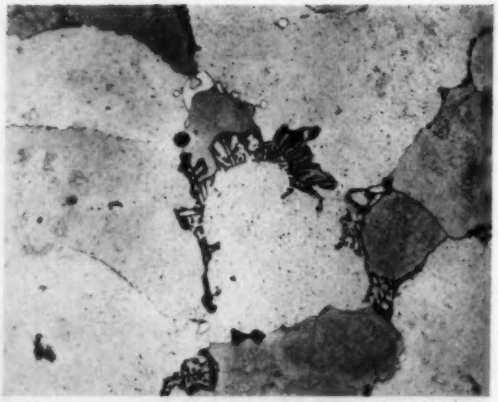


Fig. 20 — Alloy 195-T4, properly treated. Keller's etch. Absence of coring at grain boundaries. Mag. — 250X.

The heat treatment technique is on the ragged edge, times are short and temperatures slightly too high.

356 Alloy Sand Castings

The next example covers a batch of castings made of 356 alloy in the T6 temper, which are expected to be pressure-tight. Most castings of this batch are found to leak when tested, and micro-examination is used to determine the cause. An examination of a machined surface (Fig. 11) shows porosity and cracks in the casting that are apparently caused by lack of feeding. Examination at low magnification (Fig. 12)

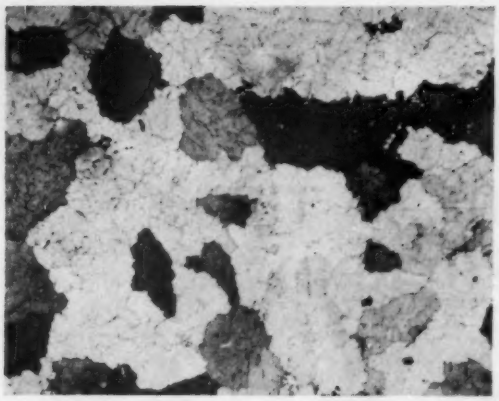


Fig. 21 — Alloy 195-T4. Keller's etch. Notice high contrast between grains. Mag. — 50X.

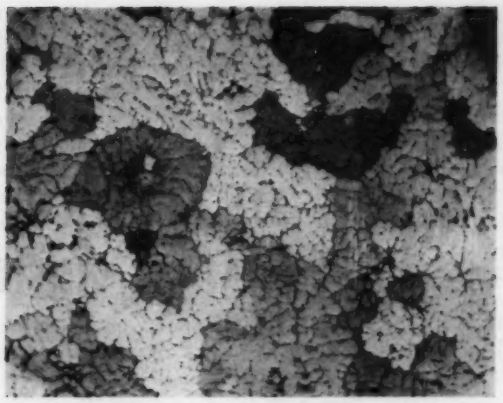


Fig. 22 — Alloy 195-T6. Keller's etch. Notice low contrast between grains. Mag. — 50X.

shows porosity at the grain boundaries, which is consistent with shrinkage porosity. Examination at higher magnification gives a somewhat different picture: (Fig. 13) the porosity is limited to the grain boundaries and the appearance of the constituents is that of a cast material rather than a heat-treated one. Also clearly visible are "rosettes" (Fig. 14) which are typical of eutectic melting. Thus the micro-examination changes the picture from one of poor gating and risering to one of over-heating in the heat treatment. The blame consequently shifts from the molding department to the heat treatment department.

195 Alloy Sand Castings

The next case is a general investigation of castings made of 195 alloys. The low magnification (Fig. 15) shows some porosity and a normal distribution of constituents. The size and shape of the constituents indicates a sand casting; high magnification (Fig. 16) positively identifies the alloy as 195 and not B 195 (Fig. 17) which is the permanent mold alloy. The main difference between the two is in the appearance of the iron-bearing constituents: α (FeSi) in the form of "Chinese script" in the 195, β (FeSi) needle form in the B 195. Figure 16 also indicates that the alloy is heat treated since the $CuAl_2$ constituent does not form an almost continuous network at the grain boundaries as in the cast structure (Fig. 18). However an appreciable amount of $CuAl_2$ is present, either due to an improper heat treament or to a local segregation. By etching, a decision can be reached between the two choices. If the heat treatment was not sufficient to dissolve all the $CuAl_2$, the matrix around it will show coring (Fig. 19); if instead there is segregation, the coring will not be present (Fig. 20). The etching also can be used to decide the temper. In the T 4 temper the contrast between grains is very pronounced (Fig. 21); in the T 6 temper the contrast is more limited (Fig. 22).

Most aluminum foundrymen use titanium as a grain refiner and many of them have wondered how and why it works. Figure 23 shows how refinement is produced. As it can be seen the dendrites start from the needles of the $TiAl_3$, which appear partially dissolved. Thus the $TiAl_3$ crystals that form above the freezing point of the alloy react peritectically with the liquid and create nuclei for the solidification.



Fig. 23 — AlTi alloy, as-cast. Keller's etch. Dendrites forming from partly consumed $TiAl_3$ crystals. Mag. — 250X.

In the few examples in this paper is shown only a limited amount of information that can be obtained by the use of the microscope. In many other cases micro-examination can be used either alone or in conjunction and support of other inspection methods. However, a word of caution: micro-examination is a subjective method of inspection and relies wholly on an experienced metallographer. Most of the clues from which a decision has to be reached are elusive and seldom so sharp as to be foolproof. Also to be remembered: under the microscope only one section and usually a small one is examined, if the sample is not truly representative, it is easy to reach a completely wrong conclusion!

METALLOGRAPHY IN THE MAGNESIUM FOUNDRY

By

G. E. Holdeman*

Metallography is a valuable tool in the magnesium foundry for measuring the control of grain size, porosity, compound solution, fusion in heat treatment, precipitation and segregation. Use of this tool for the Mg-Al-Zn alloys as well as for the newer Mg-rare earth-zirconium and Mg-thorium-zirconium alloys will be discussed.

Preparation of samples for metallographic examination will be considered briefly. It should be emphasized that samples must be representative of the casting to be examined. Therefore, samples should be taken from both thick and thin sections. Areas of equal thickness may show different structures depending on the cooling rate. Heavy sections which have been drastically chilled may exhibit different compound size and distribution than sections of the same thickness which have cooled slowly, such as sections under risers. For equal grain-sized sections, compound distribution in permanent mold castings is finer than in sand castings. As a result, heat treatment of permanent mold castings is, in general, easier than for sand castings.

It is desirable to cut samples with the least amount of working, although for the examination of as cast grain sizes, some residual working may be advantageous. Cutting by means of a hacksaw is satisfactory. The use of a large bandsaw for cutting up a casting will result in a worked layer of 1/8 to 1/4 in. in depth. This may be removed by recutting with a hacksaw or by grinding. Samples should not be taken from a casting by means of a chisel. Not only will excessive working of the specimen be encountered, but actual rupturing between grains may be found.

For samples which are small, hard to hold, or on which edge study is to be made, mounting is desirable. Materials such as bakelite, lucite, or styron may be used. If a mounting press is not available, molten sulphur may be poured around the sample. The use of low melting point metals for mounts is not recommended because of the possibility of a galvanic effect during washing and etching.

*The Dow Chemical Co., Magnesium Dept., Midland, Mich.

While various grinding techniques may be used, the following has been found to be satisfactory. Disks of 50, 150, and 320 aloxite cloth and No. 0 emery paper are mounted on a double shaft motor as shown in Fig. 1. The disks rotate over oil pans which catch the dust keeping it from the operators face and also preventing fire hazards from the accumulation of fine dust. Grinding by hand on a plate glass may be done if preferred, but the speed advantage with motor-mounted disks is obvious. All grinding marks from the previous step should be eliminated and grinding prolonged somewhat to reduce the amount of cold work present. Rotating the sample 90 degrees between disks aids in determining when scratches from the previous operation have been removed. Edges of samples or mounts should be beveled on the last grinding operation to prevent cloth tearing on subsequent polishing.

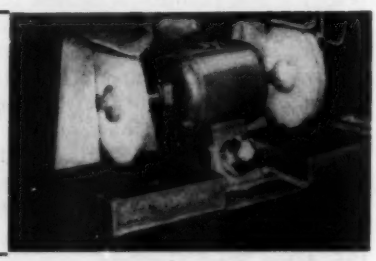


Fig. 1 — Grinding wheels for magnesium alloys.

Polishing is done on two wet laps using a medium nap polishing cloth such as cotton velveteen. A suspension of No. 600 alundum in water is used on the first wheel. The wheel is maintained just moist enough to prevent seizure. The specimen should be rotated counter to the direction of travel of the wheel and polishing should be done only long enough to remove scratches remaining from the No. 0 emery paper and to give a satiny finish. Prolonged polishing may give undesirable relief of some of the harder constituents. Speed of this wheel as well as that of the one used for final polishing is about 1000 rpm.

A suspension of levigated alumina or the newer synthetic gem materials may be used for final polishing. Filtered soap solution in the proportion of 30

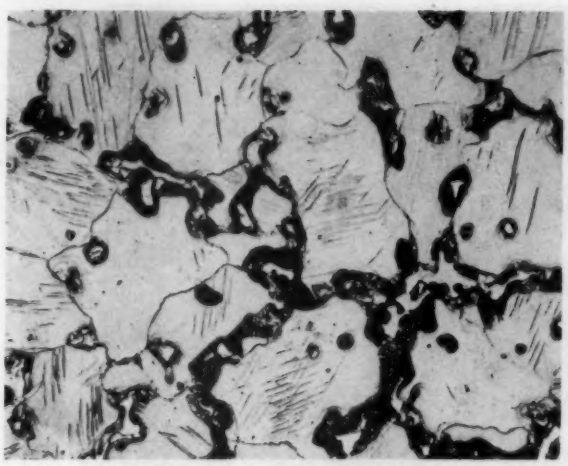


Fig. 2 — Grain boundaries in as-cast AZ92 alloy. Etch — Acetic picral. Mag. — 200X.

ml per liter is added to the suspension. Rotation counter to the direction of the wheel should also be followed in this operation. After the completion of polishing, the specimen should be rinsed first in running water, then in alcohol, and then dried by a blast of air. The specimen is now ready for etching.

The choice of etchants is dependent upon what the metallographer particularly wants to show. Of greatest interest to the foundryman are the acetic picral etchant used for estimating the amount of massive compound in the specimen, and the acetic glycol etchant which is used to show general structure and grain size in heat treated castings. Picral used in the picric acid etches may be made up and used as a stock solution. However, the picric acid etchants change their etching characteristics in a few hours and cannot be made up in advance. This is not true of the glycol type etchants. Etching is done by immersion in the case of the picral etchants and by swabbing with glycol-base etchants. Samples etched in picral etchants should be washed in running alcohol after etching rather than in water. With glycol etchants washing with running water is preferred.

It has been well established that grain size is important in developing maximum mechanical properties in magnesium alloys. Grain size is also important because of its effect on compound size and shape and ease of heat treatment. Likewise of importance is the effect of grain size on cracking tendency. Inasmuch as heavy sections of castings are particularly prone to show coarse grain, a special casting has been developed for checking the adequacy of grain refining operations. This is a 4-in. cube poured in an open dry-sand mold. The sample for metallographic examination is taken as near as possible to the center of the cube. Grain size in this casting is measured in the as-cast condition. For this an etchant consisting of 100 ml of 5 per cent picral, 10 ml of water and 5 ml of glacial acetic acid is used. Three characteristics of this etchant enter into its ability to reveal grains in as-cast magnesium alloys.

1. Grain boundaries are revealed between the massive compound particles.

2. Contrast is shown between one grain and another because of the etchants ability to stain grains differently depending on their orientation.

3. Twins are revealed that run in given directions in each grain. This produces still more contrast between grains.

Figure 2 shows the grain structure of as-cast AZ92 alloy as revealed by the acetic picral etchant. In obtaining grain size on cast material it is advantageous to leave some of the twinned metal produced by sawing on the specimen. This is generally easy to do as extended grinding and polishing are required to remove the cold-worked layer.

While massive compound often lies along grain boundaries as a semi-continuous network, this is not always true. Whenever prominent dendrites have formed, the massive compound may also be found along the dendrite interstices where it tends to give an impression of finer grain than is actually the case. The accuracy of grain size measurements may be increased if samples are first given a partial heat treatment to remove evidence of coring and then are cooled in air so that a line of precipitate occurs at the given boundaries.

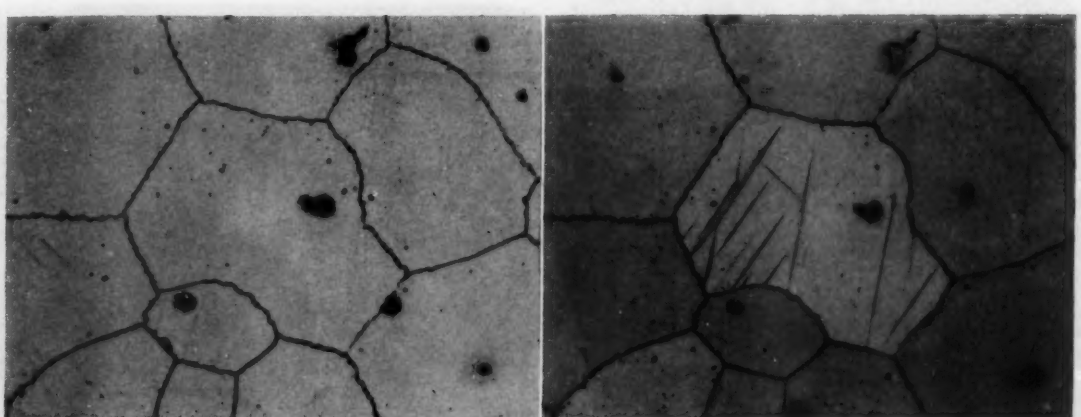


Fig. 3 — Heat-treated AZ92 alloy. Etch (Left) — Acetic glycol; (Right) — 100 picral, 10 water, 5 acetic acid. Mag. — 400X.

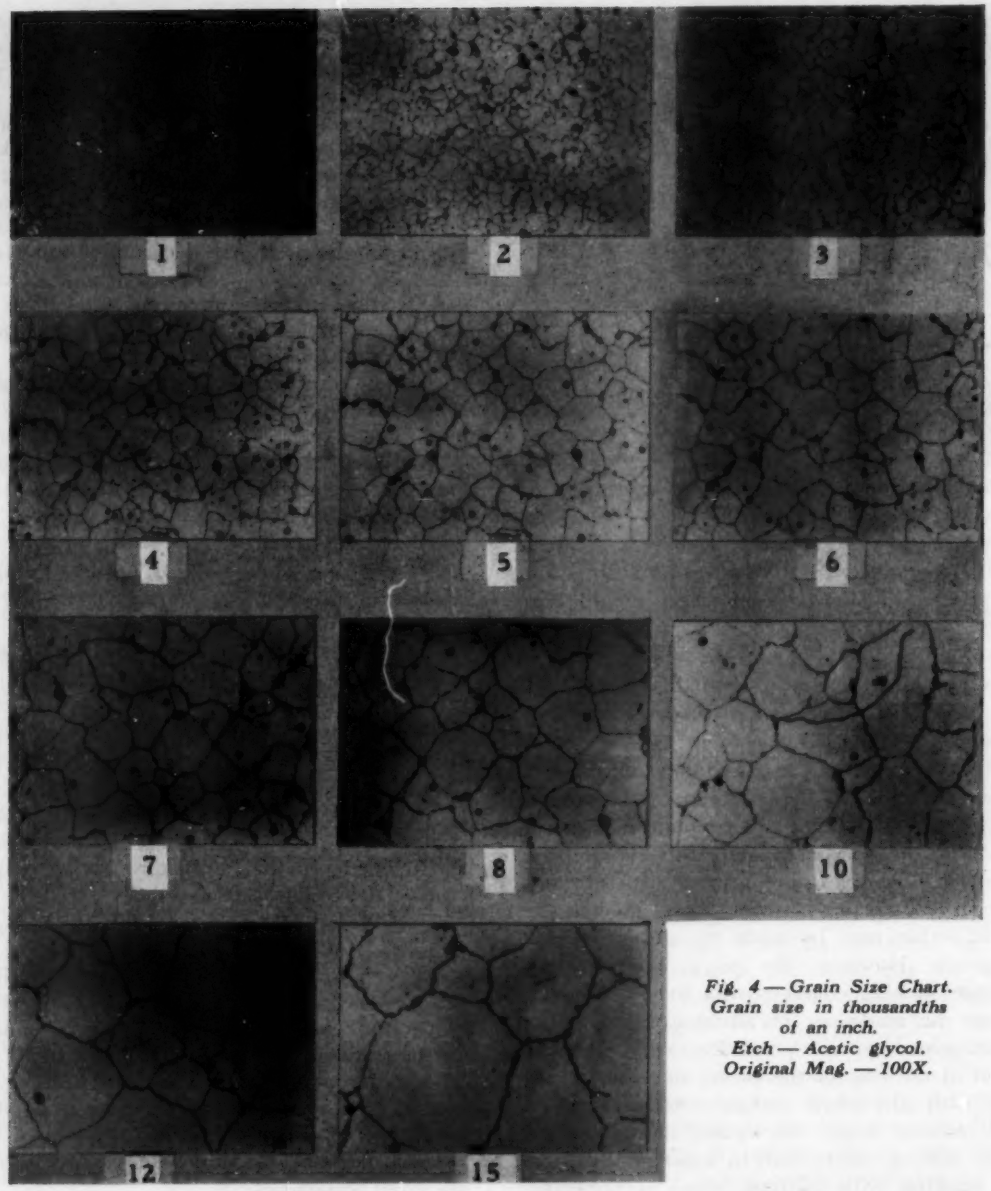


Fig. 4 — Grain Size Chart. Grain size in thousandths of an inch. Etch — Acetic glycol. Original Mag. — 100X.

For grain size measurements in the heat-treated condition either the acetic picral etchant (100 ml of 5 per cent picral, 10 ml of water, 5 ml of glacial acetic acid) or the acetic-glycol etchant (60 ml of ethylene glycol, 20 ml of glacial acetic acid, 19 ml of water, and 1 ml of concentrated HNO_3) may be used.

Figure 3 shows the same specimen of heat-treated AZ92 as etched with acetic glycol and with acetic picral. Twins may be seen in the center grain of the specimen as etched with acetic picral.

While a number of methods may be used for determining the grain size, a preferred method is to use the comparison with a grain size chart as shown in Fig. 4. Rapid determination of grain size is pos-

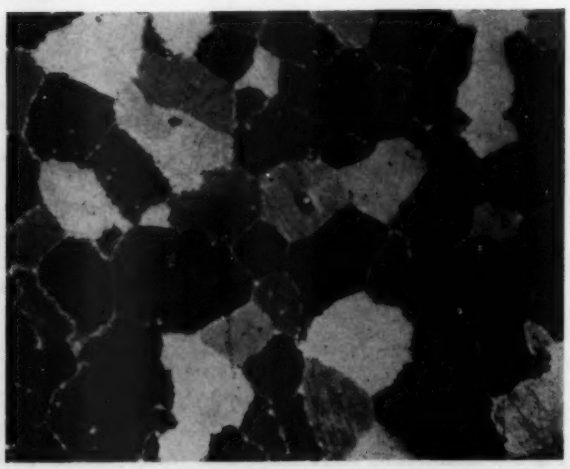


Fig. 5 — Heat-treated AZ92 alloy. Polarized light. Etch — Acetic picral. Mag. — 100X.



Fig. 6 — As-cast A-10 alloy. Etch — Glycol. Mag. — 500X.

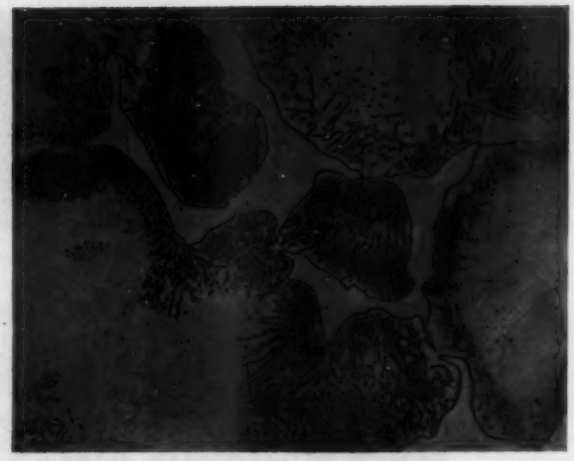


Fig. 7 — As-cast AZ63 alloy. Etch — Glycol. Mag. — 500X.

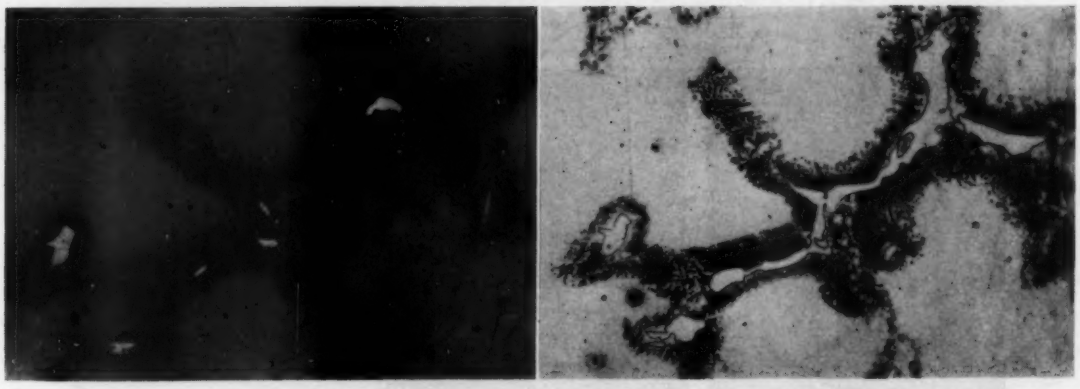


Fig. 8 — AZ63 alloy showing darkened $Mg_{17}Al_{12}$ and white $Mg_3Al_2Zn_3$. Etch (Left) — HF; (Right) — Glycol. Mag. — 200X.

sible by this method. Whole numbers on the chart indicate grain size in thousandths. Larger grain sizes may be determined by reducing the magnification to 25X and using a microscope with a revolving stage and polarized light. Comparison is again made with the chart and readings multiplied by four.

Figure 5 shows the contrast between grains due to orientation differences as seen with polarized light.

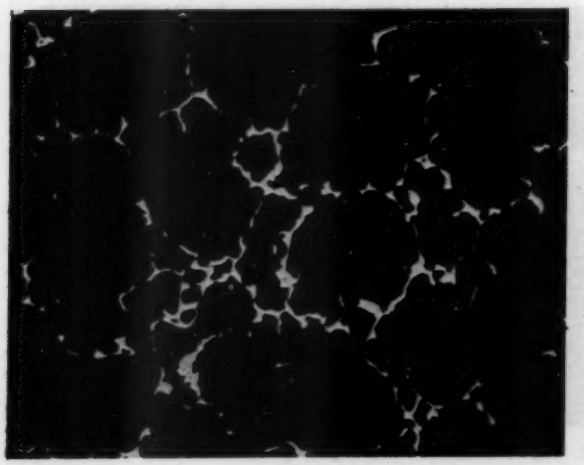


Fig. 9 — As-cast AZ92 alloy. Etch — Phospho-picral. Mag. 100X.

Also of considerable importance to the foundryman is the matter of massive compound after heat treatment in Mg-Al and Mg-Al-Zn alloys. Solution of the massive compound which is formed during solification results in marked increases in both tensile strength and elongation.

In Mg-Al alloys the massive compound is in the form of $Mg_{17}Al_{12}$ (Fig. 6). The addition of Zn results first in the formation of a divorced eutectic as shown in Fig. 7. If the amount of zinc present exceeds the ratio of 1 Zn to 3 Al, the compound $Mg_3Al_2Zn_3$ can be found in the as-cast state. By an etching technique which involves first an etch in a solution of 10 ml of 48 per cent hydrofluoric acid and 90 ml of water followed by an etch in 10 ml of 5 per cent picral plus 90 ml of water, the $Mg_3Al_2Zn_3$ compound may be separated from the $Mg_{17}Al_{12}$.

Figure 8 shows the same area of an AZ63 casting as etched with glycol etchant and the HF etchant followed by the picral water etchant. The HF stains the $Mg_{17}Al_{12}$ compound black; this leaves the

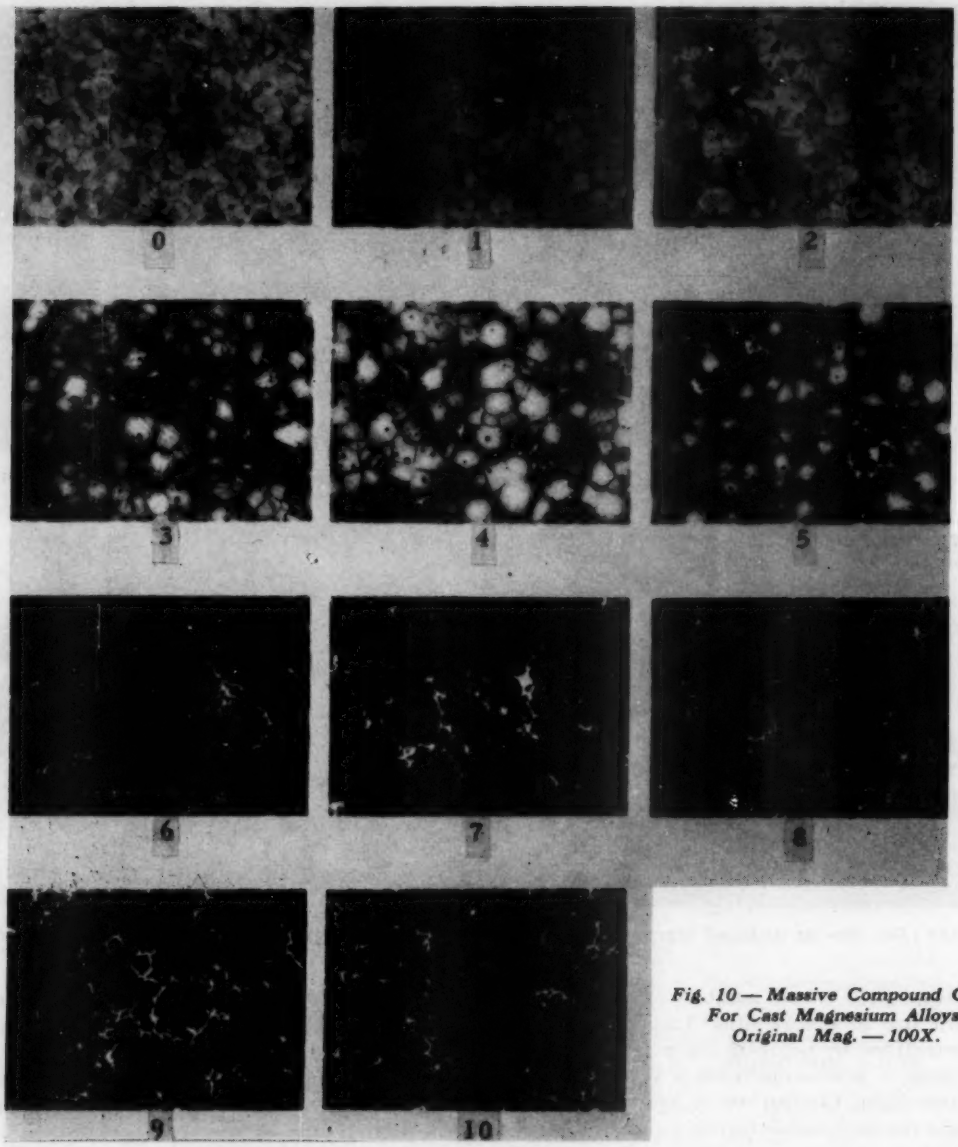


Fig. 10 — Massive Compound Chart For Cast Magnesium Alloys. Original Mag. — 100X.

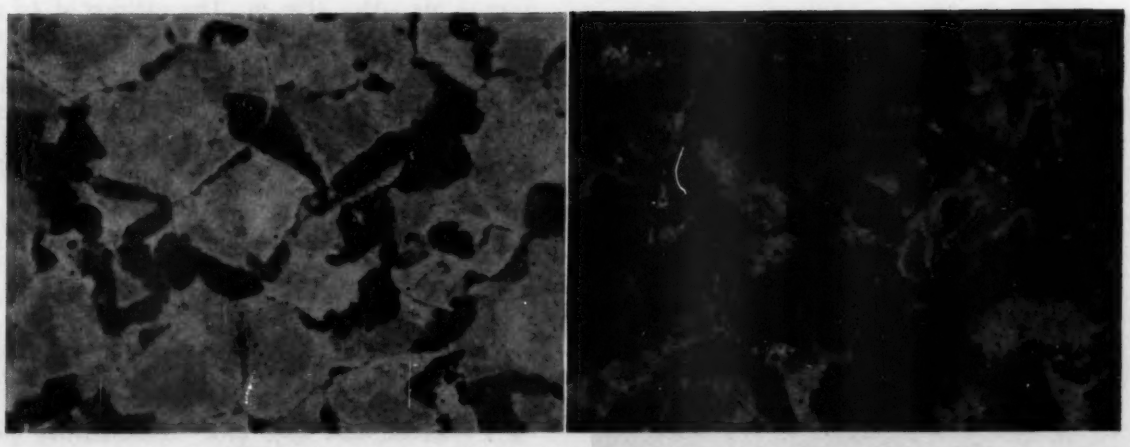


Fig. 11 — Precipitation in AZ63 (Left) and AZ92 (Right) alloys. Etch — Glycol. Mag. — 250X.

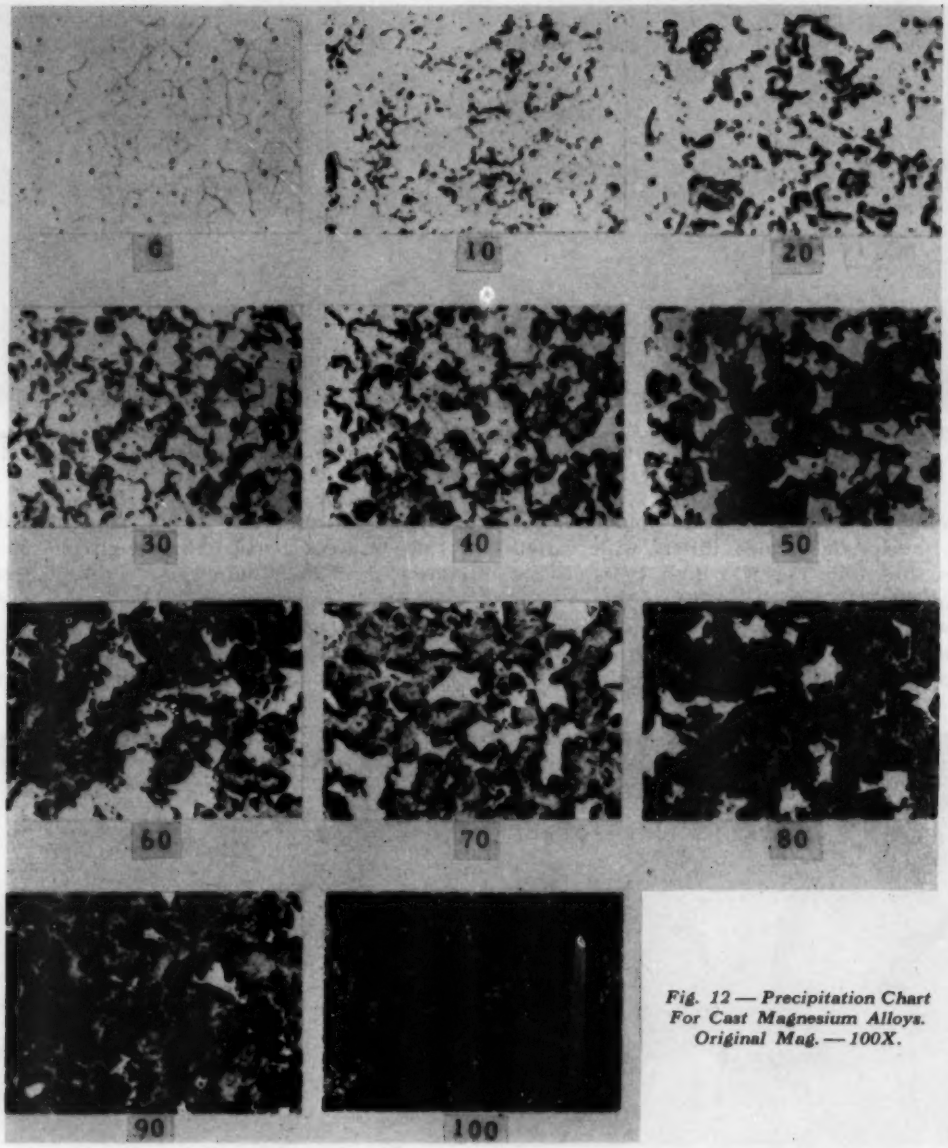


Fig. 12 — Precipitation Chart For Cast Magnesium Alloys. Original Mag. — 100X.

$Mg_3Al_2Zn_3$ unetched. The picral-water etch darkens the matrix to give better contrast between the matrix and the ternary compound. For normal foundry checking of massive compound, it is desirable to stain the matrix while leaving the compound white. This is best accomplished by the use of the phospho-picral etch which is composed of 0.7 ml of ortho-phosphoric acid and 100 ml of 5 per cent picral.

Figure 9 illustrates the stained matrix and white compound in a sample of AZ92 alloy. Massive compound may be assigned a numerical rating through the use of a chart as is shown in Fig. 10. Ratings vary from 0 where no massive compound is present to 10 for an as-cast structure of AZ92. Samples are etched in the phospho-picral etchant and the image projected on the screen of the microscope at 100X. The area in a 3-in. x 4-in. rectangle is compared with the chart. It is thus possible through the use of the compound chart to compare ratings not only between castings but between different areas in the same casting. While the lowest possible compound rating is desired for maximum property development, it is not possible to draw a definite line of satisfactory and non-satisfactory ratings. Other factors such as soundness and grain size must be considered in conjunction with the compound rating.

After solution heat treatment of a Mg-Al or Mg-Zn alloy has been effected to take the massive compound into solution, the properties may be further changed by an aging treatment. While tensile strength remains essentially unaffected, marked increases in the yield strength and hardness take place along with a reduction in elongation. Two types of precipitate are found in these alloys. The more familiar is the lamellar type resembling pearlite in appearance which forms along the grain boundaries. Another type is a finely divided precipitate which appears within the grain usually beginning at the center of the grain



Fig. 13 (Left) — Manganese constituent. Etch — None. Mag. — 500X. Fig. 14 (Center) — Typical appearance of Mg_2Si — included Mn particle. Etch — None. Mag. — 1000X. Fig. 15 (Right) — "Chinese Script" appearance of Mg_2Si. Etch — None. Mag. — 500X.

and proceeding outward to the grain boundary. The amount of precipitate formed differs with various alloys as is shown in Fig. 11. Both types of precipitate may be seen in the AZ63 specimen. The chart shown in Fig. 12 may be used for assigning ratings to the pearlitic type of precipitation. Samples before rating are etched in either the glycol (ethylene glycol – 75 ml, water 24 ml, concentrated HNO_3 – 1 ml) or acetic glycol (ethylene glycol – 60 ml, glacial acetic acid – 20 ml, water – 19 ml, concentrated HNO_3 – 1 ml) etchants. The ratings are based on the percentage of the area affected by the precipitate as seen under the microscope.

The elements Mn and Si are also found in Mg-Al



Fig. 16 — Microshrinkage (Left) and fusion voids (Right) in heat-treated AZ63 alloy. Etch — Glycol. Mag. — 150X.

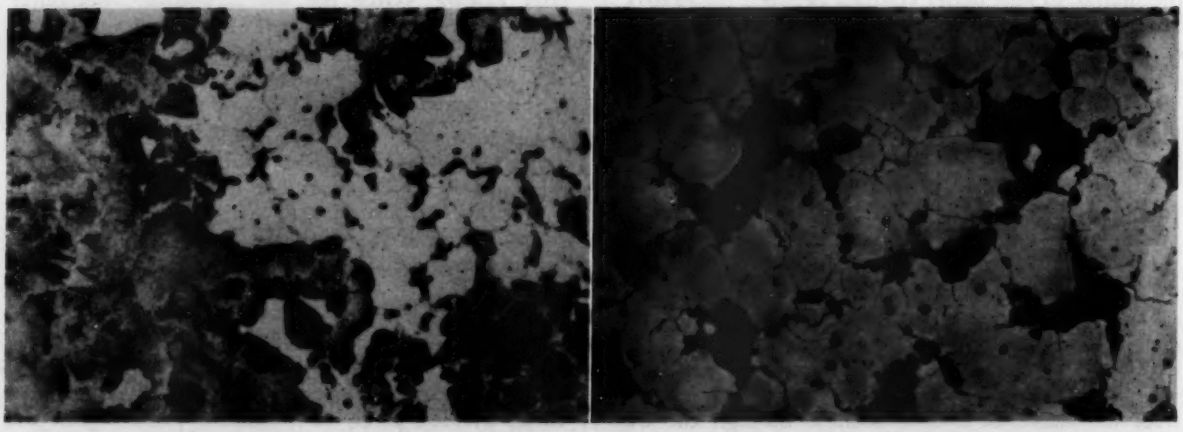


Fig. 17 — Microshrinkage (Left) and fusion voids (Right) in heat-treated AZ63 alloy. Etch — Glycol. Mag. — 100X.

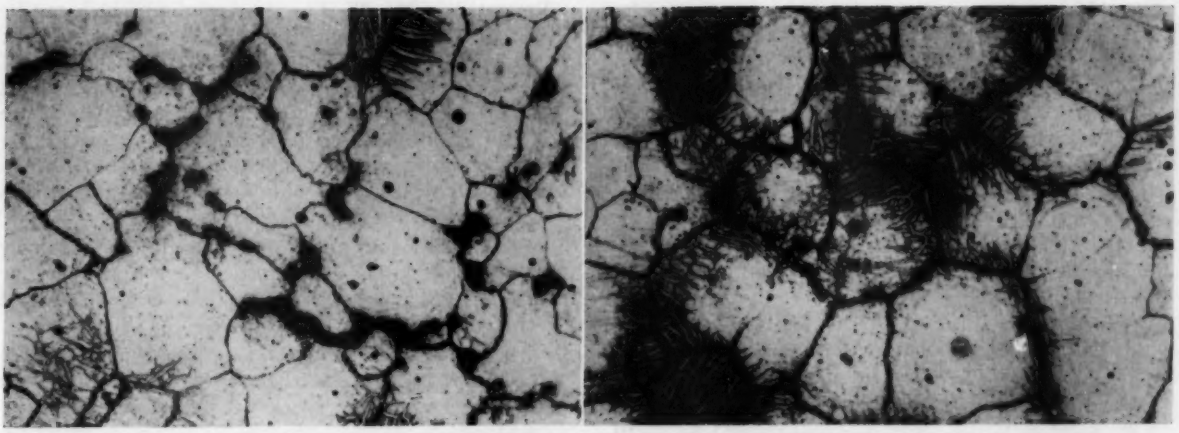


Fig. 18 — Microshrinkage (Left) and fusion voids (Right) in heat-treated AZ63 alloy. Etch — Acetic picral. Mag. — 100X.

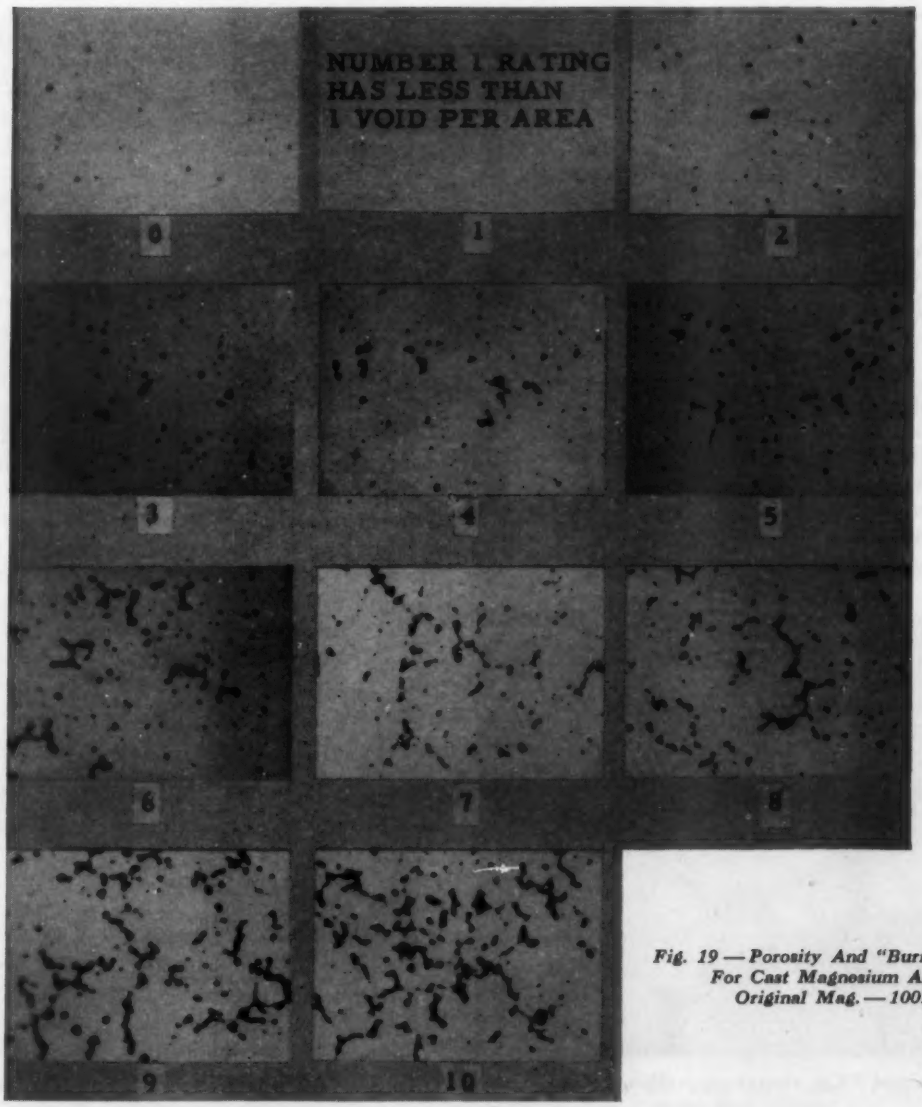


Fig. 19 — Porosity And "Burning" Chart For Cast Magnesium Alloys. Original Mag. — 100X.

alloys. Mn appears under the microscope as particles of Mn-Al with irregular contour and bluish gray color. The particles are in pronounced relief due to polishing and may have circle pits surrounding them. Mn particles are unaffected by etching but circle pits may be enlarged. Mn particles unetched are shown in Fig. 13.

Si occurs in Mg alloys as the intermetallic compound Mg_2Si. This is sometimes confused with Mn, but it differs from Mn in that it has a more watery-blue color, is more angular in shape, and shows less relief in polishing. The typical appearance of Mg_2Si is shown in Fig. 14. A Mn particle is included in one Mg_2Si particle. When the Si content exceeds about 0.15 per cent in Mg-Al alloys, the Mg_2Si assumes a more angular appearance referred to as "Chinese script" (Fig. 15.) An etch of 50 ml of 5 per cent picral and 50 ml of water gives greater contrast between Mg_2Si and Mn. The Mg_2Si becomes brilliant blue in color while the Mn remains a dull gray.

Microporosity and fusion voids both contribute to the lowering of mechanical properties in castings. Because their appearance under the microscope is quite similar, they will be discussed together. Distinguishing between microshrinkage which occurs during solidification and fusion voids or "burning" which are the result of improper heat treatment is often difficult. Three means which are useful in distinguishing these defects will be discussed. The shape of the void may be revealed by careful polishing. Fusion voids, produced by eutectic melting during heat treatment, have smooth concave sides with sharp points extending along the grain boundaries. Holes due to microshrinkage are irregular in outline (Fig. 16). A second means may be used if the casting has been aged or if some aging has occurred during cooling from the heat treating temperature. Areas in which microshrinkage has occurred will contain less precipitate than non-porous areas. On the other hand, areas around fusion voids will show the same or greater amount of precipitate than surrounding areas. The lack of precipitate around microshrinkage and the presence around fusion voids is illustrated in Fig. 17. The third method which can be used on AZ63 alloy involves the use of an etchant composed of 50 ml of 5 per cent picral, 20 ml of water, and 20 ml of glacial acetic acid. This etchant will produce a cracked film around voids which are due to fusion as the alloy content in this region is greater than the average composition of the alloy. No such cracked film will be formed around voids which are due to microshrinkage (Fig. 18).

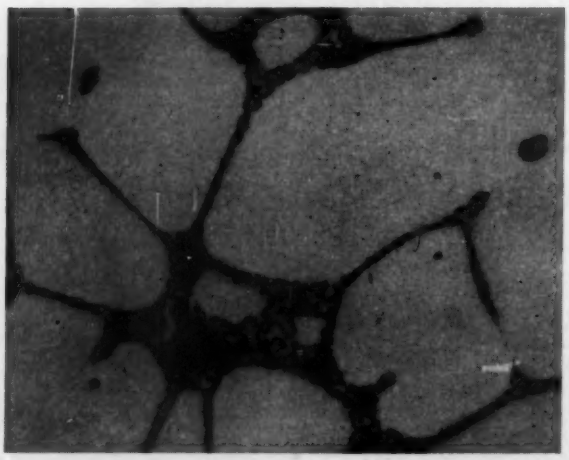


Fig. 20 — As-cast EK30 alloy. Etch — Glycol. Mag. — 500X.

For routine foundry metallographic work, it may be desirable to assign a rating to the porosity or fusion. This may be done by the use of a standard rating chart such as is shown in Fig. 19.

Magnesium-rare earth and magnesium-rare earth-zinc alloys are being used for aircraft castings where elevated temperature properties are required. Mischmetal which is about 50 per cent Ce is generally used as the alloying ingredient. Ce has limited solubility in Mg—about 1 per cent at 1070 F—and is found mostly as an eutectic phase at the grain boundaries. The compound Mg_9Ce is formed. Zr is added to the Mg-rare earth or Mg-rare earth-zinc alloys for grain refinement. It is not normally detectable in the as-cast state, but is developed upon heat treatment and aging as fine particles of zirconium hydride in the grain interiors. The as-cast structure of one of these alloys, EK30, is shown in Fig. 20. The Mg-Ce eutectic spheroidizes upon heat treatment so that the structure appears as in Fig. 21. Mg-rare earth alloys containing zinc may appear as in Fig. 22 which shows EZ33 in the T5 state. Here again compound is found at the grain boundaries, but it is believed to be a Mg-Ce-Zn compound. It is more massive and exhibits sharper points than does the compound in the EK30 alloy. Also the precipitation of zirconium hydride at the center of the grains is not present. The glycol type etchants are normally used with these alloys.

For higher temperature applications than are suitable for Mg-rare earth alloys, Mg-Th-Zn-Zr are used. The structure of HK31 alloy in the T6 state is shown in Fig. 23. The compound at the grain boundaries is a Mg-Th one which coalesces somewhat on heat treatment. No precipitation of zirconium hydride occurs within the grains. The structure of the Mg-Th-Zn-Zr alloy HZ32 commonly used in the T5 state appears quite different under the microscope. The Mg-Th-Zn compound present at the grain boundaries is in the form of sharp plates which probably contribute to the excellent creep properties of this alloy (Fig. 24).

If alloying and settling are not properly carried out in the Mg-rare earth or Mg-Th alloys, Zr segregation may be encountered such as is shown in Fig. 25. In spite of the metallographic appearance there is little effect on the mechanical properties because of such segregation.

The Mg-Zn-Zr alloys are used extensively in England and have some acceptance in this country for room temperature applications when high yield strength and good elongation are required. The alloy most commonly used in this country is ZK51A in

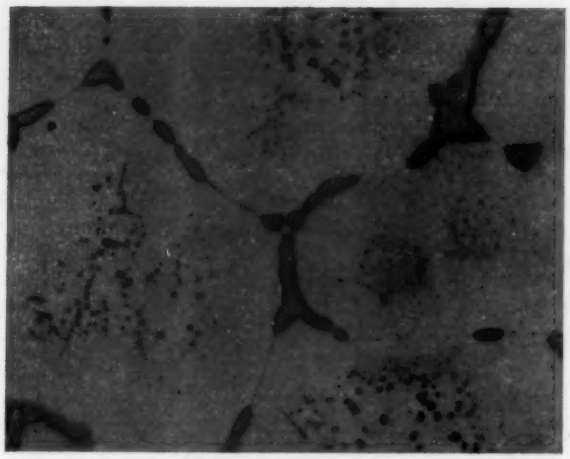


Fig. 21 — Heat-treated and aged EK30 alloy. Etch — Glycol. Mag. — 500X.

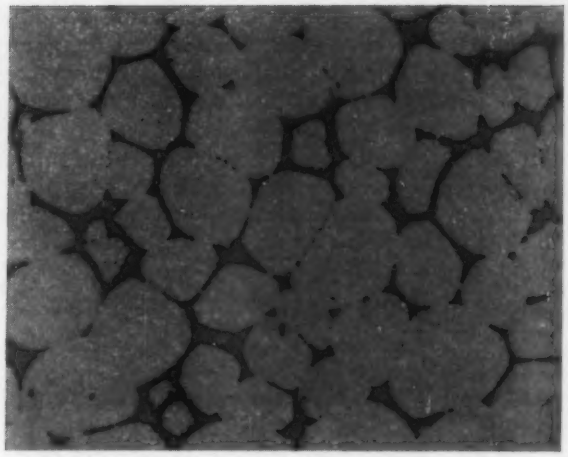


Fig. 22 — As-cast and aged EZ33 alloy. Etch — Glycol. Mag. — 250X.

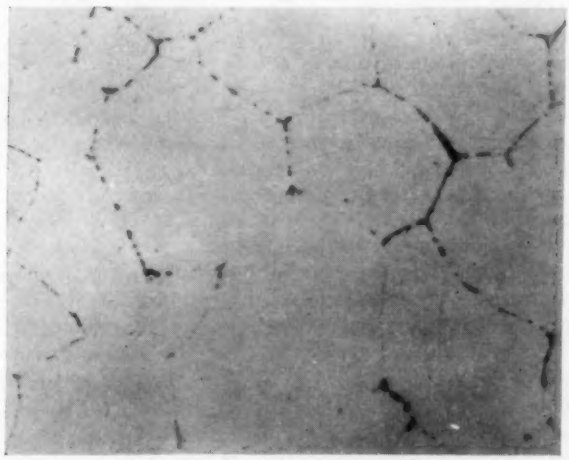


Fig. 23 — Heat-treated and aged HK31. Etch — None. Mag. — 500X.

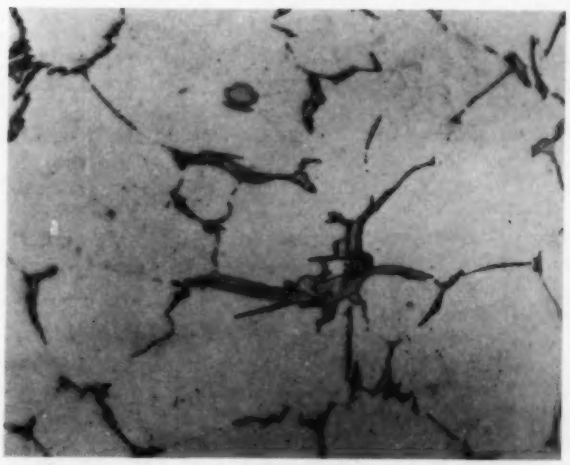


Fig. 24 — As-cast and aged HZ32 alloy. Etch — Glycol. — Mag. — 500X.

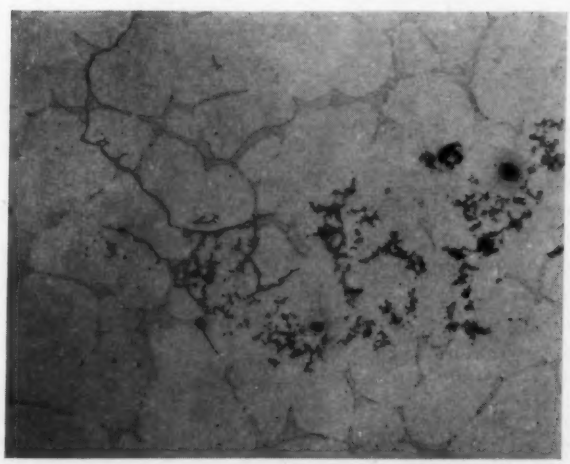


Fig. 25 — Zr segregation in EK30 alloy. Etch — Glycol. Mag. — 100X.

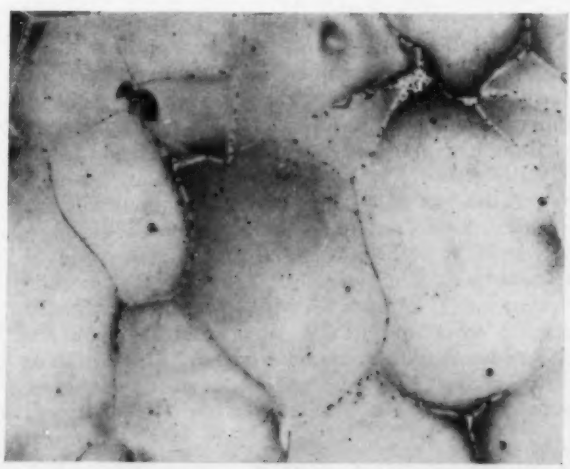


Fig. 26 — As-cast and aged ZK51 alloy. Etch — Phospho-picral. Mag. — 750X.

the as-cast and aged condition. Its structure is shown in Fig. 26. The white massive material at the grain boundaries is a Mg-Zn compound. Fog within the grains is caused by precipitated Mg-Zn and Zr-Zn compounds.

If weld repair is being done in the foundry, routine metallographic checks should be made of weld quality. Phospho-picral etchant has been found to be a suitable etchant for weld examination to check for proper fusion and cracks.

In our own magnesium foundry, metallography is a regular tool used in controlling procedures. Grain size measurements are made as previously mentioned on 4-in. cubes for control of grain refining operations. In addition, metallographic examinations are made on any test bars which fail to meet mechanical property requirements. Sections cut from heavy areas of scrap castings are placed in heat treat ovens and regularly checked for compound solution. A regular program of casting examination for grain size, porosity, fusion, compound solution, and precipitate is followed. These results are recorded and serve as a reference to compare past and present production.

FERROUS METALLOGRAPHIC PRACTICE

By

Philip C. Rosenthal*

The average foundry has at its disposal useful tools to study the shortcomings of metals or molding sand or to control these materials so that they will not deviate from specification requirements. Constant use of these tools by the foundry can prevent many metallurgical problems from getting started, and the occasional ones that do occur can frequently be solved by use of the scientific tools we have assembled for that purpose, such as the tensile test machines, chemical tests, spectrographic examination, x-ray inspection, magnetic particle inspection, metallographic inspection, etc. For best use of these tools, the foundryman must recognize their advantages and limitations, and in this present instance we are going to examine the advantages of one of the oldest and yet one of the most valuable tools available to the foundryman. This is the art of examining materials at a high enough magnification to reveal details that are not apparent to the naked eye.

In the case of the ferrous foundry alloys, to which we will devote our attention, metallography has long been an accepted method of inspection, but as in the case of many of these time-worn practices, its attributes can often be overlooked in the excitement of newer developments.

It is the author's intention to discuss first the use of metallography in cast steel metallurgy, then in malleable iron and finally in cast iron. This permits a logical introduction of the various metallographic constituents that need to be described.

The principal and most compelling reason we use metallography is to control the foundry product so that it meets specifications. It is because of the close association between structure and properties that metallography is so useful. Metallography is not a useful entity in itself, as a tensile test might be, for instance. Its usefulness lies in its ability to tell us something about metal properties that would be difficult or impossible to establish otherwise. In this paper, therefore, the relation between properties and structure will be stressed rather than the art of metallography, because it is the intent of this *Symposium* to show how metallography can work for the foundryman rather than making more work for him.

*Chairman, Dept. of Mining and Metallurgy, University of Wisconsin, Madison, Wis.

Metallography of Steel Castings

Metallography is useful in controlling those properties of cast steel that are influenced by the casting process and by its thermal history after casting.

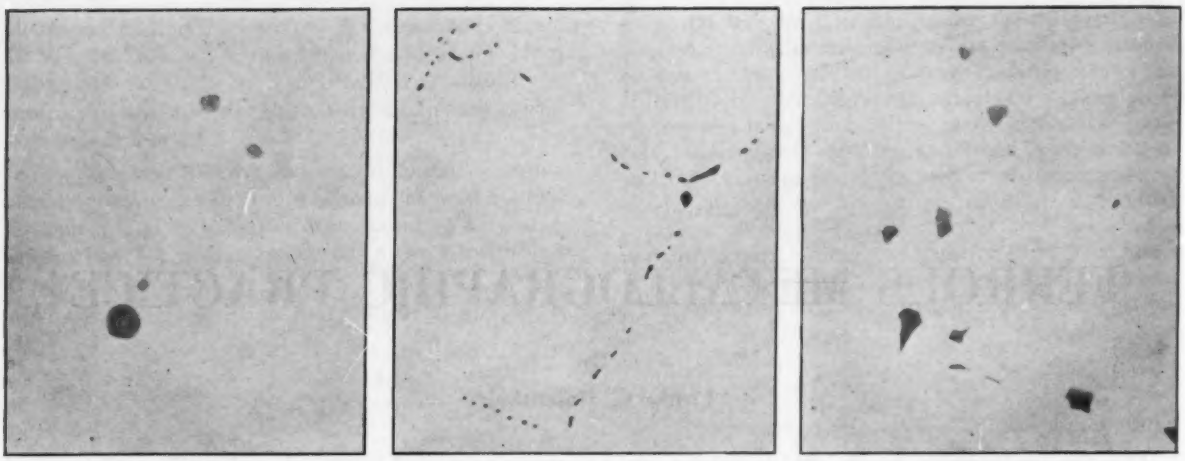
Problems that are related to the casting process include inclusion size and distribution, porosity, cracks, segregation effects, grain size, etc. These factors influence properties of the casting to a degree.

After the casting is poured, the properties produced by the heat treatment given the casting are best checked or controlled by metallographic means.

Inclusion Distribution

Metallographic inspection provides information regarding the deoxidation practice and also closely correlates with the properties obtained in the finished casting. Sims, Dahle, Gagnebin, Boulge, Saller and others have shown that the distribution of the sulfides in the steel is a fairly accurate reflection of the oxygen content of the molten metal at the time the casting is poured. The amount of oxygen in solution apparently affects the solubility of the sulfides, and this in turn changes the distribution of the sulfides by controlling the time during the solidification process at which the sulfides precipitate.

When the oxygen content is fairly high, as would be obtained by silicon or silicon-manganese deoxidation practice, the sulfides precipitate randomly as rounded globules indicative of a low solubility in the solidifying steel. Associated with the sulfides are rounded silicate inclusions. This type of distribution is shown in Fig. 1A and is referred to as Type I inclusions. The highest ductility and toughness results from this type of distribution. Consequently, the deoxidation practice leading to Type I inclusions is to be preferred except that it frequently is accompanied by pinholing near the surface which may demand that more severe deoxidation practice be used.



(A) Type I (B) Type II (C) Type III

Fig. 1 — Sulphide inclusions in cast steel[6]. Unetched. Mag. — 500X.

With a relatively small amount of aluminum as a deoxidizer, the sulfide distribution changes radically. The sulfides, rather than being randomly distributed, now are preferentially located at the grain boundaries of the freezing dendrites. On occasion, they may be severely elongated on the grain boundary, and are sometimes referred to as "stringer" or chain sulfides. The rounded silicates are not so prevalent or may be entirely absent. Clusters of fine alumina particles may be present. The localized distribution of the sulfides has a detrimental effect on ductility and toughness of the steel. An example of this type of distribution, referred to as Type II sulfides, is shown in Fig. 1B.

An addition of aluminum exceeding the critical amount which gives the Type II sulfides will cause a change in the sulfide pattern. Instead of being spread out in the grain boundaries, they now assume a more widespread distribution and are no longer elongated. This distribution is shown in Fig. 1C as the Type III sulfides. Although still segregated in the grain boundaries, the sulfides do not present the continuity found with the Type II sulfides and consequently ductility and toughness are not particularly harmed.

To a steel foundryman, the benefit gained from metallographic work that led to these important findings is quite obvious. Aluminum deoxidation is often necessary to avoid pinholing, yet it can cause low ductility. All that is now necessary to avoid pinholing and poor ductility is to use enough aluminum to produce Type III sulfides.

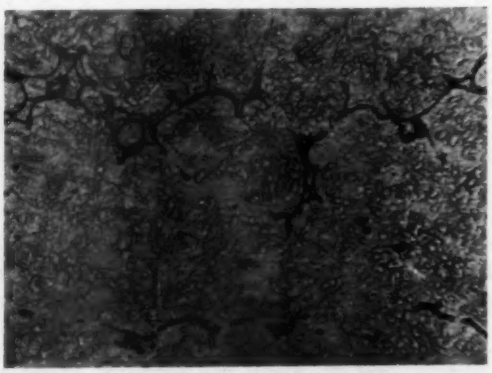
Other strong deoxidizers can act in a similar manner, but the amounts required to effect the changes from Type I to Type III sulfides will be different than that required of aluminum.

Naturally, metallography can also be used to determine the degree of cleanliness of a steel. This can be done by comparing representative photomicrographs of the unetched steel with suitable standards prepared for the purpose. The cleaner the steel, the greater will be its ductility, impact resistance and possibly also its endurance limit.

Other features of the casting that can be revealed in the unetched condition and which are associated with the freezing characteristics of the steel are pinholes near the surface; oxide penetration into the surface; cracks resulting from shrinkage effects; intergranular cracks from grain boundary precipitation; and other related defects. Figure 2, for example, shows part of a crack that developed in a steel casting. This crack is seen to show an interdendritic pattern. It can be assumed from this evidence that the crack originated during the freezing process rather than from stresses developing during subsequent cooling.

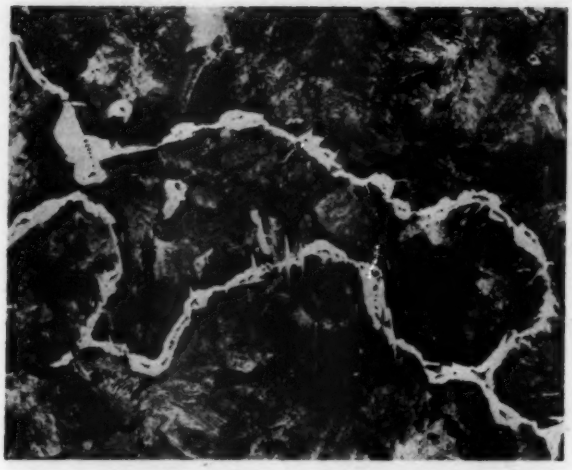
Intergranular cracking can result from aluminum nitride precipitation on the grain boundaries. Careful examination is required to reveal this precipitate since the particles are small and offer little contrast to the surrounding metal. Other unidentified inclusions have been found at times to contribute to intergranular-type failures.

In the etched condition, cast steel is usually composed of two microconstituents, ferrite and pearlite. Ferrite is the term given to the room temperature form of iron grains while pearlite represents a mixture of this ferrite with the compound iron carbide,



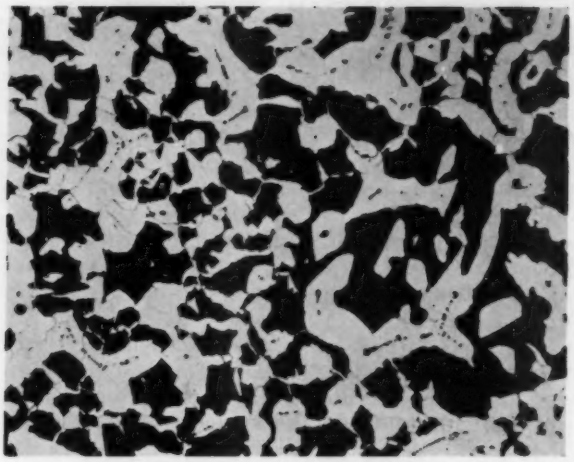
(Courtesy of Grede Foundries, Inc.)

Fig. 2 — Crack in a steel casting that developed as a result of shrinkage. Mag — 100X.



(Courtesy of R. A. Ragatz, University of Wisconsin)

Fig. 3 — Cast steel (0.42% C) showing ferrite (light), pearlite (dark) and inclusions in the ferrite. Mag. — 100X.



(Courtesy of R. A. Ragatz, University of Wisconsin)

Fig. 4 — The cast steel shown in Fig. 3 after annealing 3 hr at 1742 F. Mag. — 100X.

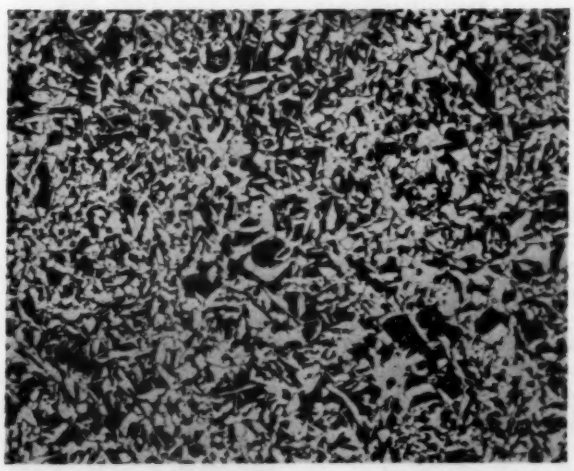
also designated cementite. Although this mixture actually occurs as alternate plates or laminations of iron carbide and iron, it normally appears as a dark phase when examined at relatively low magnification; consequently in an etched sample of cast steel, the dark areas correspond to pearlite whereas the light areas are ferrite. The proportion of these phases is controlled by the carbon content, although cooling rate from above about 1550 F is also important in determining the proportion of these constituents. A slowly cooled steel containing about 0.8 per cent carbon will show nothing but pearlite. Most cast steels have carbon contents less than this amount, however.

Because a certain degree of segregation of impurities and alloys develops during the freezing of steel, this condition influences the way ferrite and pearlite are distributed in the cast product. In fact, it is usually possible to distinguish between a cast steel and one that is heat treated. An example of a cast steel is shown in Fig. 3 which shows a number of interesting features. First, there is the typical coarse as-cast structure which is a reflection of the formation of the crystal structure during solidification. The dark-etching pearlite areas are outlined by a ferrite network. A network pattern of this type has been found to contribute to intergranular fracture exhibiting low ductility even though ferrite itself is ductile.[1] Presence of inclusions within the ferrite areas is also indicated. These inclusions were apparently precipitated during the last stages of freezing. And continuity in the inclusion pattern, as exhibited here, could also lead to an intergranular type of fracture in the steel.

If this steel is subsequently annealed, the coarse-grained structure will be broken up, but the inclusion pattern is not disturbed. An example of this is shown in Fig. 4.

Annealing infers furnace cooling from a temperature of 1600-1700 F. If the casting is cooled in air, the structure is further refined and improved, as revealed in Fig. 5. It would be expected that the refinement demonstrated by Fig. 5 would result in an improvement in mechanical properties.

It should be emphasized that the cast structure shown in Fig. 3 represents only one possibility in the distribution of pearlite and ferrite. Actually, a wide variety of distributions are obtainable depending on carbon content, alloy content and cooling rate. A few of these are illustrated in Fig. 6, showing a Widmanstatten structure, and Fig. 7, which shows an



(Courtesy of R. A. Ragatz, University of Wisconsin)

Fig. 5 — Cast steel (0.23% C) after holding at 1650 F for 1 hr and air cooling. Mag. — 100X.

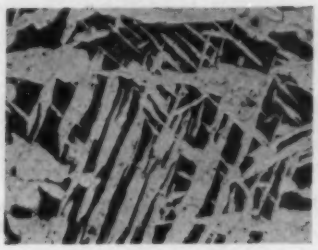


Fig. 6 — Widmanstatten structure in cast steel. Mag. — 100X.

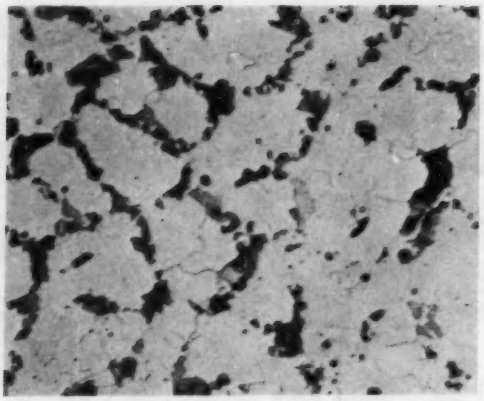


Fig. 7 — Interdendritic distribution of pearlite in cast steel. Mag. — 100X.

interdendritic distribution of pearlite in a low-carbon cast steel.

In alloy steels, the segregation pattern is not readily destroyed by heat treatment, and as seen in Fig. 8 it can occur in a variety of ways even after normalizing. These illustrations are for three grades of alloy steels normalized at 1600 F for 1 hr, 1800 F for 5 hr, and 2200 F for 19 hr, respectively, top to bottom. This evidence of persistent segregation led investigators to determine whether long-time, high-temperature annealing (often called "homogenizing") will eliminate segregation. Figure 8 shows that it is only partially eliminated even at a temperature as high as 2200 F.

In some photos in Fig. 8, the structure is not a combination of ferrite and pearlite, but a combination of fine pearlite and ferrite with martensite. This combination is shown again in Fig. 9. Either rapid cooling or alloy additions or both tend to promote formation of martensite.

Martensite is the phase that is desired when steel is heat treated by quenching. Because it is relatively brittle, it is then subsequently tempered to soften and

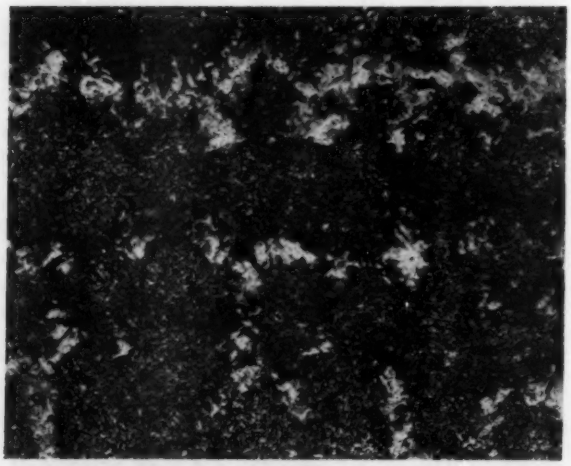
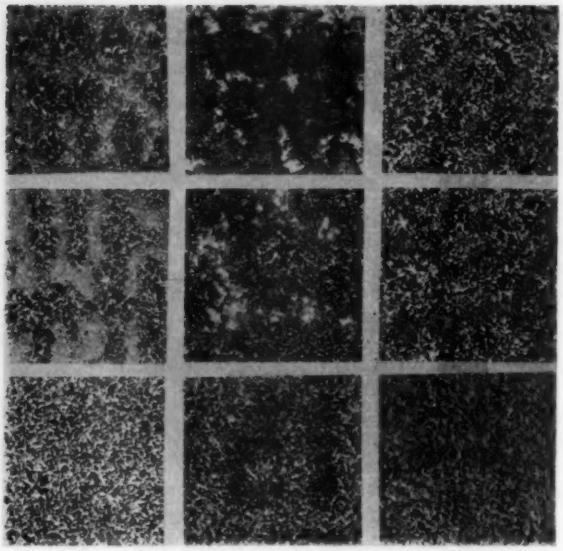


Fig. 9 — Combinations of fine pearlite and martensite in an alloy steel air cooled after 1 hr at 1600 F[7]. Mag. — 100X.



AISI 3140 AISI 4342 AISI 8642

Top Row — 1 hr at 1600 F
Center Row — 5 hr at 1800 F
Bottom Row — 19 hr at 2200 F

Fig. 8 — Effect of annealing time and temperature on microstructure of three alloy steels.[7] Original Mag. — 100X.

toughen the steel. Figure 10 shows that after a quench-and-temper heat treatment, the segregation pattern is not eliminated in an alloy nickel-chromium-molybdenum steel when it is homogenized at 1600 F for 1 hr even though the structure is all tempered martensite, but that it is fairly well eliminated after 19 hr at 2200 F. The markings are indentations made with a microhardness tester using a diamond indenter. The relative size of these indentations show differences in hardness in the quenched and tempered structure in the steel held at 1600 F.

Occasionally, the hardness of a quenched steel may not be as high as expected. Metallography can be used to establish the cause. Occurrence of some ferrite or pearlite along with the martensite will indicate that:

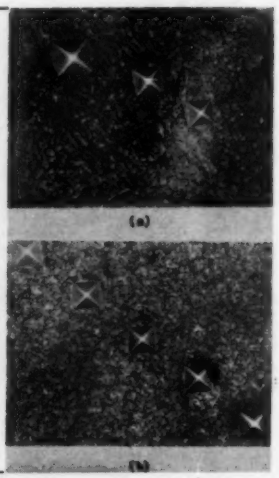


Fig. 10 — Microhardness tests indicate segregation in a Ni-Cr-Mo cast steel held at 1600 F for 1 hr (a), but its virtual elimination after 19 hr at 2200 F (b)[7]. Mag. — 100X.

1) the part was held out of the furnace too long before quenching—so called "slack quenching" or
2) the quenching rate was not high enough
3) alloy content was not high enough.

Often it is possible to determine which of these is the basic cause of the difficulty. Furthermore, the type of ferrite formed, whether it is of a "blocky" or "acicular" variety, will indicate to the experienced metallographer what alloying element can be added to alleviate the condition if lack of sufficient alloy is the cause.

Differences of opinion between a casting supplier and a customer concerning heat treatment can also

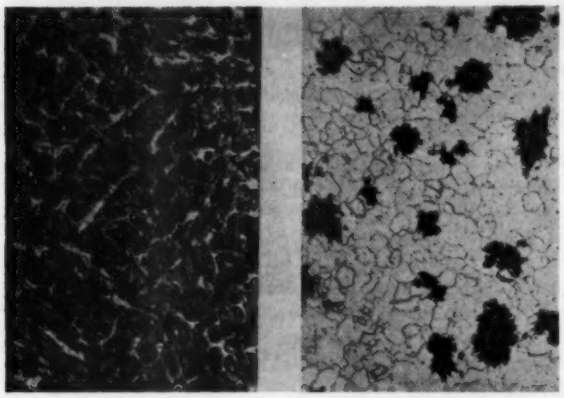


Fig. 11 — As-cast white iron (left) and malleable iron (right) resulting from annealing white iron. Mag. — 100X.

often be settled by metallographic examination. In one instance, cracks in a steel casting after quenching by the customer were blamed on poor normalizing treatment given by the casting producer. Metallographic inspection of the casting proved that ferrite appeared only in parts of the casting that were subject to slack quenching, whereas the balance of the casting showed the proper microstructure. Had the prior normalizing treatment been inadequate, any evidence of this deficiency would have been more or less uniformly distributed throughout the casting. Since the ferrite noticed in the structure came only in those parts of the casting that would most likely be slack quenched, the cracking difficulties therefore could be attributed to the quenching practice rather than the prior normalizing heat treatment.

In setting up heat treating processes for a specific casting, metallography can be used as a control tool by correlating the properties and microstructure obtained in test bars with the microstructures and consequently the properties obtained in critical sections of the casting.

Another instance of the use of metallography as a supplementary tool in the study of cast steel behavior is the paper by Christopher.[2] In this study of hot tearing of cast steel, Christopher verified metallographically that the drop in high temperature impact strength and occurrence of brittle fracture coincided with the formation of liquid metal at grain boundaries.

Metallography of Malleable Iron

Malleable iron is an alloy of iron, carbon, and silicon containing also manganese, phosphorus, and sulphur. The carbon content is in the neighborhood of 2.25 per cent. As-cast malleable iron is hard and brittle having all its carbon in the combined form, either as pearlite or as "massive" cementite (Fig. 11, left). This "white iron" is subsequently heat treated in a two-stage process to first decompose the massive cementite to graphite and second to cause the residual carbon in solution to precipitate to graphite rather than appear as pearlite, which is undesirable in plain malleable. Figure 11 (Right) shows a typical malleable iron after this annealing process. The structure is ferrite and graphite nodules.

In cases of poor annealability of malleable iron, metallography can indicate whether the difficulty arises from the first stage of the annealing cycle, the second stage, or results from the combined effects of both. If the cause is established as one or the other, corrective measures can be taken to either prolong the annealing cycle or check the analysis for residual elements that might be responsible since certain elements like chromium have a stabilizing influence in the carbides. Other elements like boron, when added in proper amounts, tend to counteract the influence of carbide stabilizers.

Conditions existing during melting of malleable can also be related to the microstructure. Excessive oxidizing conditions during the melting cycle can lead to irregular and large graphite nodules and to residual pearlite areas as shown in Fig. 12. Differences in nodule size can also be established metallographically and since these differences are believed to reflect prior processing variables, this distinction is sometimes important. Figure 13 shows the extremes that sometimes occur in this respect.

In addition to variations involving graphite nodules, the shape and distribution of the sulfides can also be used to give some idea of the degree of oxidation or deoxidation existing prior to pouring. These

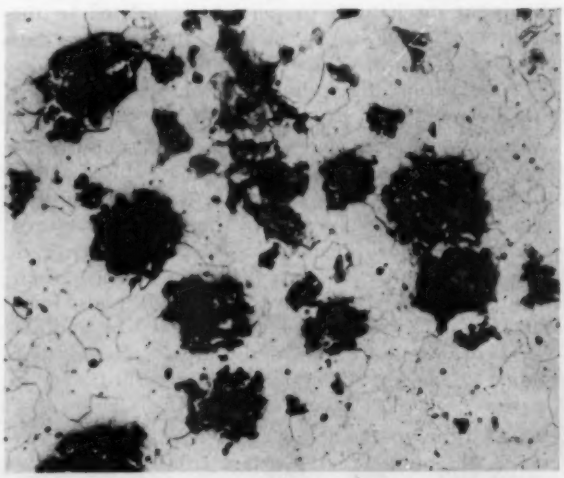


Fig. 12 — Large and irregular graphite nodules and residual pearlite patches suggestive of excessive oxidizing conditions during melting. Mag. — 100X.

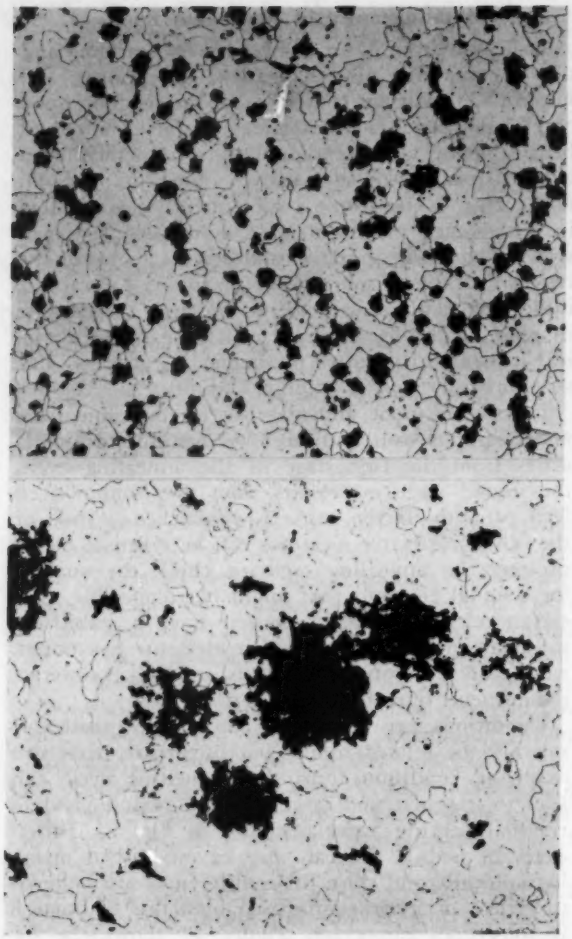


Fig. 13 — Extremes possible in graphite nodule size. Mag. — 100X.

differences are of the same order and affected in the same way as those for sulfides that appear in cast steel. Since the annealability of malleable iron and subsequent mechanical properties are related to the conditions existing during melting, these various bits of evidence furnished metallographically give valuable information that cannot be obtained readily by other means.

Just as in steel casting work, metallography offers a valuable tool for the laboratory in trouble-shooting problems. For example, a small casting after annealing was being straightened by hand and fractured with the first blow. Pearlite was suspected but metallographic examination revealed none. The brittle fracture was then attributed to a high phosphorus content which was verified by analysis. In another case, brittle fracture was definitely associated with residual pearlite and cementite as illustrated in Fig. 14.

In addition to these numerous applications, problems of pinholing, cracks, surface effects, etc. are also usefully handled by means of metallography. With pearlitic malleable iron becoming increasingly more important, it is obvious that here too the microscope



Fig. 14 — Residual combined carbon and pearlite in a malleable iron that had failed in a brittle manner. Mag. — 250X.

plays a very important role. All problems of heat treatment that occur in steel that are amenable to solution by metallographic inspection are also found in pearlitic malleable iron. In addition, features not encountered in steel must be kept under control. For example, excessive precipitation of carbon on the nodules results in a "bull's eye" structure which is undesirable. This structure is shown in Fig. 15. It comes from cooling too slowly through the transformation range of the iron.

Metallography of Gray and Nodular Iron

Despite the many applications of metallography in cast steel or malleable iron production, these seem almost insignificant when one turns to the field of cast iron to discover how metallography can be of benefit. This is due to the fact that cast iron is an exceedingly complex structure and that many of its outstanding properties are directly related to this structure.

Cast steel was shown to consist of various com-

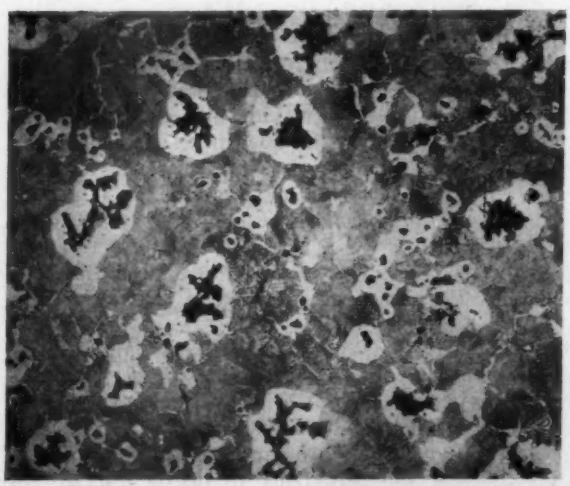


Fig. 15 — "Bull's eye" structure in pearlitic malleable. Mag. — 100X.

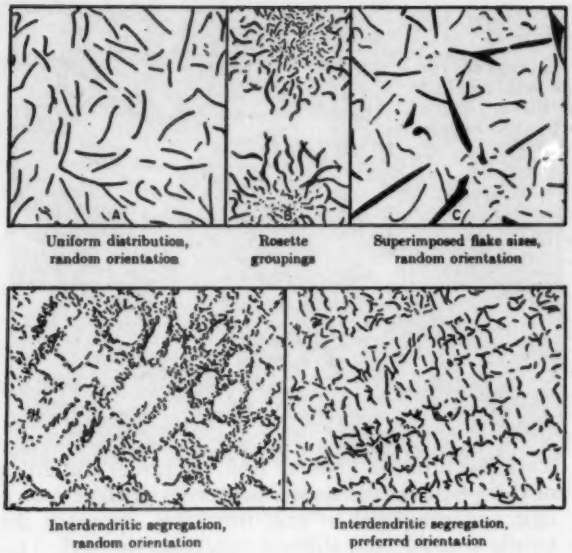


Fig. 16 — Graphite flake types[8].

binations of ferrite and pearlite. Gray cast iron has these variables plus others as well. For example, when cast iron freezes, it precipitates graphite in the form of flakes. Marked differences in the graphite flake structure and size can occur and these have an influence on the mechanical properties. Also, there usually is enough phosphorus present to cause another phase to appear which is not formed in steel or malleable iron. This phase is called "steadite". Gray iron is also sensitive to cooling rate. Consequently marked differences in structure can occur depending on the cooling rate of the casting or of the various parts of a single casting.

Figure 16 illustrates the five types of distribution of graphite flakes in gray iron. These five types of distribution have been standarized by AFS and ASTM. Type A is generally preferred for most irons. Rosette groupings such as shown in Type B reflect the eutectic freezing habit of gray iron. These are often

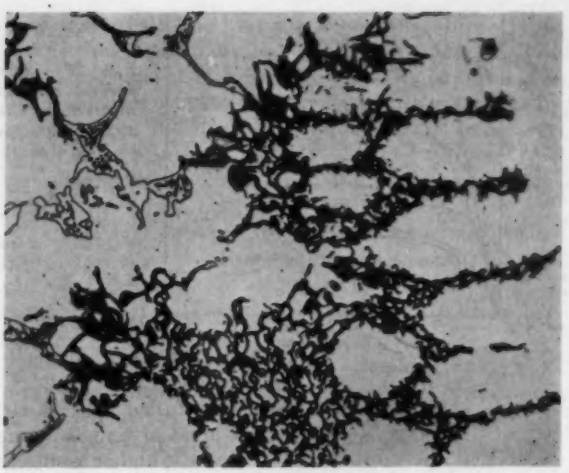


Fig. 18 — Type D graphite associated with ferrite found at a point just back of a chilled (white) iron area. Mag. — 500X.

found in irons of fairly high carbon content and near the surface of a casting where freezing rates are fairly high. Type C represents the type of structure resulting from a high carbon content. Type D occurs in irons that have been cooled rapidly or are in need of inoculation treatment. This type is assumed to occur from undercooling effects during solidification. Type E is usually found in irons of low carbon content.

It should be emphasized that all these graphite structures are essentially established at the completion of freezing. They cannot be altered by subsequent heat treatment.

Figure 17 shows the microstructure of a cylinder sleeve iron that produced a rough surface finish when machined. This condition can be attributed to the combination of Type B distribution and long graphite flakes which weakened the metal locally and which accounts for the two areas shown where metal particles were broken out during polishing.

Type D graphite associated with ferrite is shown in Fig. 18. This sample was taken from a wedge-shaped



Fig. 17 — Cast iron cylinder sleeve that produced a rough surface finish when machined. Mag. — 100X.

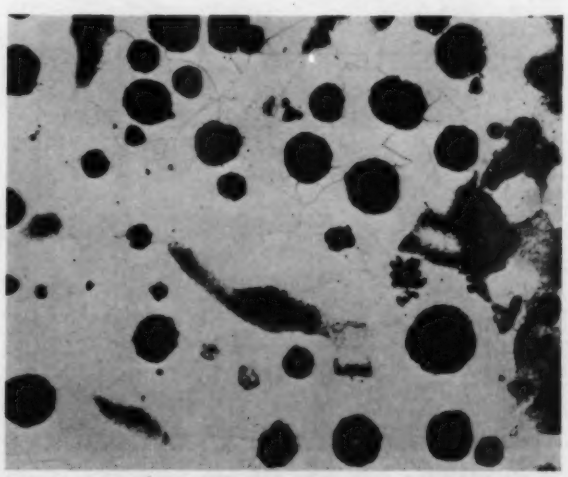
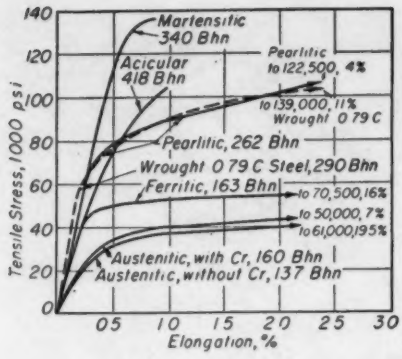
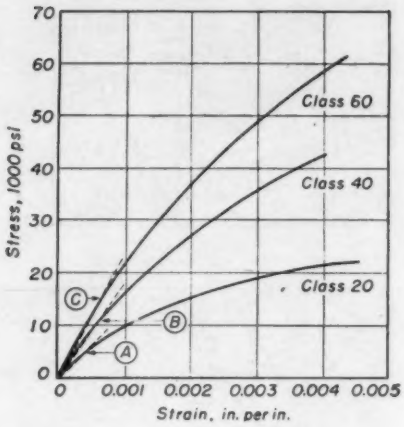


Fig. 19 — Nodular iron containing graphite nodules, pearlite patches and ferrite. Mag. — 100X.



(Left)

Fig. 20 — Stress-strain curves for several nodular irons and pearlitic malleable showing similar elastic strain characteristics (except for the austenitic irons)[9].



(Right)

Fig. 21 — Stress-strain curves for gray iron[10].

casting at a point just above the chilled zone in the tip of the wedge. In fact, a few areas of cementite are visible in the sample. At at slightly higher point in the wedge the graphite was coarser and associated with an essentially pearlitic matrix. Consequently, the sample shown represents a soft area sandwiched between a chilled portion on one side and a normal pearlitic structure on the other. This tendency for an abnormally soft area to occur immediately adjacent to a chilled area is not uncommon. The fine distribution of the graphite results from the relatively high rates of cooling, and is responsible for the breakdown of combined carbide despite the relatively high cooling rate. Since the combination of fine graphite and ferrite will wear excessively, it is obvious that this type of structure must be avoided in areas subjected to wear. Because the structure results from rather rapid cooling it is more likely to occur in thin sections.

Although Figs. 17 and 18 demonstrate marked differences in graphite size and shape, and consequently marked differences in properties and behavior in service, there perhaps is an even wider variation in properties when gray iron and nodular iron are compared. Figure 19 for example shows the structure of a nodular iron. The graphite particles here can be contrasted with those already observed in gray iron. This iron is essentially ferrite but a few pearlite areas are visible and a very few flake graphite particles also.

The essentially spherical distribution of graphite in malleable iron or nodular iron minimizes the effect of this phase on the tensile properties of these materials; and compared to gray iron, they approach the tensile strength and stiffness exhibited by steel. This is shown in Fig. 20 where the stress-strain curve for several nodular irons are compared to those for steel. In the first or elastic stage of loading, the curves almost coincide, and if the balance of the microstructure is comparable the curves are close together. Gray iron, on the other hand, with its graphite in a flake structure, will differ considerably from steel in its tensile behavior and the degree of variation is directly related to the graphite size and shape.

Stress-strain curves for three grades of cast iron are shown in Fig. 21. These illustrate the slope of the stress-strain curve is related to the graphite flake size and amount. Another way of demonstrating this difference between nodular and gray iron is shown in the curves prepared by Lindsay and Shanes[3] which show the change in properties as the graphite changes from its flake to the nodular variety (Fig. 22).

Thus we can conclude that the graphite structure of gray iron has an important bearing on its tensile properties aside from the effects resulting from differences in the amount of ferrite or pearlite.

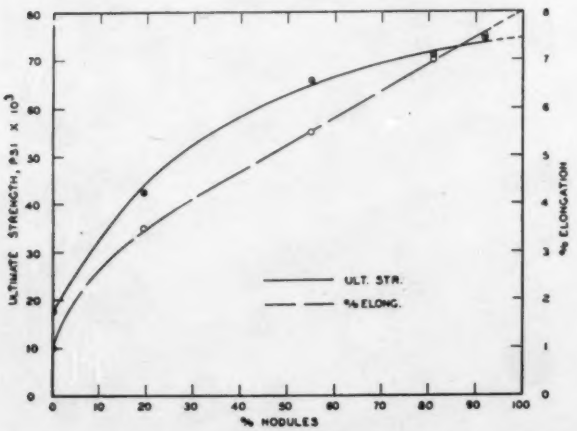


Fig. 22 — Effect of changing from flake to nodular graphite on strength and percentage elongation[3].

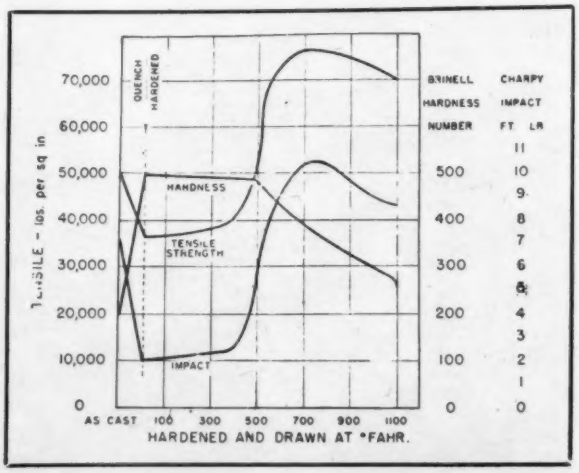


Fig. 23 — Effect of tempering quenched cast iron on hardness, tensile strength, and impact strength[4].



Fig. 24 — Excessive steadite (mottled areas) causes poor machinability. Mag. — 100X.

The complicating effect of graphite also shows up in heat-treated irons. In steel, heat treatment to a high hardness is accompanied by high tensile strength as well. Figure 23 from Austin,[4] however, shows that when cast iron is quenched and tempered, the tensile strength is moderately low until the iron becomes softer, after which the strength rises abruptly. The interpretation of this information is that the graphite flakes serve as notches in the hard iron causing it to fail prematurely. Only softening allows some plastic flow to adjust for this notching effect, and the strength to begin to rise.

In addition to the effects of the graphite, which influences such important properties as strength, hardness, machinability, wear resistance, damping capacity, etc., the structure of the metal also is subject to wide fluctuations and this must also be given consideration. Figure 24, for example, shows an iron of poor machinability which was attributed to the excessive amount of steadite in the structure. Use of the microscope readily established the cause of the difficulty.

Another cause of poor machinability is carbide, due to rapid cooling or the presence of alloying elements. This condition (Fig. 25)[5] is detrimental to machinability.

It is particularly in such properties as machinability or wear resistance of gray iron that microstructural features assume major importance. Free machining may be thought of as being achieved by having discontinuities in the metal to provide discontinuous chips and by having the metal conditioned by alloying or heat treatment so that it is readily cut by the tool. In cast iron, the discontinuities are furnished by the graphite flakes, and the metal conditioning comes from avoiding excessively hard phases or by deliberately promoting the formation of soft phases such as ferrite. A good grade of iron usually is mostly pearlite so far as the metallic phase is concerned, and this is readily machinable; but even better results are obtained by annealing to produce more ferrite. Thus, the change in microstructure shown in Fig. 26 is accompanied by more than a 100 per cent change in machinability. Of course, strength is sacrificed as annealing is used. The plus side of the machining problem is thus favored by having a ferritic matrix. The shape and size of the graphite are also important in determining machining rates and finish; but the actual extent of the effect of variations in size and shape have not been fully evaluated. Certainly, surface finish can be adversely affected by having an "open-grained" structure as was already demonstrated.

Factors in the machining of gray iron are chilled edges, excessive steadite, sand particles, abrasive inclusions, carbides, etc. Even differences in pearlite spacing makes a difference, coarser pearlite being better.

Under normal conditions, hardness can be used as a criterion of machinability, but the use of hardness measurements as an exclusive machinability control can be misleading. In one instance, machining a cast iron gear with a Bhn of 228, resulted in severe tool wear. Normalizing the casting improved the machinability but the hardness actually increased to 286 Brinell. Examination showed that the original material was a mixture of graphite, carbides, pearlite and ferrite, the soft ferrite providing the predominant influence on the hardness. Normalizing eliminated most of the carbides and ferrite areas, resulting in an essentially pearlitic structure and more machinable structure.

With respect to wear resistance, the requirements are almost the opposite of those for machinability. Ferrite in the matrix definitely impairs wear resistance. Type A graphite is preferred and Type D graphite definitely detrimental. Thus, control of the metallographic features of gray iron is important for machinability or wear resistance applications.

Many other applications of metallography to gray cast iron production could be cited and it is hoped that this brief review will give some inkling of the

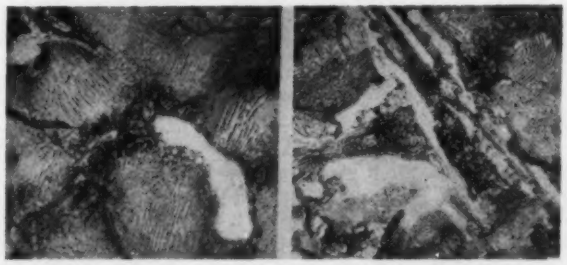


Fig. 25 — Massive carbide particles in gray iron[5].

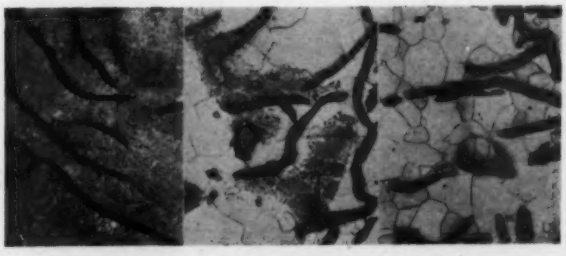
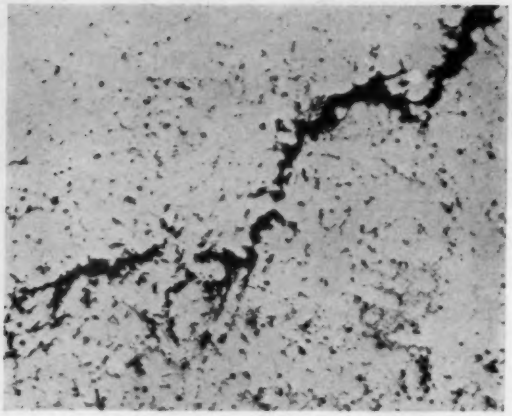


Fig. 26 — Annealing gray iron to produce a combination of ferrite and graphite causes more than a 100% improvement in machinability[11].



Fig. 27 — Interface between gray iron (bottom) and vitreous enamel (top). Mag. — 100X.

benefits derivable from metallography. Surface treatments such as enamelling can be checked to determine the nature of the interface and thus serve as a guide in the enamelling. Figure 27 shows such an interface. Cracks in castings can be studied to determine the cause. Thus the one in Fig. 28 was found to be a hot



(Courtesy of Grede Foundries, Inc.)

Fig. 28 — Hot tear in gray iron casting. Mag. — 100X.

tear rather than occurring in the annealing furnace as originally suspected. Many other examples of these and other types could be given to show how metallography is an important tool in the ferrous foundry.

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MANAGEMENT OF A METALLURGICAL LABORATORY

By

W. W. Edens*

A discussion of the management of a metallographic laboratory should be prefaced with a few remarks on the purpose and the use of this facility within any corporate structure. In the preceding discussions of this *Symposium,* various techniques and procedures have been presented which, collectively, can be considered to describe the work and general usefulness of a metallographic laboratory. Whether we deal with cast or wrought, non-ferrous or ferrous materials, a metallographic laboratory exists for the sole purpose of providing a service. The product of this service department is information.

To secure this information, we need not only cutting, grinding, polishing, etching, and microscope equipment, but we need also people to interpret intelligently what is seen under the microscope. This is necessary, so that the quality of our product, information, is true, accurate, and useful. This information becomes the basis for various operational decisions in the daily activity of the company and can have a pronounced effect on quality, scrap, and process sequence. All of these can materially affect the welfare of any company.

With this background regarding the product of a metallographic laboratory, it seems appropriate to consider its management under the general headings of operating procedure, equipment, personnel, and records. In order to carry on this discussion as directly as possible, the author will describe the operations at the research laboratories at his plant.

Operating Procedures

Work comes to the metallographic laboratory through any one of three main channels. The first of these are routine tests as might be applied to the inspection of incoming material. Such samples are submitted directly from the receiving department and could cover such things as grain size determinations, response to heat-treatment, and inspection of general metallurgical quality.

Figure 1 shows a number of furnaces used for heat-treatment tests. Approval must be given by the laboratory before such material can be released and moved to stock or processing departments. The results of the required tests are reported on a suitable report form and are directed to the inspection department who actually release the material for use.

The second source of metallographic samples comes from the processing or manufacturing departments. These may come as the result of current production problems or may be service failures that took place on test floors or in the field. In such types of work, an examination of the metallographic structure, in conjunction with chemical analysis and mechanical tests, can very frequently solve a production problem or point to desirable engineering changes that will improve a product. Information regarding such samples in the form of data and recommendations is sometimes reported in letter form but more often is released in the form of a formal laboratory report.

The third source of samples for the metallographic laboratory are those submitted as a part of various research projects within the research laboratories. In these projects, systematic studies are being carried

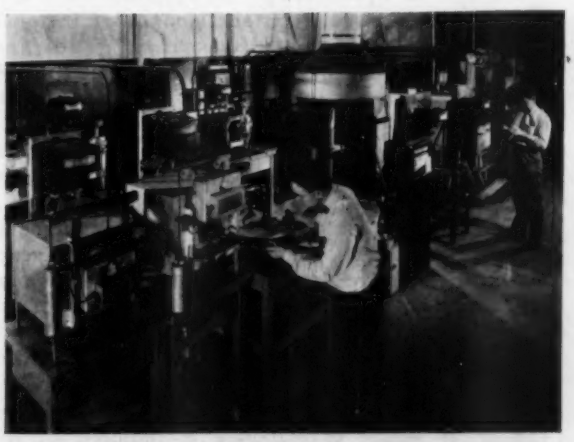


Fig. 1 — View of laboratory heat-treating furnaces

*Allis-Chalmers Manufacturing Co., Research Laboratories, Milwaukee, Wis.

out on a variety of subjects, and when the nature of the project is such that metallographic information is required, samples are submitted to the metallographic laboratory. The results are made available to the researcher, either for his personal visual observation under the microscope or as photomicrographs for his permanent record. The information thus becomes a part of a final formal report of a research project.

Regardless of the source of the metallographic sample, it must enter the laboratory through the central control desk. As a matter of fact, all samples, whether they be for metallographic examination, chemical analysis, physical test, or any other comparable determination, must go through this control desk. Here, the sample is permanently identified by a code and a suitable form is completed which gives all pertinent information regarding the history of the sample as well as information regarding the specific data that is required.

The identifying code consists of four alphabetical letters in sequence, beginning with AAAA and concluding with ZZZZ. The letters advance like the numerals of the mileage indicator of an automobile speedometer. Such a sequential code arrangement provides for over 400,000 separate identifications before the code has to be repeated. In this way, no more than four digits are ever necessary to give complete identification on the records or the actual sample.

One copy of the control sheet is kept at the central control desk and is used for expediting purposes within the laboratory. The other copies go to the laboratories concerned with making the necessary tests. After these tests have been performed in the various respective laboratories, the various data are returned to the central control desk on the original form. Here, the information is consolidated, typewritten onto a suitable printed form, and forwarded to the parties requesting the work. This, very briefly, covers the operational procedure.

Equipment

The specific equipment of the metallographic laboratory is subject to a certain amount of personal preference. Also, the particular character of the equipment will depend upon the type of work to be carried out in the facility. Our equipment, in general, includes the polishing equipment, various microscopes with magnifications up to 2500 and dark room equipment for developing, printing and enlarging photomicrographs.

Figure 2 shows some of the microscopic equipment. Our metallographic laboratory also handles the general photography requirements for the entire laboratory and photographs for permanent record, or report, any samples which can be readily moved into the metallographic laboratory. Also with the metallographic area, we have a Tukon hardness tester.

The equipment for a metallographic laboratory can be as simple or elaborate as required by the work which must be done. It is, however, very important and practically necessary that the area in which the microscopes are used, and in which the photographic work is done, be as free from dust as possible. Air conditioning is desirable, although not mandatory as long as the area can be kept free from dust.

To this end, the dry grinding and polishing portion of the metallographic laboratory should be in a separate room partitioned off from the area in which wet polishing is done and in which the microscopes are used. Also, the various dark rooms for developing and printing must be properly arranged for the exclusion of all unwanted light, yet maintaining adequate ventilation.

Water and drain facilities must be available in the dark room. Routine supplies such as polishing papers, etching chemicals, and photographic supplies must, of course, be purchased and maintained in suitable inventories. This generally describes the equipment requirements which must be supplied and maintained as a part of the management of a metallographic laboratory.

Personnel

The personnel required in a metallographic laboratory will vary as to number, type, and skill directly in accordance with the nature of the work and the amount of work coming into the laboratory. Both men and women can function equally well in this type of work, if appropriate instruction and training is given. In our operations, we distinguish between four general categories of personnel. These are routine technicians, specialized technicians, metallurgists, and research metallurgists. Under the general heading of metallurgists we have grades A, B, and C which reflect the degree of ability, training, and experience required in the respective jobs and, of course, this largely determines the salary rate.

In a one-man metallographic laboratory, the single individual does all the various types of work and can be sufficient unto himself. As the volume of work increases, division of labor can be accomplished in which routine technicians take care of the polishing and initial sample preparation. Specialized technicians take care of the general photographic requirements of the laboratory as well as to be extraordinarily proficient in other various particular phases of the work.



Fig. 2 — Microscopes and microhardness tester in the metallographic laboratory

The various grades of metallurgists have professional status and do work in such areas as special polishing and etching techniques, proper identification and interpretation of metallographic structures, and such general activities as require metallurgical knowledge and experience. These people make metallurgical decisions and formulate recommendations as to proper metallurgical procedures.

The research metallurgists work in their own fields and are not normally a part of the metallographic laboratory. The technicians perform the services of polishing, etching, and carry out any other preparation required, to the point where the research man can view the sample in accordance with the needs of his projects or specific interest. He can request photomicrographs as needed. His counsel and advice is available to the metallographic laboratory staff.

It is obvious that the division of work within any metallographic laboratory will depend upon the nature of the work therein, but where the operation is sufficiently large to warrant the use of several people, a distribution of labor as is indicated above can be made to function quite nicely.

Records

In the management of any staff or line organization, it is necessary that some records be kept. These can be only the simple payroll records which reflect the hours on the job. The amount and type of records kept can be elaborated from this to practically any degree considered desirable.

Although our people are salaried, we ask them to fill out two time cards every day. One is an IBM type card, which goes to the timekeeping and accounting departments, and shows the total time to be charged to other departments outside the laboratories. Most of the time shown on this card is spent on routine tests and this time is charged to appropriate inspection accounts and production orders as a legitimate cost of production.

The second time card is a research laboratories card. This has no official status in the sense that it affects the payroll nor are any charges derived from its use. This is purely informative within the laboratories. The headings to which time can be charged are relatively few in number. Such headings as laboratory investigations, research projects and consulting are typical. Under consulting, there is a subclassification which distinguishes the product line or product department which benefited from the consultation.

The various laboratory investigations and research projects are individually numbered and time is segregated and accumulated to these numbers. Through the use of this divisional time card, we accumulate in detail, monthly and annually, a division of time as it has been applied against the various major headings and the individual investigations and projects. These data are valuable in the overall direction of the work and in describing our work to top management.

Records are also kept of the total number of metallographic samples processed and, for that matter, this type of record is also kept by the people performing physical tests and chemical analyses to the end that numerical data are available regarding the activities of various service sections of the laboratories. As a matter of interest, there are over 1000 individual tests conducted per day in our laboratories by the various service sections.

All principles of good management naturally apply to a metallographic laboratory. Inadequate equipment can be just as wasteful as excessive personnel. Each laboratory has its own problems and there is no one single standard formula that will work in all cases.

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